

# TRANSACTIONS

of the

## American Foundrymen's Society



Proceedings of the  
Fifty-Second Annual Meeting  
Philadelphia, Pa.

May 3 to 7, 1948

VOLUME 56

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AMERICAN FOUNDRYMEN'S SOCIETY  
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In accordance with the action of the Board of Directors and the affirmative letter balloting of the eligible membership of the Society, the name of the American Foundrymen's Association has been changed to American Foundrymen's Society, effective as of July 1, 1948.

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of the

AMERICAN FOUNDRYMEN'S SOCIETY  
(INCORPORATED)

(Fiscal Year Ending June 30, 1949)

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Technical Director, S. C. MASSARI

\* Member, Executive Committee.

# SUMMARY OF PROCEEDINGS OF THE 52nd ANNUAL MEETING

THE 52ND ANNUAL CONVENTION of the American Foundrymen's Society was held in Philadelphia, May 3-7, 1948. The Biennial Foundry and Allied Industries Show was staged by the Society in conjunction with the convention. Exhibits were displayed in Convention Hall.

Credit for the cordial reception and hospitality received by those in attendance is given to the Philadelphia Chapter of A.F.S. under the general chairmanship of John M. Robb, Jr., Hickman, Williams & Co., Inc. This committee excelled in its splendid work and hospitality as host to the large number of members and guests attending.

This exhibit broke previous records from the standpoint of number and quality of exhibits. Over 270 companies exhibited. Over 13,000 foundrymen, metallurgists, and representatives from allied industries throughout the United States and from 17 foreign countries attended the convention. During the five days of the meeting, 50 sessions were held for presentation of about 100 timely technical papers and committee reports. A series of five lectures on Foundry Quality Control was featured.

A summary of the sessions which were held in the meeting rooms in Convention Hall, Bellevue-Stratford Hotel, and the Benjamin Franklin Hotel follows:

## ALUMINUM AND MAGNESIUM SESSION

Monday, May 3, 10:00 A.M.

*Presiding*—L. Brown, Magnesium Fabricators Div., Bohn Aluminum and Brass Corp., Adrian, Mich.

*Co-Chairman*—A. T. Ruppe, Bendix Products Div., Bendix Aviation Corp., South Bend, Ind.

*Effect of Gating Design on Metal Flow Conditions in Casting of Magnesium Alloys*—H. E. Elliott, Dow Chemical Co., Bay City, Mich., and J. G. Mezoff, Saginaw Bay Industries Inc., Bay City, Mich.

*A Study of Factors Affecting the Pouring Rates of Castings*—H. E. Elliott, Dow Chemical Co., Bay City, Mich., and J. G. Mezoff, Saginaw Bay Industries, Bay City, Mich.

*Step Aging of a Magnesium-Base Casting Alloy*—E. J. Vargo, Wellman Bronze and Aluminum Co., Cleveland, and G. Sachs, Case Institute of Technology, Cleveland.

## EDUCATIONAL SESSION

Monday, May 3, 10:00 A.M.

*Presiding*—W. H. Ruten, Polytechnic Institute of Brooklyn, Brooklyn.

*Co-Chairman*—L. G. Probst, National Engineering Co., Chicago.

*Engineering Education for the Foundry Industry*—G. K. Dreher, Foundry Educational Foundation, Cleveland.

*Recruiting and Training Panel*—G. J. Barker, University of Wisconsin, Madison, Wis.; I. Ivan, Whiting Corp., Harvey, Ill.; Glenn Kies, Industrial Training Service, Detroit; H. L. Rosse, Eddystone Borough Schools, Pa.; F. B. Skeates, Link Belt Co., Chicago; R. S. Tour, University of Cincinnati, Cincinnati.

## MALLEABLE SESSION

Monday, May 3, 10:00 A.M.

*Presiding*—W. D. McMillan, International Harvester Co., Chicago.

*Co-Chairman*—W. B. McFerrin, Electro Metallurgical Co., Detroit.

*Effect of the Common Alloying Elements on the Tensile Properties of Malleable Iron*—H. A. Schwartz and W. K. Bock, National Malleable & Steel Castings Co., Cleveland.

*The Construction and Operation of an Oil Fired Malleable Iron Holding Furnace*—F. Coghlin, Jr., Albion Malleable Iron Co., Albion, Mich.

*Influence of Chromium on the Graphitization of White Cast Iron*—Gabriel Joly, Technical Center of Foundry Industries, Paris, France—*French Exchange Paper*, Presented by C. O. Burgess, Formerly Union Carbide & Carbon Research Labs., Niagara Falls, N. Y., Now with Gray Iron Founders' Society, Cleveland.

## BRASS AND BRONZE ROUND TABLE LUNCHEON

Monday, May 3, 12:00 Noon

*Presiding*—B. A. Miller, Baldwin Locomotive Works, Philadelphia.

*Co-Chairman*—R. J. Keeley, Ajax Metals, Philadelphia.

*Subject*—Recent Developments in the Melting of Brass and Bronze.

*Discussion Leader*—H. L. Smith, Federated Metals Div., American Smelting & Refining Co., Pittsburgh.

## ALUMINUM AND MAGNESIUM SESSION

Monday, May 3, 2:00 P.M.

*Presiding*—C. E. Nelson, Dow Chemical Co., Midland, Mich.

*Co-Chairman*—M. H. Young, Wright Aeronautical Corp., Wood Ridge, N. J.

*Can Castings Be Engineered?*—F. G. Tatnall, Baldwin Locomotive Works, Eddystone, Pa.

*Engineering For Aluminum Alloy Castings*—T. G. Gauthier and H. J. Rowe, Aluminum Co. of America, Cleveland.

## MALLEABLE SESSION

Monday, May 3, 2:00 P.M.

*Presiding*—C. F. Joseph, Central Foundry Div., General Motors Corp., Saginaw, Mich.

*Co-Chairman*—Eric Weland, Union Malleable Iron Works, Deere & Co., East Moline, Ill.

*Application of Correlation in the Malleable Iron Foundry*—R. G. Seidel, National Malleable & Steel Castings Co., Cleveland.

*Production Hardness Testing in a Malleable Shop*—C. Schneider and L. Ulsenheimer, National Malleable & Steel Castings Co., Cleveland.

## BRASS AND BRONZE SESSION

Monday, May 3, 4:00 P.M.

*Presiding*—H. M. St. John, Crane Co., Chicago.

*Co-Chairman*—J. J. Curran, Walworth Co., Greensburg, Pa.

*An Approach to Quality Control in Casting Production*—G. K. Eggleston and V. A. Simpson, Barnes Mfg. Co., Mansfield, Ohio.

*Ingot Metal vs. Virgin Metal*—F. L. Wolf, Mansfield, Ohio.

## LECTURE COURSE SESSION

Monday, May 3, 4:00 P.M.

*Presiding*—M. E. Brooks, Dow Chemical Co., Bay City, Mich.

*Co-Chairman*—D. Basch, Almin. Ltd. of Great Britain, Schenectady.

*Subject*—Test Procedures for Quality Control of Aluminum and Magnesium Castings.



Discussion Leader—E. V. Blackmun, Aluminum Co. of America, Cleveland.

#### EDUCATIONAL DINNER

Monday, May 3, 6:30 P.M.

Presiding—A. W. Gregg, Whiting Corporation, Harvey, Ill.  
Co-Chairman—D. F. Lane, Bethlehem Steel Co., Sparrows Point, Md.  
Subject—*Management's View of Foundry Educational Activities*.  
Discussion Leader—M. J. Gregory, Peoria.

#### GRAY IRON SHOP COURSE SESSION

Monday, May 3, 8:00 P.M.

Presiding—E. J. Burke, Hanna Furnace Co., Buffalo, N.Y.  
Co-Chairman—A. N. Kraft, Wilkening Mfg. Co., Philadelphia.  
Subject—*Front Slagging of the Cupola*.  
Discussion Leader—T. W. Curry, Lynchburg Foundry Co., Radford, Va.

#### SAND SHOP COURSE SESSION

Monday, May 3, 8:00 P.M.

Presiding—R. F. Thomson, International Nickel Co., Detroit.  
Co-Chairman—C. A. Sanders, American Colloid Co., Chicago.  
Subject—*Non-Ferrous Foundry Sands*.  
Discussion Leader—J. J. Shannon, Jenkins Bros., Bridgeport, Conn.

#### ALUMINUM AND MAGNESIUM SESSION

Tuesday, May 4, 10:00 A.M.

Presiding—H. Brown, Solar Aircraft Co., Des Moines, Iowa.  
Co-Chairman—H. R. Youngkrantz, Apex Smelting Co., Chicago.  
Aluminum-Zinc-Magnesium-Copper Casting Alloys—L. W. Eastwood, Battelle Memorial Institute, Columbus, and L. W. Kempf, deceased, formerly with Aluminum Co. of America, Cleveland.

*Effect of Titanium on Grain Size and Tensile Properties of An Aluminum 4.5% Cu (No. 195) Casting Alloy*—W. E. Sicha and R. C. Boehm, Aluminum Co. of America, Cleveland.

*A Fluidity Test for Aluminum Casting Alloys*—W. E. Sicha and R. C. Boehm, Aluminum Co. of America, Cleveland.

#### BRASS AND BRONZE SESSION

Tuesday, May 4, 10:00 A.M.

Presiding—R. M. Brick, University of Pennsylvania, Philadelphia.  
Co-Chairman—C. A. Robeck, Gibson & Kirk Co., Baltimore, Md.

*The Technology of Copper-Lead Alloys*—R. W. K. Honeycombe, Melbourne, Australia, *Institute of Australian Foundrymen Exchange Paper*, presented by Wm. Ball, Jr., Magnus Brass Div., National Lead Co., Cincinnati.

*The Effect of Foundry Practice on the Properties of Some Binary Copper-Silicon Alloys*—A. I. Krynskiy, W. P. Saunders, and H. Stern, National Bureau of Standards, Washington, D.C.

#### MALLEABLE SESSION

Tuesday, May 4, 10:00 A.M.

Presiding—G. A. Vennerholm, Ford Motor Co., Dearborn, Mich.

Co-Chairman—W. B. McFerrin, Electro Metallurgical Co., Detroit.

*Effect of Manganese Sulphur Ratio on the Rate of Anneal of Blackheart Malleable Iron*—J. E. Rehder, Bureau of Mines, Ottawa, Canada.

*Pearlitic Malleable Irons, Plain and Alloyed*—D. J. Reese, International Nickel Co., New York, and Richard Schneidewind, University of Michigan, Ann Arbor, Mich.

#### ALUMINUM AND MAGNESIUM ROUND TABLE LUNCHEON

Tuesday, May 4, 12:00 Noon

Presiding—W. J. Klayer, Aluminum Industries, Inc., Cincinnati.

Co-Chairman—M. E. Gantz, American Magnesium Co., Cleveland.

*The Development of a Permanent Mold for Aluminum Tensile Test Bars*—L. J. Ebert, R. E. Spear, and G. Sachs, Case Institute of Technology, Cleveland.

Discussion on *Permanent Mold Casting of Aluminum and Magnesium Alloys*.

#### MALLEABLE ROUND TABLE LUNCHEON

Tuesday, May 4, 12:00 Noon

Presiding—J. H. Lansing, Malleable Founders' Society, Cleveland.

Co-Chairman—A. M. Fulton, Northern Malleable Iron Co., St. Paul, Minn.

Subject—*Malleable Foundry Mechanization*.

Discussion Leader—R. J. Anderson, Belle City Malleable Iron Co., Racine, Wis.

#### BRASS AND BRONZE SESSION

Tuesday May 4, 2:00 P.M.

Presiding—R. W. Parsons, Ohio Brass Co., Mansfield, Ohio.

Co-Chairman—A. K. Higgins, Allis-Chalmers Mfg. Co., Milwaukee.

*A New Permeable Metal Casting Plaster*—K. A. Miericke and E. S. Johnson, U. S. Gypsum Co., Chicago.

*The Value of Pressure Tests and Radiographs of Gun Metal Castings*—W. H. Baer, Naval Research Laboratory, Washington, D.C.

#### HEAT TRANSFER SESSION

Tuesday May 4, 2:00 P.M.

Presiding—E. C. Troy, National Engineering Co., Philadelphia.

Co-Chairman—H. F. Taylor, Massachusetts Institute of Technology, Cambridge, Mass.

*Heat Transfer Committee Report*—H. A. Schwartz, Chairman, National Malleable and Steel Castings Co., Cleveland.

*Thermal Conductivity of a Sand Mixture*—C. F. Lucks, O. L. Linebrink, and K. L. Johnson, Battelle Memorial Institute, Columbus.

*Studies of Solidification of Aluminum Castings*.

*Studies of Solidification of White Iron Castings*.

*A Study of Solidification of Steel Spheres*.

V. Paschkis, Columbia University, New York.

#### PATTERN SESSION

Tuesday May 4, 2:00 P.M.

Presiding—A. F. Pfeiffer, Allis-Chalmers Mfg. Co., Milwaukee.

Co-Chairman—L. F. Tucker, City Pattern & Foundry Co., South Bend, Ind.

*Core Box Design and Rigging for Core Blowing*—H. J. Jacobson, Industrial Pattern Works, Chicago.

*Metallizing Wood Patterns*—H. A. Erbe, General Steel Casting Co., Granite City, Ill.

#### LECTURE COURSE

Tuesday, May 4, 4:00 P.M.

Presiding—C. A. Robeck, Gibson & Kirk Co., Baltimore.

Co-Chairman—W. W. Edens, Badger Brass & Aluminum Co., Milwaukee.

Subject—*Test Procedures for Quality Control of Sand Castings, Brass, Bronze & Nickel Alloys*.

Discussion Leader—Wm. Romanoff, H. Kramer & Co., Chicago.

#### LECTURE COURSE

Tuesday, May 4, 4:00 P.M.

Presiding—H. N. Whitmore, National Malleable & Steel Castings Co., Cleveland.

Co-Chairman—Milton Tilley, National Malleable & Steel Castings Co., Cleveland.

Subject—*Test Procedures for Quality Control of Malleable Iron Castings*.

Discussion Leader—M. O. Booth, Central Foundry Div., General Motors Corp., Saginaw, Mich.

#### CHAPTER OFFICERS AND DIRECTORS DINNER

Tuesday, May 4, 7:00 P.M.

Presiding—W. B. Wallis, Vice-President, American Foundrymen's Association.

#### PLANT AND PLANT EQUIPMENT SESSION

Tuesday, May 4, 8:00 P.M.

Presiding—James Thomson, Continental Foundry & Machine Co., East Chicago, Ind.

Co-Chairman—E. W. Beach, Campbell, Wyant & Cannon Foundry Co., Muskegon, Mich.

*Modernization of the Small Foundry*—L. B. Knight, L. B. Knight & Assoc., Inc., Chicago.

## GRAY IRON SHOP COURSE SESSION

Tuesday, May 4, 8:00 P.M.

*Presiding*—H. H. Wilder, Eaton Mfg. Co., Vassar, Mich.*Co-Chairman*—B. A. Miller, Baldwin Locomotive Works, Eddystone, Pa.*Subject*—*Effect of Blast Variation on Cupola Control.**Discussion Leader*—B. P. Mulcahy, Consultant, Indianapolis, Ind.

## SAND SHOP COURSE SESSION

Tuesday, May 4, 8:00 P.M.

*Presiding*—D. F. Sawtelle, Malleable Iron Fittings Co., Branford, Conn.*Co-Chairman*—F. S. Brewster, Harry W. Dietert Co., Detroit.*Subject*—*Malleable Foundry Sands.**Discussion Leader*—R. P. Schauss, Illinois Clay Products Co., Chicago.

## ANNUAL BUSINESS MEETING AND CHARLES EDGAR HOYT ANNUAL LECTURE

Wednesday, May 5, 9:30 A.M.

*Presiding*—Max Kuniansky, President, American Foundrymen's Society.

President Kuniansky called the meeting to order as the Annual Business Meeting of the American Foundrymen's Society. He then presented the President's Annual Address. See page xiii.

Following this presentation President Kuniansky called on Secretary-Treasurer W. W. Maloney who reported on the nominations of Officers and Directors for the coming year and stated that no additional nominees had been received in accordance with the procedure prescribed in the Society By-Laws. He therefore cast the unanimous ballot of the membership of A.F.S. for the election of the following:

*President (to serve one year):*

W. B. Wallis, Pittsburgh Lecomelt Furnace Corp., Pittsburgh.

*Vice President (to serve one year):*

E. W. Horlebein, Gibson &amp; Kirk Co., Baltimore.

*Directors (to serve three years):*

T. H. Benners, Jr., T. H. Benners &amp; Co., Birmingham.

N. J. Dunbeck, Eastern Clay Products, Inc., Jackson, Ohio.

A. M. Fulton, Northern Malleable Iron Co., St. Paul.

R. Gregg, Reliance Regulator Co., Alhambra, Calif.

V. E. Zang, Unitcast Corp., Toledo, Ohio.

*Director (to serve one year):*

Max Kuniansky, Lynchburg Foundry Co., Lynchburg, Va.

Following announcement of elections, President Kuniansky introduced the newly elected Officers and Directors present. He then called on Secretary-Treasurer W. W. Maloney who announced the 1948 Apprentice Contest winners as follows:

*Patternmaking Division*

1st—Dwight L. French, Caterpillar Tractor Co., Peoria, Ill.

2nd—James J. Kennedy, Jr., Burns &amp; Wohlgenuth, Philadelphia.

3rd—Robert J. Schneider, Midwest Pattern Co., Minneapolis.

*Gray Iron Molding Division*

1st—John Dean Morehead, Caterpillar Tractor Co., Peoria, Ill.

2nd—John Coreno, Hill Acme Foundry, Cleveland.

3rd—John Valichnac, Fulton Foundry &amp; Machine Co., Cleveland.

*Steel Molding Division*

1st—Adolph Scanzoni, Continental Foundry &amp; Machine Co., East Chicago, Ind.

2nd—Joseph J. Wienclawski, Birdsboro Steel Foundry &amp; Machine Co., Birdsboro, Pa.

3rd—Frank Alabaise, West Steel Castings Co., Cleveland.

*Non-Ferrous Molding Division*

1st—Thomas D. Craven, Howard Foundry Co., Chicago

2nd—Frank Posanki, Allis-Chalmers Co., Milwaukee.

3rd—John Wegner, Waukesha Foundry, Waukesha, Wisc.

The Society arranged to have the four first-prize winners present at the convention and to receive their awards in person. The first-prize winners were called to the platform and President Kuniansky presented the awards to each of them individually. The first-prize award consisted of \$100 and an engraved Certificate of Award.

Following presentation of Apprentice Contest awards, President Kuniansky introduced Mr. E. Longden, author of the IBF exchange paper to A.F.S. He brought with him a message of congratulations from the I.B.F. and expressed his pleasure on his attendance at the 52nd Annual A.F.S. Convention.

President Kuniansky then called on Fred G. Seifing, Chairman of the Educational Division of A.F.S. who spoke briefly on the plans of the Educational Division of A.F.S. for the coming year.

President Kuniansky then called on Mr. H. Bornstein, Chairman of the Annual Lecture Committee who introduced the Charles Edgar Hoyt Lecturer, Charles E. Nelson who delivered his lecture on "Grain Size Control in Magnesium Castings" which begins on p. 1.

## ENGINEERING SCHOOL GRADUATES LUNCHEON

Wednesday, May 5, 12:00 Noon

*Presiding*—H. H. Judson, Minneapolis Moline Power Implement Co., Minneapolis, Minn.*Co-Chairman*—C. V. Nass, Pettibone-Mulliken Corp., Chicago.  
*Engineering Education*—R. M. Brick, University of Pennsylvania, Philadelphia.

## PATTERN ROUND TABLE LUNCHEON

Wednesday, May 5, 12:00 Noon

*Presiding*—L. F. Tucker, City Pattern & Foundry Co., Inc., South Bend, Ind.*Co-Chairman*—H. K. Swanson, Swanson Pattern & Model Works, East Chicago, Ind.*Subject*—*Patternmaking.**Discussion Leader*—Martin Rintz, Continental Foundry & Machine Co., East Chicago, Ind.

## GRAY IRON SESSION

Wednesday, May 5, 2:00 P.M.

*Presiding*—V. A. Crosby, Climax Molybdenum Co., Detroit.*Co-Chairman*—D. E. Krause, Gray Iron Research Institute, Columbus.*Relation of Cupola Research to Progress in Cast Iron Development*—R. G. McElwee, Vanadium Corporation of America, Detroit.*A Suggested Method for the Determination of Coke Reactivity to Carbon Dioxide at Combustion Temperatures*—H. E. Flanders, University of Utah, Salt Lake City.

## REFRACTORIES SESSION

Wednesday, May 5, 2:00 P.M.

*Presiding*—R. H. Stone, Vesuvius Crucible Co., Swissvale, Pittsburgh, Pa.*Co-Chairman*—A. S. Klopff, Western Foundry Co., Chicago.*Testing Refractories for the Foundry*—S. M. Swain, North American Refractories Co., Cleveland.*Review of Refractories Used in Steel and Iron Foundries*—W. H. Owen, Harbison-Walker Refractories Co., Bellevue, Pa.

## FOUNDRY EDUCATIONAL FORUM

Wednesday, May 5, 2:00 P.M.

## LECTURE COURSE SESSION

Wednesday, May 5, 4:00 P.M.

*Presiding*—G. A. Vennerholm, Ford Motor Co., Dearborn, Mich.*Co-Chairman*—F. J. Dost, Sterling Foundry Co., Wellington, Ohio.*Subject*—*Test Procedures for Quality Control of Gray Iron Castings.**Discussion Leader*—F. J. Walls, International Nickel Co., Detroit.

## SAND SESSION

Wednesday, May 5, 4:00 P.M.

*Presiding*—E. C. Troy, National Engineering Co., Philadelphia.*Co-Chairman*—C. W. Briggs, Steel Founders Society, Cleveland.*Surface Gas Pressure of Molding Sands and Core Sands*—H. W. Dietert, H. H. Fairfield, and F. S. Brewster, Harry W. Dietert Co., Detroit.*Changes in Chemistry of Liquid Steel in Contact with Sand—Report of Mold Surface Committee, J. B. Caine, Chairman, Sawbrook Steel Castings Co., Lockland, Cincinnati, Ohio.**A Survey of Sieve Series and Grade Scales*—R. E. Morey, Naval Research Laboratory, Washington, D.C.

## CANADIAN DINNER

Wednesday, May 5, 7:00 P.M.

Presiding—E. N. Delahunt, Warden King Ltd., Montreal.

## REFRACTORIES SESSION

Wednesday, May 5, 8:00 P.M.

Presiding—A. S. Klopff, Western Foundry Co., Chicago.

Co-Chairman—C. F. Joseph, Central Foundry Div., General Motors Corp., Saginaw, Mich.

Information Please—Question and Answer Panel.

## GRAY IRON SHOP COURSE SESSION

Wednesday, May 5, 8:00 P.M.

Presiding—W. W. Levi, Lynchburg Foundry Co., Radford, Va.

Co-Chairman—C. L. Lane, Florence Pipe &amp; Machine Co., Florence, N.J.

Subject—Cupola Coke.

Discussion Leader—J. A. Bowers, American Cast Iron Pipe Co., Birmingham, Ala.

## SAND SHOP COURSE SESSION

Wednesday, May 5, 8:00 P.M.

Presiding—E. L. Thomas, Cadillac Motor Car Company, Detroit.

Co-Chairman—N. J. Dunbeck, Eastern Clay Products Co., Inc., Jackson, Ohio.

Subject—Gray Iron Foundry Sands.

Discussion Leader—T. W. Curry, Lynchburg Foundry Co., Lynchburg, Va.

## GRAY IRON SESSION

Thursday, May 6, 10:00 A.M.

Presiding—A. E. Schuh, U. S. Pipe &amp; Foundry Corp., Burlington, N.J.

Co-Chairman—F. G. Sefing, International Nickel Co., New York.

A Laboratory Evaluation of Some Automotive Cast Irons—A. B. Shuck, Koppers Co., Inc., American Hammered Piston Ring Div., Baltimore, Md.

Contraction and Distortion in Ferrous Castings—E. Longden, P. R. Jackson &amp; Co. Ltd., Manchester, England—Institute of British Foundrymen Exchange Paper.

## JOB EVALUATION AND TIME STUDY SESSION

Thursday, May 6, 10:00 A.M.

Presiding—R. J. Fisher, The Falk Corp., Milwaukee.

Co-Chairman—M. E. Annich, American Brake Shoe Co., Mahwah, N.J.

Grinding Standards Help to Eliminate Cleaning Room Bottlenecks—Dean Van Order, Burnside Steel Foundry Co., Chicago.

## STEEL SESSION

Thursday, May 6, 10:00 A.M.

Presiding—W. W. Moore, Burnside Steel Foundry Co., Chicago.

Co-Chairman—H. F. Taylor, Massachusetts Institute of Technology, Cambridge, Mass.

Observation on Knock-Off Risers as Applied to Steel Castings—J. A. Duma, Portsmouth Foundry &amp; Engr. Corp., and S. W. Brinson, Norfolk Naval Shipyard, Portsmouth, Va.

A Theoretical Approach to the Problem of Dimensioning Risers—J. B. Caine, Savbrook Steel Castings Co., Lockland, Cincinnati, Ohio.

## STEEL ROUND TABLE LUNCHEON

Thursday, May 6, 12:00 Noon

Presiding—J. A. Duma, Portsmouth Foundry &amp; Engr. Corp., Portsmouth, Va.

Co-Chairman—J. A. Rassenfoss, American Steel Foundries, East Chicago, Ind.

Motion Picture on Gating Systems for Metal Casting—Wm. H. Johnson and Wm. O. Baker, Naval Research Lab., Washington, D.C.

Followed by Discussion.

## FOUNDRY COST SESSION

Thursday, May 6, 2:00 P.M.

Presiding—R. L. Lee, Grede Foundries, Inc., Milwaukee.

Co-Chairman—C. E. Westover, Westover Engineers, Milwaukee.

Question and Answer Panel—Cost Problems of General Interest, Method of Cost Accounting, Distribution of Costs, etc.

## GRAY IRON SESSION (SYMPOSIUM ON HEAT TREATMENT OF GRAY CAST IRON)

Thursday, May 6, 2:00 P.M.

Presiding—J. S. Vanick, International Nickel Co., New York.

Co-Chairman—T. E. Eagan, Cooper Bessemer Corp., Grove City, Pa.

Fundamentals of Heat Treating Gray Cast Iron—A. Boyles, U. S. Pipe &amp; Foundry Co., Burlington, N.J.

Gray Iron Hardenability and Its Relation to Air Quenching of Castings—R. A. Flinn and R. J. Ely, American Brake Shoe Co., Mahwah, N.J.

## LECTURE COURSE SESSION

Thursday, May 6, 4:00 P.M.

Presiding—G. A. Lillieqvist, American Steel Foundries, East Chicago, Ind.

Co-Chairman—C. H. Lorig, Battelle Memorial Institute, Columbus.

Subject—Test Procedures for Quality Control of Steel Castings.

Discussion Leader—J. W. Juppenlatz, Lebanon Steel Foundry, Lebanon, Pa.

## GRAY IRON SESSION (SYMPOSIUM ON HEAT TREATMENT OF GRAY CAST IRON)

Thursday, May 6, 4:00 P.M.

Presiding—H. Bornstein, Deere &amp; Co., Moline, Ill.

Co-Chairman—D. A. Paull, Sealed Power Corp., Muskegon, Mich.

Stress Relief of Gray Iron Castings—J. H. Schaum, Naval Research Laboratory, Washington, D.C.

Conventional vs. Salt Bath Hardening of Cast Iron Cylinder Liners—G. Lahr, Detroit Diesel Engine Div., General Motors Corp., Detroit.

## SAND SESSION

Thursday, May 6, 4:00 P.M.

Presiding—J. A. Rassenfoss, American Steel Foundries, East Chicago, Ind.

Co-Chairman—E. C. Zirzow, National Malleable &amp; Steel Castings Co., Cleveland.

Progress Report on High Temperature Properties of Steel Molding Sands—P. E. Kyle, Cornell University, Ithaca, New York.

Strainer Cores—H. L. Campbell, Western Foundry Co., Chicago.

Causes of Rat-Tail Casting Defects—Report of A.F.A. Committee on Physical Properties of Iron Foundry Molding Materials at Elevated Temperatures, H. W. Dietert, Chairman, Harry W. Dietert Co., Detroit.

## JOB EVALUATION AND TIME STUDY SESSION

Thursday, May 6, 4:00 P.M.

Presiding—R. J. Fisher, The Falk Corp., Milwaukee.

Co-Chairman—M. E. Annich, American Brake Shoe Co., Mahwah, N.J.

Question and Answer Panel.

## A.F.A. ALUMNI DINNER

Thursday, May 6, 7:00 P.M.

Presiding—Past President S. V. Wood, Minneapolis Steel Castings Co., Minneapolis.

## GRAY IRON SHOP COURSE SESSION

Thursday, May 6, 8:00 P.M.

Presiding—K. H. Priestley, Vassar Electroloy Products Inc., Vassar, Mich.

Co-Chairman—Ralph Koch, U. S. Pipe &amp; Foundry Co., Burlington, N.J.

Subject—Factors Affecting Fluidity of Cupola Iron.

Discussion Leader—R. F. Flora, Clover Foundry, Muskegon, Mich.



## SAND SHOP COURSE SESSION

Thursday, May 6, 8:00 P.M.

Presiding—R. H. Jacoby, Key Co., East St. Louis, Ill.

Co-Chairman—E. E. Woodliff, Foundry Sand Service Engineering Co., Detroit.

Subject—Steel Foundry Sands.

Discussion Leader—Charles Locke, Armour Research Foundation, Chicago.

## GRAY IRON SESSION

Friday, May 7, 10:00 A.M.

Presiding—J. T. MacKenzie, American Cast Iron Pipe Co., Birmingham, Ala.

Co-Chairman—W. W. Levi, Lynchburg Foundry Co., Radford, Va.

Improvement of Machinability in High Phosphorus Gray Cast Irons—W. W. Austin, Jr., Southern Research Institute, Birmingham, Ala.

The Production of Nodular Graphite Structure in Gray Cast Iron—H. Morrogh, British Cast Iron Research Assn., Alvechurch, Birmingham, England.

## STEEL SESSION

Friday, May 7, 10:00 A.M.

Presiding—Frank Kiper, Ohio Steel Foundry, Springfield, Ohio.

Co-Chairman—V. E. Zang, Unitcast Corporation, Toledo, Ohio.

Tensile Properties vs. Composition of Double Normalized Cast Steel—W. K. Bock and H. A. Schwartz, National Malleable &amp; Steel Castings Co., Cleveland.

The Delayed Quench for Steel Castings—S. L. Gertsman, Bureau of Mines, Ottawa, Canada.

Electro-Chemical Cleaning of A Large Steel Casting—An Experiment—J. A. Wettergreen, General Electric Co., Schenectady.

## GRAY IRON SESSION

Friday, May 7, 2:00 P.M.

Presiding—R. G. McElwee, Vanadium Corp. of America, Detroit.

Co-Chairman—E. C. Jeter, Ford Motor Co., Dearborn, Mich.

Oxygen-Enriched Cupola Blast—W. C. Wick, Armour Research Foundation, Chicago.

Solidification Characteristics of Gray Iron—J. H. Schaum, Naval Research Lab., Washington, D.C., and J. E. Fifield, International Nickel Co., Hartford, Conn.

## STEEL SESSION

Friday, May 7, 2:00 P.M.

Presiding—J. F. Randall, Ford Motor Co., Detroit.

Co-Chairman—D. C. Zuege, Sivy Steel Casting Co., Milwaukee.

Statistical Quality Control—A New Tool for the Foundryman—H. H. Johnson and G. A. Fisher, National Malleable &amp; Steel Castings Co., Sharon, Pa.

Development of Techniques for Quality Welding of Steel Castings—J. D. Wozny and E. L. LaGrelus, American Steel Foundries, East Chicago, Ind.

## ANNUAL BANQUET

Friday, May 7, 7:00 P.M.

Presiding—Max Kuniansky, President, American Foundrymen's Society.

The Annual A.F.S. Dinner of the Society's 52nd Annual Convention was called to order by President Max Kuniansky, presiding. Mr. Kuniansky read a telegram from A. M. Cadman, President of Pittsburgh Foundrymen's Association addressed to W. B. Wallis, A.F.S. President-Elect as follows:

"The officers and members of the Pittsburgh Foundrymen's Association congratulate you upon your election as President of the American Foundrymen's Society. We are proud in the fact that you are the sixth Pittsburgher to receive that single honor. It is a tremendous job but we all know that it will be well taken care of." Following singing of the National Anthem by Miss Geraldine Willier, President Kuniansky introduced Past President S. V. Wood, Chairman of the A.F.S. Board of Awards. After brief introductory remarks Past President Wood introduced Past President J. L. Wick, Jr., who presented the Wm. H. McFadden Gold Medal to Egbert H. Ballard.

## PRESENTATION OF MCFADDEN GOLD MEDAL

Mr. Wick stated:

"Tonight we partake in a ceremonial sponsored in 1921 by

four devoted members of the American Foundrymen's Society. They believed that 'accomplishment' is its own reward, yet realized that public recognition of 'accomplishment' completes this satisfaction. So they established a means by which conspicuous performance might be recognized in a tangible way.

"They did this by entrusting to the American Foundrymen's Society an endowment fund, the income from which made gold medals of merit available to the Board of Awards. The membership of the Board of Awards is composed of the last seven living presidents of the Society.

"By the wise provision of these sponsors, these awards are made only at such times as in the judgment of the Board of Awards some individual merits such recognition.

"Tonight's recipient of this Wm. H. McFadden Medal is, in my opinion, a man who has much in common with Wm. H. McFadden. Their 'backgrounds, approach to life, and points of view,' harmonize completely—Egbert H. Ballard. One's first impression of a man is not always borne out by fuller acquaintance with him. With you, however, my first impression was likewise my second; and since 1927 when I first met you, I have always found you just the same—serene and gracious, cheerfully confident, assuring all who know you that you could and would take care of your assignments. The spirit of friendship has always seemed to flow right from you.

"According to the record, you became the 31st president of the American Foundrymen's Society, after having served as vice president during 1931, and after having previously served as a member of the board of directors from 1927-30.

"The record states that you were born in Hartford, Conn., and received your education in Lynn, Mass., to which city your family had moved along with the former Thomson-Houston Electric Co. which eventually became the nucleus of the Lynn plant of the General Electric Co.

"I discovered that you began doing part-time work during summer vacations when you were only 13 years of age.

"Except for the few years when you were superintendent of the Massachusetts Steel Casting Co. at Everett, Mass., you have been associated with the General Electric Co. throughout your entire life until your retirement in 1943.

"Industry does not prosper under the leadership of those whose fitness is such that five years or so comprises their cycle of complete usefulness. High responsibility is long responsibility; and we never find continuity in an enterprise without finding a quality of devotion in its administration.

"Industry is more than mechanization and material structure; it is made of human stuff, and those who grow with it over the years must be a part of it.

"You have always been definitely a part of the foundry industry. For the same reason, the American Foundrymen's Society is great today because it has been able to command the continuing loyalty and the continuous devotion of men like you.

"Egbert H. Ballard, it was your lot to be elected president in 1931-32, a year beset by troubles and uncertainty.

"We were just finishing the boom years from 1921-31. We called it the 'new era' and in it repeated all the errors that have characterized similar periods in the past. There was over-expansion, over-capitalization, extravagances, attempts at monopolistic control, unsound credits, labor unions pushing wages up to the levels that killed off business.

"That year of strife and turmoil, 1931, produced substantial losses in our membership in common with practically all associations and societies.

"You were the president of the American Foundrymen's Society and apparently found joy in testing your prowess against the forces of adversity which fell into your lap.

"Life had proved a testing ground to you, and you immediately showed your mettle in the way you stood up against misfortune.

"You rose to your full stature and demonstrated your fitness to solve problems by cancelling the convention and exhibition which had been planned for Philadelphia. This action on your part received the unanimous approval of all our loyal exhibitors and members.

"You, and your Board, and the A.F.S. staff, immediately brought the operating budget of our Society in harmony with conditions as they then existed.

"You saw in a little clearer light and had a keener appreciation of the value to our members and to our engineers in particular, of the precious data, methods, practices, and proceed-

is known to and available to certain groups and sections in our industry, and you undertook its organization, correlation and assembly, and through your devotion and follow-up, established committees who produced and made available, four years later, the first edition of the CAST METALS HANDBOOK which is now in its third edition.

Egbert H. Ballard, in the name of the Board of Awards and the Board of Directors, and membership of the American Foundrymen's Society, I have the privilege and honor to present to you the Wm. H. McFadden Gold Medal.

*"Awarded in recognition of your steadfast and sincere efforts in behalf of the Society while serving as its president during most difficult times, and for your continual contribution to the foundry industry throughout your entire business career."*

"May this Gold Medal always remain a permanent testimony to your accomplishments and be a perpetual reminder that your associates were not unmindful of your continual contributions to our great industry."

#### RESPONSE BY MR. BALLARD

Mr. Ballard graciously accepted the Wm. H. McFadden Gold Medal saying:

"With humility, deep appreciation, and many thanks I accept this medal."

#### PRESENTATION OF PENTON GOLD MEDAL

Mr. Wood then introduced Fred J. Walls who presented the John A. Penton Gold Medal to Richard G. McElwee.

Mr. Walls stated:

"On but few occasions during a man's lifetime is he afforded the opportunity and pleasure of presenting a gold medal to a long-time friend and fellow-worker, and I am deeply appreciative of this privilege.

"Tonight we join those who have preceded us in honoring those of the present, and in so doing, dedicate the future progress of the American Foundrymen's Society, and the great industry for which it was founded, to men of vision, loyalty and devotion.

"John A. Penton, in whose honor this medal is awarded, was one of these men; and so is the 1948 recipient, Richard G. McElwee. Mac, as he is affectionately known by his many friends throughout the foundry industry, began his career 34 years ago at the Muncie Foundry & Machine Co. as Chief Inspector. He is a native of Williamsport, Pa., where he was born in 1890, and he received his education there and in Galeton, Pa. From Muncie he became casting buyer for General Motors Truck Co. in Pontiac, Mich. and then became associated with American Car & Foundry Co., Detroit, where he worked on Engineering Specifications.

"After a short time as foundry foreman at Whitehead and Kales Co. of Ecorse, Mich., he joined the Ecorse Foundry Co., and resigned from the position of General Manager there in 1936 to accept his present position as Manager of the Foundry Alloys Division of the Vanadium Corp. of America, Detroit.

"While associated with the Ecorse Foundry, during the period of 1925-36, he pioneered electric furnace melting of high strength irons, special alloys, and short time anneal malleables.

"When Mr. Penton presented the medal endowment to A.F.S., he made the suggestion that the medal should be awarded,

*"For the most noteworthy contribution along the line of electric furnace practice in melting ferrous and non-ferrous metals."*

Therefore, it is significant that Richard G. McElwee fulfills the desire of John A. Penton in this respect, and, in addition, the Board of Awards emphasizes his vision, loyalty, and devotion in the citation which reads,

*"Outstanding contributions to the dissemination of information to the Foundry Industry, especially for his splendid effort on behalf of the Cupola Research Project."*

I need not list these contributions, which were in the form of papers before many related technical societies and talks before most of our Chapters and conferences on foundry procedures and engineering properties of cast iron.

"Mr. McElwee's interest in the encouragement and development of young foundrymen has always been sincere and fruitful to the industry. He has been a loyal committeeman, and, at present, is Chairman of the Gray Iron Division of A.F.S.

"Richard G. McElwee, it is my great pleasure, on behalf of the Board of Awards of the American Foundrymen's Society to present to you the 1948 John A. Penton Gold Medal, and with it, this certificate of Honorary Life Membership in this great technical society."

#### RESPONSE BY MR. McELWEE

Mr. McElwee graciously accepted the John A. Penton Gold Medal saying:

"Needless to say, I feel highly honored and very grateful to have been the recipient of this Award.

"In view of its being awarded for 'dissemination of information' I would like to comment that I feel that I have been doubly paid. On each occasion on which information has been given, whether at Chapter meetings, conference or private interview, I have found that one takes at least as much as he gives. Having already received my reward you can readily see why I feel doubly grateful for this honor."

#### PRESENTATION OF SIMPSON MEMORIAL MEDAL

Mr. Wood then introduced W. L. Seelbach who presented the Peter L. Simpson Memorial Gold Medal Award to Peter E. Rentschler.

Mr. Seelbach stated:

"I can think of no greater pleasure and honor than being able to present the Peter L. Simpson Gold Medal Award of the American Foundrymen's Society, to a man who so richly deserves it.

"Many men, when making an award presentation, wait until the last sentence of their presentation before naming the man who is to receive the award. In my case I want to give you the name without further ado and then talk about him later. Peter Earl Rentschler, President of The Hamilton Foundry & Machine Co. of Hamilton, Ohio, is the man to whom I am going to present this award.

"Pete, the name that everybody uses after they have once met him, was born in 1897, and graduated from Princeton University in 1920. In the summer of 1920 he started in the foundry and has worked there continuously. He has seen active employment in all departments of the foundry, and became president of his company in 1927.

"In addition to his duties of operating a large modern production gray iron foundry, he has carried on many other society, industrial, and public and civic activities. Peter Rentschler's work in the Gray Iron Founders' Society is known to all the members of the Society, and to others in the foundry industry. The outstanding work he did for the Society was on Cost Activities, and the preparation and adoption of Gray Iron Trade Customs. On Oct. 22, 1946, he was given an award from the Gray Iron Founders' Society, Inc. for his outstanding work in these activities.

"His interest in the educational program of the City of Hamilton, Ohio has been continuous since 1939, that is the date he was first elected a member of the Hamilton City Board of Education.

"On January 1, 1944, he was elected President of the Hamilton City Board of Education, and held this position for four years.

"His civic duties consisted of active participation and whole-hearted cooperation, and in many cases, Chairman of Budget Drives for the Hospitals; Community Chest; Safety Council, and Welfare Associations of the City of Hamilton.

"How Peter Earl Rentschler has time for all this work is something many of us have not been able to understand, but like so many men of the type of Pete Rentschler, a leader in one field becomes a leader in many others,—there seems to be no limit in the capacity for work of these kind of men.

"Peter Rentschler has always had the idea that the foundry industry had never been recognized as a very important cog in the industrial machinery of the United States, and when Pete has an idea he carries it out as evidenced by his bulletins to his employees; his open houses; his articles in national magazines and newspapers, and the accomplishment of being awarded from the Navy Department, Sept. 23, 1943, the first Army and Navy 'E' to go to the industry.

"He operates one of the outstanding fine, clean and well-managed, gray iron foundries. His public relations work, his promoting of better housekeeping and safety practices in his and other foundries, is known by everybody in the industry. Safety and hygiene in the foundry was one of his accomplish-

ments,—he preached this wherever he went,—spent many long days before legislative committees in Columbus, Ohio so as to have workable rules and regulations set forth covering hygiene and safety.

"His work along this line is always with the vision of a perfect employee-employer relationship. It is no wonder then that the Board of Awards consisting of the last seven living past presidents of the American Foundrymen's Society, chose Peter Earl Rentschler as the recipient of this year's Peter L. Simpson Gold Medal Award.

"I am proud to be able to present this award to you, and wish you the best of everything, which you so richly deserve."

#### RESPONSE BY MR. RENTSCHLER

Mr. Rentschler graciously accepted the Peter L. Simpson Gold Medal Award saying:

"Thank you Mr. Seelbach. It is an honor to have been selected the recipient of the Peter L. Simpson Memorial Medal for 1948. I wish to thank the Board of Awards and the Board of Directors of the American Foundrymen's Society for this recognition. The work that merited this award is not my work alone but it is the cooperative efforts and energies of the loyal group

of associates, which resulted in the accomplishments being recognized.

"I am sorry that these people are not here tonight to see this award presented to me, but which I want to think of as being presented to them through me. Thank you again."

#### PRESENTATION OF HONORARY LIFE MEMBERSHIP

Mr. Wood then presented a Certificate of Honorary Life Membership to the retiring President, Max Kuniansky who graciously accepted it with an expression of appreciation.

#### GUEST SPEAKER'S ADDRESS

Following presentation of Awards, the guest speaker of the evening, Dr. Karl T. Compton, President, Massachusetts Institute of Technology, Cambridge, Mass. presented his talk entitled "Team Play of Hand and Brain" in which he reviewed the discovery of the electron, of radioactivity, and of x-rays. In discussing the development and production of artificially radioactive materials he cited metallurgical and medical uses of radioactive isotopes. In conclusion, Dr. Compton pointed out that research and development pay off, and that the practical techniques of production and instrumentation must team with theoretical techniques for their mutual success.



# PRESIDENT'S ANNUAL ADDRESS

**Max Kuniansky**

Vice President and General Manager  
Lynchburg Foundry Co.  
Lynchburg, Va.

WE ARE PLEASED that our 52nd Annual Convention and Exhibit are being held in the city where the American Foundrymen's Society was founded in 1896. This exhibit is the largest in the history of AFS. Our Society is deeply grateful to the Foundry Equipment Manufacturers Association and the Foundry Supply Manufacturers Association for the splendid co-operative job they have done in making this a truly great show. We are particularly grateful to our Technical Director and the some 500 technical committee members of our various committees in the Technical Divisions who have put into being an excellent and well-developed technical program. We are very grateful to the British Foundrymen for their exchange paper (page 36) and also for the excellent paper by Mr. Morrogh (page 73). We wish to express thanks also to the Australian and French foundrymen for their exchange papers (pp. 57 and 66). It is our earnest hope that the many foreign visitors who have come to this show will feel that they are amply rewarded for their efforts.

Your membership stands today at its highest level in the history of the Society. As of March 31, 1948, for the first time in its history we passed 10,000 members. The exact figure as of that date was 10,174. We are proud that in this total we include 134 student members and 450 foreign members. It is gratifying to note that 90 per cent of our entire membership is affiliated with our 39 regular and five student chapters. This is truly a tribute to your Vice President, Mr. Wallis, and to each membership chairman of each chapter.

## Seven New Chapters Formed

We are pleased to report that since July 1, 1947, we have seven new chapters. These are, chronologically: Eastern New York at Albany, Tennessee at Chattanooga, and five student chapters as follows: University of Minnesota, Ohio State University, Missouri School of Mines and Metallurgy, Massachusetts Institute of Technology, and Oregon State. It is most gratifying to note our growth in student chapters because it is this type of endeavor on the part of AFS with its Educational and Technical Development Programs that bears great promise for a future influx of highly desirable technical personnel into the foundry industry.

During this past year your President and Vice President, aided at times by your Secretary and your Technical Director, visited every chapter in the Society with the exception of Mexico City and the Oregon State College Student Chapter. The growth of the Society is such that it is no longer possible for your President to visit all the chapters as he would very much like to. However, dividing the chores with the Vice President and the two staff officers mentioned has made this close contact with our chapters possible. We have had the privilege at each point to sit in with the Board of Directors and discuss the problems of the Chapter as related to the National organization, and we feel that in this manner we are forging a bond that will be highly beneficial to both.

## Society's Finances Sound

I am pleased to report that your Society's finances are in excellent shape at the present time. The four factors contributing to this condition are:

1. The constantly increasing membership, especially in company and sustaining memberships
2. Publication and sale of more and better technical books and pamphlets
3. The growing recognition of the AMERICAN FOUNDRYMAN as a magazine for the foundryman and, of course
4. The successful completion of the 1948 Exhibit.

Our policy to solicit advertising for the AMERICAN FOUNDRYMAN is bearing fruit and that endeavor to date has convinced us that we were right in entering this field. You may be sure that our efforts along this line will be pursued relentlessly.

Due to the lack of balance in income between our exhibit year and the non-exhibit year, your Finance Committee will put the Society on a two-year budget and it will also make possible the preparing of a long-range program of activities for the Society.

One of the significant developments has been the formation of the National Castings Council, of which your Society is a member. It appears to us that its objective of covering the overall industry in the future will be of great benefit to all the supporting members as well as to the foundry industry.

This year again in many of our Chapter areas we held successful management meetings and we feel

that the closer tying in of management with AFS interests is much to be desired. We also had a number of successful regional conferences.

### Staff Retirement Plan

This year your Board has approved a staff retirement plan for its National Headquarters personnel. We feel that this retirement plan will assure us of having a lower turnover in our National Headquarters and also of being able to attract a good calibre of personnel in the future.

We have scheduled our Chapter Chairmen Conference for the last three days in June and this year, for the first time, we are inviting both the Chairmen and Vice-Chairmen. These have proven very helpful to the National Headquarters Staff as well as to the Chapters. This year also your by-laws committee has done an excellent job in revising our by-laws to meet the changing needs of the activities of your Society.

To those of you who have not read in the last issue of the *AMERICAN FOUNDRYMAN* of the untimely passing of our director, R. Hudson McCarroll, I wish to make note at this time of the very serious loss AFS has had, especially because of Mr. McCarroll's great interest in the technical phases of AFS.

### Technical Activities

I wish to call to your attention briefly the highlights of the technical activities of the Society. These activities are headed by your Technical Director, S. C. Masari, and the Technical Correlations Committee. We have recently employed Philip D. Johnson to head up the Educational Activities as a part of our Technical Development Program and look forward to a concise and well-directed program for our educational activities now in process of formulation. We have eight Divisions and numerous General Interest Committees. There are approximately 450 members of these committees that comprise the best foundry talent available and are the source of all the technical information published by AFS. Three major activities of the Technical Committees are:

1. Research projects.
2. Research and technical papers, and preparation of the technical program for the Annual Convention.
3. Preparation of special publications.

Research projects are now actively under way in the following institutions:

Sand Research.....Cornell University  
Heat Transfer.....Columbia University  
Fundamentals of Flow in Aluminum and  
Magnesium Molds...Battelle Memorial Institute  
Heat Treatment of Pearlitic Malleable  
Iron.....University of Michigan  
Fracture Tests for Bronze...University of Michigan  
Hot Tears in Steel Castings  
.....Armour Research Foundation

Cupola Research employs a full time research investigator under the supervision of the Cupola Research Committee. Our publications and activities are in the following fields:

Technical articles in the *AMERICAN FOUNDRYMAN*.

Technical papers for presentation at the Convention.

Round Table Discussions.

Preprints of all Convention Papers.

Annual TRANSACTIONS.

Special publications total about 25, the most recent being the *CUPOLA HANDBOOK* and the book on the *ANALYSIS OF CASTINGS DEFECTS*. Five of the existing publications are now in process of revision by committees and when completed new editions will be published.

We also have a program of liaison with many of the principal technical societies in the country, including such organizations as:

The Society of Automotive Engineers.

American Society for Testing Materials.

American Society of Mechanical Engineers.

American Institute of Mining and Metallurgical Engineers.

American Society for Metals.

American Standards Association.

We also have a program of active co-operation with governmental agencies on matters relating to cast metals.

The current educational program is incorporating some of the following phases of educational work:

Foundry Visitations.

Foundry Talks.

AFS Apprentice Contests.

Chapter Educational Courses.

Co-operation with Boy Scouts.

Use of Recruiting Literature and Teaching Aids.

### Publication of Textbooks

We have authorized the publication of textbooks by the Educational Division at the college level; at the trade training school level, and at the high school level. We have already published course outlines for vocational schools, engineering colleges, training college graduates in industry, co-operative courses for engineering schools and co-operative courses for technical high schools. We shall publish within the next few months a Foreman Training Course Outline and an Outline of Apprentice Training Standards, as well as an Apprentice Training Course Outline.

### Annual Apprentice Contest

Once again we have staged our Annual Apprentice Contest which has been very successful. We wish at this time to express our thanks to all chapters and companies who have co-operated in making this Apprentice Contest successful.

I wish to express my thanks to the members of our Board of Directors and our Executive Committee for the very magnificent and excellent job they have done in running the affairs of your Society.

I wish to extend my thanks to the National Headquarters Staff, and to Mr. Maloney in particular, for the job they are doing in gearing their work to the constantly growing activities of the Society.

I wish to thank particularly the army of volunteer workers on our many technical committees who are, after all, the AFS.

## Report of the Secretary for the Fiscal Year 1947-48

THIS REPORT of the Secretary must, necessarily, be in large part a report on what the Society has accomplished through the efforts of its membership, rather than any picture of Staff accomplishments. It also reports accomplishments resulting from the studied actions taken and policies established by the Officers and Directors of the Society, because if it be a good report, it is good only because of the continuation and further development of the cooperative spirit without which A.F.S. as a Technical Society could not progress.

### Membership

During the fiscal year ended June 30, 1948, membership in the Society continued to show a substantial increase, in spite of predictions that have been made now for several years that each year we might expect a definite decrease in new members. Thus, the membership on June 30, 1948 totaled 10,403, against 9,683 on June 30 one year ago, or a net increase of 720.

Of the total membership on June 30, 1948, 91.5 per cent were affiliated with the 39 regular and 5 Student Chapters. This percentage compares with 90.8 per cent a year ago. At the same time, the number of members without Chapter affiliation increased from 772 to 884, including 449 members residing in countries outside North America on June 30, 1948, against 392 in 1947. A chart of membership growth shows no material change in the continued almost vertical climb of AFS membership since the first Chapter was established in 1934.

To the Membership Committee Chairmen of the Chapters can be attributed, as in past years, this continued accession of new members and growth of the Society. These men, without compensation in any way, are constantly adding to the influence of the Society wherever castings are made and, of course, to the strength of the Chapters as well.

The big problem to be undertaken during 1948-49 and thereafter, must be the retention of a maximum number of the new members enrolled each year. This problem is pointed up by the following figures for the past 11 years, 1937-38 to 1948-48 inclusive:

NEW MEMBERS VS. MEMBERS LOST VS. NET MEMBERSHIP GAIN  
1937-38 to 1947-48, inclusive

	Total	11-Year Average
New Members (including reinstatements) .....	14,103	1,282
Members Lost (all causes) .....	6,152	559
Net Membership Gain .....	7,951	723

From the above it is evident that out of every 100 members gained during any one fiscal year, 43 other members are lost. Thus every effort must be made to keep the membership sold on its value to the individual, and this we feel is directly associated with the development of Chapters in areas where Chapters do not now exist but could be supported, and the constant betterment of those services now being rendered.

To this end, we feel that continued distribution of paper-bound copies of the bound volume of TRANSACTIONS gratis to all members, on request, is of considerable importance. Of equal if not greater importance, is constant betterment of AMERICAN FOUNDRYMAN.

For the past several years special letters have been prepared over the signature of the President and sent to Personal, Company and Sustaining Members whose dues are renewable on July 1, suggesting conversion to a higher membership classification, or increasing the Sustaining Membership support. President Kuni-ansky's letter of suggestion thus far has developed six conversions from Personal to Company Memberships, six conversions from Company to Sustaining Memberships.

### Chapters

During the past year two new regular Chapters and five Student Chapters were approved by the Board of Directors. These include, in order of approval: the Tennessee Chapter at Chattanooga, and the Eastern New York Chapter at Albany; Student Chapters: University of Minnesota, Minneapolis; Ohio State University, Columbus; University of Missouri School of Mines and Metallurgy, Rolla, Missouri; Massachusetts Institute of Technology, Boston; and Oregon State College at Corvallis, Oregon. As of June 30, 1948, the five Student Chapters included 133 members total.

Our 40th regular Chapter now apparently is well under way at Kansas City, Missouri, with a Steering Committee formed and intending to call a general meeting at an early date. In addition, our sixth Student Chapter has been established, awaiting only the approval of the new Board of Directors on a petition already received.

Attention again should be called to the non-existence of Chapters in two important foundry areas; namely, Pittsburgh, Pa. and New England. No steps have been taken or interest of any kind shown with reference to a Chapter in Pittsburgh, and no further activities have developed in New England since the Secretary met with the Executive Committee of the New England Foundrymen's Association in Boston in October, 1947. Chapters in these two areas alone, it is estimated, would result in a minimum increase of 300 members within one year.

It is the intention of the Staff during the coming year to work toward establishment of Student Chapters at all universities now benefiting under the Foundry Educational Foundation, the list including Cornell University, University of Wisconsin, Case Institute of Technology at Cleveland, University of Cincinnati, and Northwestern Technological Institute, Evanston, Ill.

### Chapter Problems

During the past two years the President, the Vice-President and the Secretary have visited practically every Chapter, sometimes at regular Chapter meetings, and in practically every case, with the Chapter Boards of Directors. Out of these visits has come a greater understanding between the Chapters and the National organization, emphasizing in particular the need for keeping the Chapter records of membership entirely up to date and accurate.

As a result, the National Office has purchased and installed equipment and has worked out a method of membership handling which should accomplish practically all Chapter purposes in the way of information and records, on a much more efficient basis. Necessarily, this new system has had to be installed on a basis of "changing horses in the middle of the stream," and errors have occurred; but it is believed that the system, as it is now working, is a tremendous improvement over any system operative in the National Office since the membership began a very rapid climb about five years ago.

At the same time, with some 60 per cent of all memberships renewable on July 1, at the beginning of a short-staffed vacation period, the Staff is of the definite belief that steps must be taken to spread the load over the entire year, thus making for more efficient handling and necessarily decreasing our costs of operation. For the past five years it has been necessary to employ temporary clerical help during the summer vacation months in order to keep up with the game, and this relatively untrained personnel, without an absolute system of operation, has resulted in most of the errors about which our Chapters have complained.

### Chapter Visits

During the past three years the burden of visits to Chapters by our National Officers has become of increasing importance, especially as the number of Chapters has been constantly growing. Thus far the National Officers necessarily have timed their



visits in accordance with Chapter requests, which frequently have been overlapping territorially, placing a great burden on the National Officers' time.

The suggestion has been made that the National Officers confine their Chapter Visits in large part to meetings with Chapter Boards of Directors with or without any member of the National Office present, scheduling such visits in advance to the best advantage. This should represent a real improvement. Nevertheless, the problem must still be faced that if official visits to all or most of the Chapters must be made each year, a policy may have to be established by the Board of Directors as to how such visits can best be made . . . that is, by one or more of the National Officers, by members of the existing National Staff, or by (as has been suggested more than once) a special "Chapter Contacts Man."

We again recommend that this matter be given Board consideration during the coming year in setting a policy for the future, for without such a policy, the task of the AFS Nominating Committee in selecting chief officers of AFS who can give the time to the job, may become increasingly difficult.

### Chapter Chairman Conference

The 5th Annual Chapter Chairman Conference was staged at the Stevens Hotel, Chicago on June 28-29-30, and with invitations extended for the first time to both the incoming Chapter Chairman and Vice-Chairman, a total of 66 Chapter Officers attended, including 58 Chairmen and Vice-Chairmen. Only the Toledo and Mexico City Chapters were not represented. Of the total number of 78 possible Chairmen and Vice-Chairmen, 74 per cent, or 58, were present at the Conference.

From the above figures, it is evident, even considering the growing importance of the Conferences, that the full intent of the double invitation to incoming Chairmen and Vice-Chairmen, cannot be realized. It has become customary for AFS to permit other Officers and even Directors to attend the Conferences at AFS expense, if only to have a Chapter represented. We recommend, therefore, that invitations hereafter be limited strictly to Officers of the Chapters and not to any other Chapter representatives. The value to the Chapter and to the Association of the attendance of miscellaneous Chapter Directors may be questionable.

For the second consecutive year the Chapter Chairman Conference covered two and a half days. However, a number of those present left at the end of the second day, not being able to spend the full time at the Conference. We recommend, therefore, that future Conferences be limited to two full days and that an intensive program be provided for each day and evening.

It is unfortunate that the Conferences must be held during the latter part of June, at the beginning of the vacation periods. To hold the Conferences later would not serve the purposes for which these meetings were intended. To hold them earlier would make it impossible to issue invitations, for practically all Chapters elect their new officers either in May or June. It is believed, therefore, that the last week in June should be established as the week for the Chapter Chairman Conference.

### American Foundryman

AMERICAN FOUNDRYMAN has now completed three full years in its present form, with regular commercial paid advertising, and experience of the past fiscal year indicates that the magazine is now established in the Industry on a fairly solid basis. For example, gross advertising revenue for these three years was as follows:

1945-46	— \$95,000
1946-47	— 85,000
1947-48	— 96,000

The first year of operation had the benefit of considerable "support" advertising. The second year showed a decrease when such "support" was dropped as it was realized that advertising was permanently established. The past year has shown an encouraging increase over the previous year and indicates that the increase can be maintained so long as industrial conditions do not affect advertising generally. There still exists considerable of the traditional opposition to acceptance of paid advertising by AFS as a Technical Society, and the fact that some companies who originally were a party to this opposition have since become advertisers, does not by any means indicate that any considerable degree of opposition, organized or individual, has been won over.

It is our intention to continually improve the editorial content of the publication and we believe that its value to the membership has been very materially increased during the past six months under the editorship of H. F. Scobie. The quality and quantity of technical articles obtained for AMERICAN FOUNDRYMAN has steadily improved, now enabling us to schedule entire issues from four to six months in advance, and thus to promise authors of papers definite schedules for the publication of their manuscripts. In this connection, it has been established as a rule of operation that no manuscript will be retained in this office for more than 60 days without either being definitely scheduled for publication in a certain issue or else returned to the author with our thanks for submission.

Of particular editorial interest are the following: (1) the establishment of a *Question-and-Answer* Round Table section effective with the July issue, and (2) the keying of *New Products* and *New Literature* sections in the magazine, which produced 132 traceable inquiries during the first four months since this practice was adopted.

Advertising rates in AMERICAN FOUNDRYMAN were increased on January 1, 1948 to \$225 basic page rate, from the previous \$195. The full effect of this increase will not be felt before the calendar year 1949, since advertisers current prior to January 1, 1948, necessary have been protected at the old rate during the calendar year 1948. This increase in rates was the first since regular advertising was established in May, 1945 and follows action taken by practically every other publication, in view of constantly rising production costs. It is expected that further rate increases may be recommended for AMERICAN FOUNDRYMAN as the membership (i.e., circulation) constantly increases and production costs rise still further.

Total circulation of AMERICAN FOUNDRYMAN on June 30, 1948 was 10,529, including 126 paid subscriptions and 10,403 member subscriptions, compared with 9,791 on June 30, 1947—a net increase of 738. During the coming year efforts shall be directed toward materially increasing the number of paid subscriptions to AMERICAN FOUNDRYMAN, especially in non-Chapter areas, to the end that our circulation (membership and paid) may include a greater percentage of the total number of foundries in the industry. This broader coverage is necessary from the standpoint of advertising promotion as well as membership promotion, since the best membership prospect is one acquainted with AMERICAN FOUNDRYMAN and AFS activities.

### 1948 Convention and Exhibit

The 52nd Annual Meeting of the Society was staged at Philadelphia May 3-7, 1948 in conjunction with one of the finest Foundry Shows held by the organization. An estimated attendance in excess of 13,000 was recorded and benefited from the extensive technical program and displays of equipment, materials and services offered by 273 exhibiting companies. A total of 91,000 square feet were occupied by all exhibits, of which 82,000 sq ft were sold at the rate of \$2.50 per sq ft, representing a gross exhibit rental revenue of \$205,420, compared with 1946 totals of 273 exhibitors, 99,433 sq ft occupied, 92,000 sq ft sold at \$1.50 per sq ft, Gross Revenue \$142,260.

Of particular interest at this Foundry Show were: The operating exhibit of the Murrell Dobbins Vocational Training School of Philadelphia, and of six Foundry Societies (Foundry Educational Foundation, Gray Iron Founders Society, Non-Ferrous Founders Society, National Founders Association, Foundry Equipment Manufacturers Association, and Meehanite Metals Group). These same organizations exhibited in 1946, and in addition the Magnesium Association, Malleable Founders Society and American Society for Metals.

Appreciation should be expressed to the Officers, Directors and Committees of the Philadelphia Chapter for their complete cooperation toward the success of the 1948 Convention and Exhibition. In addition, the Society had excellent cooperation from both the Auditorium management and the local Convention & Visitors Bureau, who handled our housing arrangements. Also, and by no means in lesser degree, appreciation should be expressed to the hundreds of authors, session chairmen, vice-chairmen and discussion leaders, who made successful the entire technical program.

Discussion has been held on the advisability of staging the Biennial Exhibits of AFS on odd-numbered instead of even-

numbered years, and of staging these Exhibits in the fall instead of the spring. It is recommended that further consideration be given these possibilities by the Board of Directors during 1948-49.

Beginning with the 1946 Convention and Exhibit, some 20 or more members of the Staff have been taken to the Convention city for servicing the entire program, necessarily at considerable expense to the Society. Experience shows that ordinarily this number will not be required at future conventions, especially in non-Exhibit years, even though a heavier-than-usual staff was necessary in Detroit because of dual and triple registration and service facilities being maintained. With the present trained Staff it is expected that advance planning will make unnecessary the attendance at future conventions of a maximum number.

### By-Laws Amendments

During the past year the By-Laws of the Society were generally revised by a Committee under the Chairmanship of Past President H. Bornstein, and following approval by the Board of Directors, were submitted to the Membership through letter ballot, ballots being returned May 15, 1948. A total of 2,223 ballots were cast by some 22 per cent of the Membership voting, of which 1,986 full ballots were cast, and of these 1,946 full ballots were cast for all Amendments and 40 against all Amendments. A total of 2,219 ballots were accepted for count, and all Amendments were approved by the membership by overwhelming majorities, the new By-Laws thus became effective July 1, 1948.

Included in the Amendments was a change in the name of the Association to American Foundrymen's Society, and articles of incorporation have since been amended to accomplish this change of name.

It is recommended that consideration be given further Amendments to the Articles of Incorporation by the Board of Directors during 1948-49, since these Articles were prepared in 1916 and their reading today is considerably out of date with the purposes of AFS as a Technical Society.

### Staff Retirement Plan

Following original motion by the Board of Directors in July, 1946, negotiations for the establishment of a Staff Retirement Plan proceeded during the past year and a plan was approved by the Board of Directors in January, 1948. Since then the American Foundrymen's Society Retirement Trust has been set up with four Trustees appointed by the Executive Committee, subject to Board approval, to initiate the operations of the Staff Retirement Plan beginning July 1, 1948. Thus, with the exception of a few remaining details, the Society, for the first time, has a Staff Retirement Plan in operation.

The members of the National Office Staff desire to express to the Board of Directors appreciation for the consideration given in thus setting up a retirement plan, with provisions in keeping with those of similar plans in effect in industry today.

### Staff Organization

During the past year 4 employees were added to the National Office Staff in Executive Capacities. These include John D. Murray, Advertising Manager of AMERICAN FOUNDRYMAN; A. A. Hilbron, Convention & Exhibit Manager; R. N. Sheets, Assistant Editor of AMERICAN FOUNDRYMAN, and Philip D. Johnson, Educational Assistant of the Technical Department. These employees should materially strengthen the service rendered by AFS to the membership and to the industry.

### Technical Department

Activities of the Technical Department covering Committee work, Convention Technical Program, technical publications, apprentice training and educational activities, are not covered by this report, since they are fully dealt with in the report of the Technical Director for 1947-48.

Respectfully submitted,

WM. W. MALONEY, *Secretary-Treasurer*



## Report of the Technical Director for the Fiscal Year 1947-48

PROGRESS REPORTS WERE MADE during the year to the Executive Committee and the Board of Directors at their several meetings that they would be fully informed of important activities. The most important phases of our technical activities are briefly covered in the following report so as to consolidate the entire year's work.

### Technical Activities

Comprehensive examination of the technical committee structure was initiated approximately a year ago. This re-organization has now been completed. It involved elimination of certain committees which have become obsolete and activation of others. The committee structure now constitutes a well integrated organization headed by the Technical Correlations Committee as the policy-making group. Based upon observations over the past year, there is adequate evidence that the committees are functioning more efficiently and their efforts are considerably more specific.

A complete National Committee Personnel Roster was published and sent to all members of the Society, and in addition to the executive officers of each of the companies whose staff members are represented on our technical committees. Numerous complimentary letters were received from these executives attesting to the value of the Roster. The Roster likewise includes Suggested Rules for Division and Committee Procedure, assisting in the more orderly functioning of the committees.

Following the adoption of the clearly defined Publication Policy and logical procedure for the review of technical papers the Program and Papers Committees of the several Divisions are functioning efficiently in developing a well-planned technical program of a high standard of quality. The review of technical papers is now systematic and thorough and thereby lends prestige to A.F.S. conventions and its technical publications.

The Educational Division, during the last year, formed a Text Book Committee, and authors have been selected for the preparation of a text book for the high school and college levels. The College Research Projects Committee has completed its list of research projects and these will be published in the near future for distribution to the several educational institutions throughout the country interested in foundry work. The Apprenticeship Contest was unusually successful at the last Convention, both in terms of contestants and their geographical distribution. Active liaison is being maintained with the Foundry Educational Foundation. At its last Executive Committee meeting, the Educational Division developed a well planned program for the coming year.

### A.F.S.-Sponsored Research

At the present time a total of eight individual research projects are underway and in several instances progress reports were presented at the Convention covering the findings up to that time. Under each of these projects quarterly reports are submitted for review by the sponsoring committee to familiarize them with the research work and place them in the position to give helpful advice in the future prosecution of the project. These projects comprise the following:

- Sand Research—Cornell University
- Brass and Bronze—University of Michigan
- Malleable—University of Michigan
- Aluminum and Magnesium—Battelle Memorial Institute
- Heat Transfer—Columbia University
- Steel—Armour Research Foundation
- Centrifugal Casting of Light Alloys—Canadian Bureau of Mines
- Cupola Research—Under the guidance of the Cupola

Research Committee, utilizing a full time research worker and at present a cooperative research involving Ford Motor Company, Battelle Memorial Institute, Canadian Bureau of Mines and the United States Bureau of Mines.

The Sand Research and the Aluminum and Magnesium Research Projects are near the close of the first year's contract, and following recommendation of the Committee and approval by the Board of Directors funds have been already made available for the continuation of these projects for another year.

It is thus noted through the foresight and appreciation of research, the Board of Directors have made possible these several research projects which are all progressing in a very satisfactory manner.

### Convention Program

Following adoption of a Publication Policy and establishment of a well-planned and orderly procedure for the review of technical papers by the several Program and Papers Committees the Convention Technical Program for 1948 developed in a very satisfactory manner. Over 90 per cent of the technical papers presented at the Convention were preprinted and made available to all members on request more than 30 days in advance of the Convention, affording ample opportunity for the review of the paper and the presentation of constructive written and oral discussion. At the present writing several papers have already been submitted for consideration by the Program and Papers Committee for the 1949 Convention.

Plans are now being made for a jointly-sponsored Symposium on Centrifugal Casting with the American Society for Testing Materials. This Symposium will be scheduled during the 1949 Convention of our Society.

### Technical Publications

The 1946 Annual TRANSACTIONS, vol. 54, comprising a total of approximately 950 pages was available for distribution to the membership during November, 1947. Volume 55, the 1947 Annual TRANSACTIONS, representing approximately 700 pages was available for distribution to the membership during April of this year. Paper-bound copies of the latter TRANSACTIONS were made available to the membership gratis on request and a total of approximately 2,100 of the paper-bound copies were sent to members requesting same. Thus, for the first time, the publication of the Annual TRANSACTIONS is current for the year in which the papers were presented. It is contemplated that vol. 56, covering the 1948 Annual Convention will be made available to members during the month of December, 1948.

At the beginning of the year, ANALYSIS OF CASTING DEFECTS was published and offered for sale. Ten thousand copies were printed, of which approximately 2,100 have already been sold, giving ample evidence of the excellent reception which this book has received. In addition, the RECOMMENDED PRACTICES FOR SAND CASTING ALUMINUM AND MAGNESIUM ALLOYS was published and made available at the Annual Convention in Philadelphia.

Several books are in process of revision and manuscripts will be received from the committees doing this work in the near future. The revised manuscript of the CORE PRACTICE BOOK has just been received and after editing will be submitted to the printer for quotation. It is anticipated that the final manuscript for THE MICROSCOPE IN ELEMENTARY METALLURGY will be ready for publication this Fall. The Committee on Recommended Practices for Malleable Castings has likewise completed its manuscript and it is now in our hands ready to prepare for the printer. The book on PRECISION INVESTMENT CASTING is being energetically pursued by the Committee and will be available for publication some time

within the coming year. The book, *THE DEVELOPMENT OF THE METALS CASTING INDUSTRY* is in process of printing at the present time and will be offered for sale in the near future. Under these circumstances it becomes apparent that the publication load will be heavy during the coming year, and that numerous revised or new publications will be available to the membership, satisfying one of the primary purposes of the Society.

### Cooperation With Other Societies

Active cooperation is maintained with numerous technical societies who have common interests and make such liaison of mutual interest. This liaison is maintained by the Technical Director with the American Society of Mechanical Engineers, the American Society of Metals, the American Society for Testing Materials, the American Institute of Mining and Metallurgical Engineers, Society of Automotive Engineers, Army Ordnance Association, the American Industrial Radium and X-Ray Society

and several foreign foundry societies with which we have maintained intimate contact.

### Technical Inquiries

During the period June 30th, 1947 to June 30th, 1948 the Technical Staff has replied to 521 individual technical inquiries received from members or others interested in matters relating to the castings industry.

The Technical Staff is highly appreciative of the continued support and understanding cooperation received from the Executive Committee and the Board of Directors of the Society. Their counsel and encouragement have been extremely helpful in achieving progress and added interest in the technical affairs of A.F.S.

Respectfully submitted,

S. C. MASSARI, *Technical Director*

## Comparative Condensed Balance Sheet

As of June 30, 1946, 1947 and 1948

ASSETS	June 30, 1946	June 30, 1947	June 30, 1948	LIABILITIES	June 30, 1946	June 30, 1947	June 30, 1948
Cash.....	\$131,341.19	\$ 58,216.09	\$106,394.93	Accounts Payable.....	\$ 18,197.03	\$ 26,618.98	\$ 21,397.81
Securities.....	181,949.60	181,949.60	178,805.60	Reserves.....	51,069.06	52,134.53	53,263.93
Accounts Receivable..	9,379.56	9,457.28	10,003.03	Accrued Items.....	2,013.17	733.05	1,128.23
Inventories.....	33,182.95	23,898.63	33,167.40	Deferred Income.....	12,991.53	24,264.13	61,439.16
Prepaid Expenses.....	1,780.21	13,815.04	19,721.40	Unexpended Special			
Furniture & Fixtures..	3,765.59	6,354.21	10,744.57	Funds.....	164,699.80	174,346.16	164,028.57
				Surplus.....	112,458.51	15,594.00	57,579.23
<b>TOTAL.....</b>	<b>\$361,399.10</b>	<b>\$293,690.85</b>	<b>\$358,836.93</b>	<b>TOTAL.....</b>	<b>\$361,399.10</b>	<b>\$293,690.85</b>	<b>\$358,836.93</b>

## Condensed Statement of Income & Expense

(All Expenses Distributed to Major Activities)

Fiscal Year July 1, 1947-June 30, 1948

### INCOME

Membership Dues.....	\$167,140.48
Publications.....	35,855.27
AMERICAN FOUNDRYMAN.....	96,327.00
Convention & Exhibit 1948.....	216,327.00
Other Income.....	501.55
<b>Total.....</b>	<b>\$516,668.80</b>

### EXPENSE

MAJOR ACTIVITIES	Total Expense	
	Amount	Per Cent
General Administration.....	\$113,366.18	23.9
Chapter Activities.....	64,035.29	13.5
Technical Activities & Publications.....	127,121.91	26.8
AMERICAN FOUNDRYMAN.....	96,290.10	20.3
Convention & Exhibit.....	73,521.99	15.5
	<b>\$474,335.47</b>	<b>100.0</b>

### Comparative Membership Report As of June 30, 1944-1948 Inclusive

	June 30, 1944	June 30, 1945	June 30, 1946	June 30, 1947	June 30, 1948
Sustaining Members.....	169	239	183	193	199
Company Members.....	1,471	1,571	1,462	1,548	1,525
Personal Members.....	4,720	5,722	6,426	7,358	7,860
Student and Apprentice Members.....	42	23	20	89	307
Honorary Life Members.....	55	55	58	64	63
Foreign Members.....	163	205	390	430	449
<b>Total.....</b>	<b>6,620</b>	<b>7,815</b>	<b>8,539</b>	<b>9,683</b>	<b>10,403</b>
New Members, 12 months.....	1,510	1,803	1,743	2,164	1,979
Resignations.....	215	235	370	186	286
Delinquents dropped.....	316	356	631	829	948
Removed by death.....	24	17	18	5	25
Net gain for year.....	955	1,255	724	1,145	720
Members in Chapters.....	5,902	7,111	7,767	8,809	9,519

billion  
0, 1948

Chapter	Sustaining	Company	Personal	Affiliate	Associate	Student and Apprentice	Honorary and Life Membership	Totals
Birmingham.....	4	37	69	182	6	2	1	301
British Columbia.....	0	16	17	44	5	2	0	84
Canton District.....	0	32	34	97	4	0	0	167
Central Illinois.....	1	10	21	80	3	8	1	124
Central Indiana.....	3	38	51	173	4	6	0	275
Central Michigan.....	0	11	30	69	0	0	0	110
Central New York.....	1	34	32	84	3	5	0	159
Central Ohio.....	1	31	32	121	2	0	1	188
Chesapeake.....	3	28	51	85	21	0	4	192
Chicago.....	22	118	149	491	16	22	9	827
Cincinnati District.....	9	49	81	128	6	8	1	282
Detroit.....	15	59	132	313	19	14	5	557
E. Canada and Newfoundland.....	7	50	78	180	17	4	1	337
Eastern New York.....	0	12	12	47	3	3	0	77
Metropolitan.....	10	71	118	184	14	4	2	403
Mexico City.....	0	11	33	14	2	0	0	60
Michiana.....	3	40	44	120	4	3	0	214
Northeastern Ohio.....	18	85	131	278	14	11	9	546
Northern California.....	8	28	97	100	13	3	0	249
No. Illinois and So. Wisconsin.....	4	20	12	53	2	0	1	92
Northwestern Pennsylvania.....	2	25	29	83	1	5	0	145
Ontario.....	2	81	113	111	2	1	1	311
Oregon.....	2	10	30	67	4	2	0	115
Philadelphia.....	18	67	110	195	12	0	3	405
Quad City.....	4	36	56	137	5	2	2	242
Rochester.....	2	13	44	60	4	0	0	123
Saginaw Valley.....	7	19	18	215	3	19	2	283
St. Louis District.....	3	47	99	138	8	1	0	296
Southern California.....	3	57	128	119	13	2	1	323
Tennessee.....	0	12	22	71	1	0	0	106
Texas.....	6	31	39	58	2	0	0	136
Timberline.....	0	6	50	23	6	0	0	85
Toledo.....	0	20	34	46	2	2	1	105
Tri-State.....	0	11	50	46	3	5	1	116
Twin City.....	5	42	52	123	7	2	0	231
Washington.....	0	14	30	24	6	2	0	76
Western Michigan.....	5	25	28	159	0	0	2	219
Western New York.....	1	38	82	82	3	1	1	208
Wisconsin.....	17	89	121	351	7	10	2	597
<b>Total.....</b>	<b>186</b>	<b>1,423</b>	<b>2,359</b>	<b>4,951</b>	<b>247</b>	<b>149</b>	<b>51</b>	<b>9,366</b>
<b>STUDENT CHAPTERS</b>								
University of Minnesota.....	—	—	—	—	—	20	—	20
Ohio State University.....	—	—	—	—	1	31	—	32
Missouri School of Mines.....	—	—	—	—	2	26	—	28
Massachusetts Institute of Technology.....	—	—	—	—	—	42	—	42
Oregon State College.....	—	—	—	—	—	31	—	31
<b>Total Student Chapters.....</b>	<b>—</b>	<b>—</b>	<b>—</b>	<b>—</b>	<b>3</b>	<b>150</b>	<b>—</b>	<b>153</b>
<b>Total All Chapters.....</b>	<b>186</b>	<b>1,423</b>	<b>2,359</b>	<b>4,951</b>	<b>250</b>	<b>299</b>	<b>51</b>	<b>9,519</b>
Foreign.....	0	28	390	17	11	2	1	449
Non-Chapter.....	13	102	149	109	42	8	12	435
<b>GRAND TOTAL.....</b>	<b>199</b>	<b>1,553</b>	<b>2,898</b>	<b>5,077</b>	<b>303</b>	<b>309</b>	<b>64</b>	<b>10,403</b>

## Minutes Special Meeting 1947-48 Board of Directors

Palmer House, Chicago—July 29, 1947

Present: President Max Kuniansky, presiding  
 Directors: E. H. Horlebein R. H. McCarroll  
 H. H. Judson John Robb, Jr.  
 J. H. Smith A. C. Ziebell  
 F. M. Wittlinger H. A. Deane  
 S. V. Wood B. L. Simpson  
 E. H. Delahunt S. C. Wasson  
 W. J. MacNeill  
 Secretary-Treasurer Wm. W. Maloney  
 Technical Director S. C. Massari

1. President Kuniansky called the new Board of Directors to order and stated the sole purpose of this special meeting was to announce, prior to the regular meeting of the 1947-48 Board on July 30, the formation of his Executive Committee. He requested Board approval for the following members of his Executive Committee:

Chairman, President Max Kuniansky  
 Vice President W. B. Wallis  
 Director & Past President S. V. Wood  
 Director J. H. Smith  
 Director S. C. Wasson  
 Director John M. Robb, Jr.

On motion by Director Horlebein, seconded by Director McCarroll, the Executive Committee was approved as presented.

2. President Kuniansky then declared that no further business would be taken up at this special meeting and the meeting was declared adjourned.

Respectfully submitted,  
 WM. W. MALONEY,  
 Secretary-Treasurer

APPROVED:  
 MAX KUNIAISKY, President

## Minutes First Meeting 1947-48 Board of Directors

Palmer House, Chicago—July 30, 1947

Present: President Max Kuniansky, presiding  
 Directors: E. W. Horlebein B. L. Simpson  
 H. H. Judson E. H. Delahunt  
 J. H. Smith W. J. MacNeill  
 F. M. Wittlinger R. H. McCarroll  
 S. V. Wood John M. Robb, Jr.  
 H. A. Deane A. C. Ziebell  
 Secretary Wm. W. Maloney  
 Technical Director S. C. Massari

Also present as members of the Advisory Board: Past Presidents F. J. Walls, L. C. Wilson

Absent: Vice-President Wallis, Directors Kolb, Lamker and Wasson; also R. Gregg, newly appointed.

### Election of Executive Committee

The Executive Committee of the Board of Directors was duly elected at the Special Meeting of the 1947-48 Board of Directors, July 29, 1947 and consists of the following Board members: Chairman, President Max Kuniansky, Vice President W. B. Wallis and Directors S. V. Wood, J. H. Smith, S. C. Wasson and John M. Robb, Jr.

### Staff Appointments

President Kuniansky called the meeting to order and immediately appointed Directors Simpson, Smith and Wood as a Board Nominating Committee to recommend candidates for election as Secretary and Treasurer, as called for by the By-Laws, Art. XVI, Sec. 1. The committee retired and, in executive session, Secretary-Treasurer Wm. W. Maloney was declared re-elected for the year 1947-48.

Continuing in executive session, the Board approved salary authorizations for the fiscal year 1947-48 as recommended by the Finance Committee, S. V. Wood, Chairman.

### 1947-48 Budget

The Finance Committee, S. V. Wood, Chairman, presented for consideration a Budget of Estimated Income and Expense for the fiscal year July 1, 1947 to June 30, 1948. The Secretary-Treasurer was asked to comment on items of the budget as presented and, following brief discussion, a motion was made by Director Horlebein, seconded by Director Judson and carried, approving the 1947-48 budget as recommended.

### Standing and Special Committee

President Kuniansky entertained a motion authorizing presidential appointment of Standing and Special Committees of the Board. A required motion was made by Director Judson, seconded by Director Wittlinger, and carried.

President Kuniansky then announced appointments to Standing and Special Committees of the Board as follows:

Finance Committee—Past President S. V. Wood, Chairman, President Kuniansky, Vice President Wallis.

National Membership Committee—Vice President Wallis, Chairman, All Chapter Membership Committee Chairmen as members.

Chapter Contacts Committee—Director Simpson, Chairman, All other directors as members.

By-Laws Committee—H. Bornstein, Chairman, Vice-President Wallis, N. J. Dunbeck, L. C. Wilson, F. W. Shipley, D. C. Zuege, J. A. Wetherspoon, Secretary Maloney, *ex officio*.

Annual Lecture Committee—H. Bornstein, Chairman, H. M. St. John, G. Vennerholm, C. F. Joseph, R. E. Ward.

\* Technical Correlations Committee—Past President Wood, Chairman, President Kuniansky, Vice-Chairman.

International Relations Committee—Secretary Maloney, Chairman, V. Delpont, England, European representative, Miguel Siegel, Brazil, South American representative, Wm. A. Gibson, Australian representative.

Representation on National Castings Council—President Kuniansky, Vice-President Wallis.

\* Technical Correlations Committee—The Secretary pointed out that, by Board action in 1946, the immediate Past-President had been designated Chairman of the Technical Correlations Committee, with the incumbent Vice-President as Vice-Chairman; but that this provides no year-to-year continuity from the Board. Accordingly Director Horlebein moved

THAT Board representation on the Technical Correlations Committee henceforth consist of the immediate past-President of AFA as Chairman, with the incumbent President of AFA as Vice-Chairman superseding previous Board action. Motion was seconded by Director Simpson and carried.



### 1949 Convention City

The Board considered recommendations of the 1946-47 Board on the question of staging a Convention and Exhibit on the West Coast in 1949. On motion by Director Judson, seconded by Director Horlebein and carried, the recommendations of the Old Board were approved, reaffirming as a Board policy the staging of Exhibits not more frequently than every two years.

At this point, as called for by previous Board action the Secretary presented for consideration the following resolution:

WHEREAS the American Foundrymen's Association in recent years has invited Manufacturers and Suppliers of Foundry Equipment and Materials to exhibit their products and services at Annual Conventions of the Association at two-years intervals, and

WHEREAS the staging of AFA Exhibits biennially, maintained as a continuous practice since 1938, has come to be recognized as a real or implied policy of the Association, and

WHEREAS the Board of Directors of the American Foundrymen's Association believes it to be to the best interests of the Association, the Exhibitors, and the Foundry Industry as a whole to continue this exhibit practice, now therefore be it

RESOLVED, that this Board of Directors of the American Foundrymen's Association approves and establishes as a stated policy the staging of foundry exhibits at intervals not more frequent than every two years, and recommends that this policy be reaffirmed by successive Boards of Directors of the Association.

#### APPROVED:

MAX KUNIANSKY, President  
For the Board of Directors  
American Foundrymen's Association

On motion by Director Judson, seconded by Director Horlebein the above resolution was approved unanimously by the Board of Directors.

The Board then considered recommendations of the 1946-47 Board pertaining to an invitation from the Southern California Chapter for staging of a Convention and Exhibit in Los Angeles in 1949, the old Board having recommended that, in view of the approved resolution restaging Exhibits of AFA not more frequently than every two years, this invitation could not be accepted for 1949.

On motion by Director Judson, seconded by Director Horlebein the recommendation of the 1946-47 Board, authorizing the Secretary to notify the Southern California Chapter of the Board's decision, was accepted.

### Convention Registration Fees

The Board considered and approved, on motion by Director Judson, seconded by Director Horlebein and carried, recommendation of the 1946-47 Board that the question of registration fees at AFA Conventions in non-exhibit years be deferred until the formulation of definite plans concerning the next non-exhibit Convention.

The Secretary requested Board action on the question of the registration fees at the 1948 Convention and Exhibit in Philadelphia, pointing out that in 1944 and 1946 the registration fees charged were: Members, gratis; non-members, \$1.00 each. A motion was made by Director Horlebein,

THAT registration fees for the 1948 Convention and Exhibit in Philadelphia be continued as in effect at the 1946 Convention and Exhibit. Seconded by Director Judson, and carried.

### Centrifugal Castings Research

The Board considered recommendations of the 1946-47 Board approving an expenditure of \$1,000.00 during 1947-48 for studies on the Centrifugal Casting of Light Alloys, as a project of the Centrifugal Castings Committee of the Aluminum and Magnesium Division. On motion by Director Judson, seconded by Director Horlebein and carried, the old Board's recommendation was accepted.

### Brass and Bronze Research

The Board considered recommendations of the 1946-47 Board for a maximum expenditure of \$5,000.00 during 1947-48 for a study of the fracture test as an indication of the quality of tin bronze, as an AFA-sponsored research project for the Brass and

Bronze Division. On motion by Director Judson, seconded by Director Horlebein and carried, the old Board's recommendation was accepted.

### Malleable Iron Research

The Board considered recommendations of the 1946-47 Board for a maximum expenditure of \$5,000.00 during 1947-48 for a study on the most suitable microstructure for pearlitic malleable iron castings preparatory to selective hardening, as an AFA-sponsored research project of the Malleable Iron Division. On motion by Director Judson, seconded by Director Horlebein and carried, the old Board's recommendation was accepted.

### Preparation of Text Books

The Board considered recommendations of the 1946-47 Board that AFA undertake the preparation of several textbooks on foundry practice for use at college, trade school, vocation school and high school levels, such projects having been recommended by the Technical Correlations Committee. On motion by Director Judson, seconded by Director Horlebein and carried, the old Board's recommendation was accepted.

### International Foundry Committee

The Board considered recommendations of the 1946-47 Board that AFA approve acceptance of a membership application to the International Committee of Foundry Technical Associations from a reorganized Italian Foundry Technical Association. On motion by Director Judson, seconded by Director Horlebein and carried, the old Board's recommendation was accepted.

### Extension of Retirement Compensations

The Board considered recommendations of the 1946-47 Board that AFA enter into a contract with C. E. Hoyt and R. E. Kennedy, retired, for a period of 7 years from July 1, 1947, making provision for extensions of retirement compensations for Messrs. Hoyt and Kennedy as follows: Extension of present retirement compensations for the period of the contract, with provision that such compensations might be reduced by future Boards of Directors, by not more than 50 per cent, if economic conditions warrant, compensations to terminate in the event of death of the recipients.

Following discussion, motion was made by Director Simpson, seconded by Director Deane and carried, accepting the recommendations of the 1946-47 Board of Directors. Director Judson dissenting.

On motion by Director Simpson, seconded by Director Deane and carried, the Board also approved the recommendation of the 1946-47 Board that resolutions adopted by the Board of Directors on July 19, 1945 pertaining to continuation of retirement compensations for Messrs. Hoyt and Kennedy be made a part of Board Minutes.

Director Simpson then moved

THAT the President reappoint a Committee on Retirement for the express purpose of preparing, for acceptance by Messrs. Hoyt and Kennedy, the contract called for by Board action above.

Motion seconded by Director Deane and carried. President Kuniansky then appointed a Retirement Committee consisting of Director Simpson, Chairman, Vice President Wallis, and Director Wasson.

### Staff Retirement Plan

To progress Board action on the Staff Retirement Plan, B. Russell Thomas of Marsh & McLennan, Chicago, insurance actuaries, presented revised proposals, as requested by the Executive Committee at the July 28 meeting. Discussion brought out the fact that such a plan will be entirely voluntary on the part of employees. In response to question, Mr. Thomas stated that in his estimation the plan proposed was in line with similar plans in industry as to employee versus employer contributions, retirement benefits, eligibility and other provisions. Mr. Thomas also pointed out that while the proposed plan made no allowance for Association service prior to adoption of the plan, this could be worked out so as to include more employees under eligibility requirements.

President Kuniansky pointed out that it would be impracticable for the Board to consider thoroughly and approve the proposals at this Board Meeting, but that further consideration

was contemplated for the scheduled meeting of the Executive Committee on November 3, with full discussion at the mid-year Board Meeting in January, 1948, for possible adoption of an approved Staff Retirement Plan on July 1, 1948. He therefore urged all Directors to carefully study the proposals of Marsh & McManis, and submit their comments in writing to the Secretary in advance of the November 3 meeting of the Executive Committee.

#### *Approval of Bank Resolutions*

The Secretary presented for approval the following resolutions required by the banks for the signing of checks and withdrawal of AFA funds during 1947-48:

RESOLVED that resolutions required by the Harris Trust & Savings Bank of Chicago, and by the Northern Trust Co. of Chicago, authorizing withdrawal of funds, are hereby approved and the Secretary authorized to certify thereto.

RESOLVED that checks for the withdrawal of funds deposited in the name of the Association with depository banks, including all general checking accounts and interest savings accounts, and for the withdrawal of all securities held in the various funds of the Association by the Trust Department of the Harris Trust & Savings Bank of Chicago, shall require the signatures of any two of the following officers: President, Vice-President, Secretary-Treasurer, Technical Director.

RESOLVED that the Board of Directors authorize a Treasurer's expense account of One Thousand dollars (\$1,000.00), said account to be reconciled at the end of each month by full statement of expenditures, and withdrawal checks to be signed by any one of the following: President, Vice-President, Secretary-Treasurer, Technical Director.

RESOLVED that the Secretary be authorized to rent a Safety deposit box at the Harris Trust & Savings Bank of Chicago for the safe-keeping of Association securities, and that any two of the following have authority to obtain access to such safety box: President, Vice-President, Secretary-Treasurer, Technical Director.

RESOLVED that the Secretary be authorized to reimburse travelling expenses for members in attendance at any regularly called Board of Directors, Executive Committee, or Technical Committee meeting, with the following exceptions: No expenses shall be paid to Directors or Committee members for attendance at meetings held during the week of the Annual Convention of the Association, unless specifically authorized by the Executive Committee. When meetings are held in conjunction with other committees or associations the Secretary is authorized to determine what portion of the expense of such attendance shall be paid by the Association.

RESOLVED that the present Blanket Indemnity Bond be renewed covering all Staff members for Five Thousand dollars (\$5,000.00), premiums to be paid by the Association.

Discussion brought out the feeling from the Board that, since both the Secretary-Treasurer and the Technical Director are authorized to sign bank checks, the amount of their bond is at present insufficient. Accordingly, Director Smith moved that the above resolution be amended to read as follows:

RESOLVED that Indemnity Bonds be purchased covering the Secretary-Treasurer and Technical Director for \$50,000.00 each, and all other employees for \$5,000.00 each, premiums to be paid by the Association.

Amendment was seconded by Director Horlebein and carried. Director Horlebein then moved that the above resolutions be approved by the Board with the amendment covering provision for increased bonding of Officers. Motion was seconded by Director Smith and carried.

The Secretary then requested approval of the following resolution representing continuation of authorities voted by previous Boards for the execution of necessary contracts for the Association:

RESOLVED that the Secretary be authorized to execute all contracts for the administration of Association affairs, subject to specific approval by the Executive Committee. In the case of AFA-sponsored research projects, approval of the project by the Executive Committee includes the authority for the Secretary to execute contracts for performance of such projects on a bid basis.

On motion by Director Horlebein, seconded by Director Smith and carried, the above resolution was approved.

#### *Auditing Committee*

The new Board considered action of the old Board toward appointment of an Auditing Committee for examining and making recommendations on the Association's financial condition as of June 30, 1947, an action felt desirable in view of inability to audit the AFA books in sufficient time for complete analysis prior to the Annual Board meeting. On motion duly made, seconded and carried, the new Board authorized retiring President S. V. Wood to appoint an Auditing Committee of three Directors for this purpose, to report back to the new Board with recommendations at a later date.

#### *Expenses of Sand Division Chairman*

The Technical Director called attention to the fact that AFA has in the past reimbursed Dr. H. Ries for travel and incidental expenses as Chairman of the Sand Division, and requested Board Approval of an expenditure for this purpose during 1947-48. He stated that the amount was already included in the 1947-48 budget.

On motion by Director Horlebein, seconded by Director Smith and carried, the expenditure was approved.

#### *1948 Chapter Chairman Conference*

The Secretary requested approval for a 1948 Chapter Chairman Conference, to be the 5th Annual event of its kind, on dates to be announced later. On motion of approval by Director Horlebein, seconded by Director Robb, it was carried.

The Secretary requested Board reactions on the matter of inviting the Chairman and/or the Vice-Chairman as Chapter delegates to the 1948 Chapter Chairman Conference, recalling that the previous Board had voted to invite only the Chairman at AFA expense. Following discussion, including the fact that the Vice-Chairman usually acts as Program Chairman in the Chapters, Director Robb moved

THAT invitations to the 1948 Chapter Chairman Conference be extended to both the Chairman and the Vice-Chairman of each AFA Chapter, both at AFA expense.

Motion was seconded by Director Horlebein and carried.

#### *National Castings Council*

The President pointed out that the next scheduled meeting of the National Castings Council will be held August 4 and that, because of the impossibility of attendance by either the President or the Vice-President, Director Smith has consented to represent AFA. He stated that discussion on a set of proposed by-laws was expected at the August 4 meeting, and asked Board agreement for Director Smith to act on behalf of the Association. Approval granted, without vote.

#### *Foreign Membership Dues and Subscription Rates*

The Secretary called attention to the increasing cost of servicing foreign memberships and subscriptions to AMERICAN FOUNDRYMAN, both of which now are on the same basis as memberships and subscriptions held within the United States. Without vote, the matter of subscription rates was referred to the Executive Committee for discussion, with Staff recommendations as a guide; foreign membership dues to be referred to the By-Laws Committee of the Board for consideration and recommendations.

#### *1948 Exhibit Fees*

The Board considered recommendations of the 1946-47 Board that Exhibit space rates at the 1948 Foundry Show at Philadelphia be established at \$2.50 per square foot, as against \$1.50 per square foot at the 1946 and 1944 Exhibits. On motion by Director Simpson, seconded by Director Judson and carried, the 1948 space rate of \$2.50 per square foot was approved.

The Board also considered the question of continuing the \$25.00 Exhibit permit fee charged all exhibitors at AFA Foundry Shows for many years past. On motion by Director Simpson, seconded by Director Horlebein and carried, a \$25.00 Exhibit permit fee was authorized for each Exhibitor Company at the 1948 Exhibit.

#### *Adjournment*

There being no further business to be considered, the meeting was declared adjourned.

Respectfully submitted,

WM. W. MALONEY,  
Secretary-Treasurer

APPROVED:  
MAX KUNIANSKY, President

## Minutes Special Meeting 1947-48 Executive Committee

Palmer House, Chicago—July 29, 1947

Present: President Max Kuniansky, presiding  
Director & Past President S. V. Wood  
Director J. H. Smith  
Director S. C. Wasson  
Director John M. Robb, Jr.  
Secretary-Treasurer Wm. W. Maloney

The meeting was called to order and it was stated that the purpose of this special meeting was to appoint members of the 1948 Nominating Committee, in view of the fact that the next meeting of the Executive Committee had been scheduled for November 3. However Art. X, Sec. 2 of the Association By-Laws states that "on or before November 1 of each year the Executive Committee of the Board of Directors shall appoint a Nominating Committee. . ."

The Secretary stated that of the 24 Chapters eligible to submit candidates for the Nominating Committee in accordance with provisions of the By-Laws Art. X, Secs. 1-3 inclusive, only 14

Chapters had submitted lists. The lists thus submitted were considered carefully by the Executive Committee giving due consideration to equitable regional representation and to proportional representation for the purpose of division of membership, in the light of representation remaining on the Board after the 1948 Annual Board Meetings.

It was pointed out that the two past Presidents (S. V. Wood and F. J. Walls) automatically become members of the Nominating Committee as provided in Art. X, Sec. 2 of the By-Laws.

(Note: All actions of this Executive Committee meeting are included in the Minutes of the Executive Committee meeting of Nov. 3, 1947).

Respectfully submitted,  
WM. W. MALONEY,  
Secretary-Treasurer

APPROVED:  
MAX KUNIAISKY, President

## Minutes Meeting of Executive Committee

Palmer House, Chicago—November 3, 1947

Present: President Max Kuniansky, presiding  
Vice-President W. B. Wallis  
Directors: James H. Smith  
S. V. Wood  
S. C. Wasson  
John M. Robb, Jr.  
Secretary-Treasurer Wm. W. Maloney  
Technical Director S. C. Massari

### Reading of Minutes

Minutes of the following meetings were read:

Special Meeting 1947-48 Board of Directors, July 29

Special Meeting 1947-48 Executive Committee, July 29

First Meeting 1947-48 Board of Directors, July 30

On motion by Director Smith, seconded by Vice-President Wallis, the Minutes were approved as read.

### Membership Report

The Secretary presented report of membership, showing total of 9741 on September 30 as against 9683 on June 30, 1947, a net increase of 58. AMERICAN FOUNDRYMAN circulation was given as 9849, including 107 subscriptions. The report also showed total dues income to September 30 of \$143,888.08, including \$26,633.63 deferred dues income (paid prior to June 30, 1947).

### Financial Report

The Secretary-Treasurer presented report of finances of the Association as of September 30, 1947, showing Income and Expense as follows, with all overhead distributed to major activities:

	INCOME	
	Budget	Actual
	7/1/47-6/30/48	7/1/47-9/30/47
Membership Dues	\$175,000.00	\$143,888.08
Publications	50,500.00	5,665.65
Advertising	100,000.00	18,852.00
1948 Convention & Exhibit	148,550.00	
Miscellaneous	2,200.00	82.00
	\$476,250.00	\$168,487.73

### EXPENSE

General Administration	\$113,372.00	\$ 20,697.92
Chapter Activities	63,968.00	11,691.30
Publications	65,707.00	12,037.70
Technical Activities	60,866.00	11,171.68
AMERICAN FOUNDRYMAN	96,580.00	17,580.25
Convention & Exhibit	75,757.00	13,423.34
	\$476,250.00	\$ 86,602.19

### Death of Peter Blackwood

The Secretary announced with regret the death of Peter Blackwood, AFA Gold Medalist of 1946, on Wednesday, October 29, burial Saturday, November 1 at Windsor, Ont. Peter Blackwood a native of Scotland, and a graduate of the Royal Technical College at Glasgow in Metallurgy, foundry practice, chemistry, and mining, was awarded the John H. Whiting Gold Medal "for his development work in, and his influence on the free interchange of information pertaining to centrifugal casting." Long connected with the Buick and Pontiac Divisions of General Motors, as Superintendent of Foundries, he took over a similar position with Ford Motor Company at Windsor when that foundry was installed in 1936, where he and his associates conducted important work on the centrifugal casting process.

Expressing its sorrow at the loss of Mr. Blackwood, the Executive Committee instructed the Secretary to communicate its expressions of sympathy to Mr. Blackwood's relatives and to his company.

### President's West Coast Trip

President Kuniansky reported fully on his visit to the West Coast Chapters from October 5-24, inclusive, during which time he visited the new British Columbia Chapter at Vancouver, B.C.; Washington Chapter at Seattle; Oregon Chapter at Portland; Northern California Chapter at San Francisco, and Southern California Chapter at Los Angeles. He was accompanied by the Secretary at the Northern California and Southern California Chapters.

The President recommended that the Secretary seriously consider visits to the West Coast in the future by Technical Director Massari, for the purpose of greater cooperation between the National Organization and the West Coast Groups.



*Annual Meeting of FEMA*

The Secretary reported his attendance at the Annual Meeting of the Foundry Equipment Manufacturers Association on September 18-20, and stated that he had presented there the resolution of the AFA Board of Directors, adopted at the July Board Meeting, expressing the intention of AFA to stage an Exhibit not more frequently than every two years. The Secretary reported that the following return resolution had been ratified unanimously by the FEMA membership present at the above meeting:

RESOLVED, that the action of AFA, by their Board of Directors' resolution of July 30, 1947, submitted by the secretary of AFA to our Executive Secretary, confirms the understanding existing among members of the Association for many years, namely, that the exhibits in connection with AFA National Conventions shall not be held more frequently than every two years; and that this Association is wholeheartedly in accord therewith as being in the best interests of the foundry industry and the AFA, and that the AFA be so advised by copy of this resolution.

*Report on 1948 Convention & Exhibit*

The Secretary and the Technical Director presented progress reports on various phases of the 1948 Annual Meeting in Philadelphia, and reported that Rules and Regulations for the Foundry Show had been thoroughly discussed and approved at a meeting of the 1948 Exhibits Committee held October 10, thus assuring just cooperation on the part of Exhibitors. He stated that floor plans had been completed for mailing earlier in November, and that Exhibit details, space assignments, etc., would be handled by A. A. Hilbron, AFA Convention and Exhibits Manager.

Members of the 1948 Committee on Exhibits are as follows:

AFA Vice-President W. B. Wallis, *Chairman*

*For Equipment Exhibitors*

S. H. Hammond, Whiting Corp., Harvey, Illinois  
Thomas Kaveny, Jr., Herman Pneumatic Machine Co., Pittsburgh, Pa.  
W. L. Hartley, Link Belt Co., Philadelphia, Pa.  
C. V. Nass, Beardsley & Piper Co., Chicago, Illinois  
P. J. Potter, Pangborn Corp., Hagerstown, Maryland  
C. A. Barnett, Foundry Equipment Co., Cleveland, Ohio

*For Supplies Exhibitors*

F. R. Fleig, Smith Facing & Supply Co., Cleveland, Ohio  
E. H. King, Hill & Griffith Co., Cincinnati, Ohio  
N. J. Dunbeck, Eastern Clay Products, Inc., Jackson, Ohio  
J. A. Gitzen, Delta Oil Products Co., Milwaukee, Wisconsin  
L. H. Heyl, Federal Foundry Supply Co., Cleveland, Ohio  
E. T. Kindt, Kindt-Collins Co., Cleveland, Ohio  
T. G. Johnston, Republic Steel Corp., Cleveland, Ohio  
W. W. Maloney, AFA Secretary-Treasurer  
A. A. Hilbron, AFA Convention and Exhibits Manager

The Secretary pointed out that the Exhibits Committee had voluntarily recommended to the AFA Board that a strong letter be sent to the President of the Foundry Equipment Manufacturers Association, and the Foundry Supplies Manufacturers Association, urging the full cooperation of their members in abating excessive and unwarranted entertainment at the 1948 Convention. Without vote, it was agreed that President Kuniansky would so notify the two Exhibitor Trade Associations.

Without vote, the Executive Committee approved the continuation of the practice of offering to the Trade Associations of the Foundry Industry booth space gratis at the 1948 Convention and Exhibit, as an indication of AFA cooperation with these groups.

*New Chapter Possibilities*

The Secretary reported that meetings of a proposed new Chapter in the vicinity of Albany, New York had been held September 8 and October 14, attended by Vice-President Wallis, Directors Robb and Deane and Technical Director Massari. As a result, a petition for approval of a new Eastern New York Chapter was presented, signed by 50 members and prospective members. On recommendation of Messrs. Wallis, Robb and Massari, motion was made by Director Robb, and seconded by Vice-President Wallis, recommending to the Board of Directors

THAT the petition for an Eastern New York Chapter, as the 39th Chapter of AFA be approved.

Motion was unanimously carried, with instruction that the Secretary obtain letter ballot approval of the Board of Directors.

The Secretary reported that, following correspondence with the New England Foundrymen's Association, and a meeting of the NEFA Executive Committee, attended by the Secretary on September 10, he has since been informed that the matter has been indefinitely postponed. He indicated that there apparently exists, as in the past, definite objection on the part of some New England foundrymen to changing from the present status to Chapter status.

The Secretary reported briefly on a request from foundrymen in Provo, Utah for consideration of a Chapter, and on the basis of a minimum membership of 25, but stated that the local foundrymen had been advised that AFA could accept no petition signed by less than 50 members and prospective members.

*Report of Special Retirement Committee*

The Secretary reported that the Special Retirement Committee appointed by the Board at the July Board meeting, with Director Simpson as Chairman, had prepared and executed contracts with Messrs. C. E. Hoyt and R. E. Kennedy for retirement compensations over a period of seven years, as authorized. Contracts having been signed by all parties, including the AFA President and Secretary, the Executive Committee instructed the Secretary to express to the Retirement Committee its thanks for a job well done, and discharge the committee from further action.

*Staff Retirement Plan*

The Executive Committee considered the Staff Retirement Plan proposed by Marsh & McLennan, Insurance Actuaries, and presented at the July 30 Board Meetings, at which time specific action was deferred to the Executive Committee with the request that all Board Members study the proposed plan and present comments. The Executive Committee recommended that the plan and the Committee's recommendations for revision thereof be brought before the next meeting of the Board of Directors, scheduled for January 22, 1948, and the Secretary was instructed to forward the plan and recommended amendments to each Director for study in advance of the January meeting.

*American Standards Association Request*

The Secretary read a lengthy letter from the American Standards Association, dated September 29, 1947, dealing with Safety Code for Foundries, and requesting action by AFA toward revision of existing safety codes. This letter reviewed the history of AFA code making in conjunction with ASA, and pointed out that labor has recently indicated its intention of drafting its own foundry safety code, an action which thus far has not been encouraged by the U. S. Department of Labor. President Kuniansky then called for discussion by the Executive Committee which developed the consensus that AFA is not financially able at this time to embark on a full-fledged safety and hygiene program involving an estimated expenditure of \$25,000.00 or more per year. Instead the committee expressed the opinion that Safety and Hygiene work is a subject which the National Castings Council should be in a peculiarly excellent position to undertake and finance authoritatively.

Accordingly, on motion by Director Smith, seconded by Director Wasson, and carried, the Secretary was instructed to reply to the American Standards Association, in effect, that "the Executive Committee has approved the By-Laws of the National Castings Council, the most logical body to carry on a strong safety and hygiene program as suggested, and that AFA will seek to obtain action by the Council at the earliest possible date."

*Foreign Requests for AFA Service*

The Secretary requested clarification on AFA policy with respect to acceptance of requests from foreign countries for literature and memberships, where the degree of existing cooperation is entirely in one direction. Following brief discussion, motion was made, seconded and carried, approving continuation of the present policy of extending cooperation to foundrymen in foreign countries, both with respect to memberships and technical literature, where similar cooperation can be expected or is offered in return.



*Chapter Advertising Solicitation*

The question of solicitation of advertising by Chapters was discussed, with respect to complaints that such advertising requests may have increased to the point of becoming burdensome. Following discussion, motion was made, seconded and carried, that AFA should take no official stand on the question of solicitation of advertising by Chapters, that such advertising should be considered by prospects solely on the merits of the advertising itself, and that AFA should not attempt to "police" this matter.

*Cooperation With Munitions Board*

President Kuniansky and Technical Director Massari reported on a meeting held September 28, in Washington, D.C. with the Munitions Board, a meeting attended by a number of representatives of AFA and Foundry Trade Associations. The President expressed satisfaction with the outcome of the meeting, indicating that considerably more thought is being given in confidential circles to foundry industry representation on any future organizational set-up, as in the event of another mobilization of industry.

*1948 Nominating Committee*

The Executive Committee appointed July 29, and recommended to the 1947-48 Board of Directors for approval, the following 1948 Nominating Committee:

Past-President, S. V. Wood, Minneapolis, *Chairman*.  
 Past President F. J. Walls, Detroit  
 L. E. Roby, Peoria Malleable Casting Co., Peoria, Ill., rep. Central Illinois Chapter—Malleable Iron.  
 L. D. Wright, U. S. Radiator Corp., Geneva, N.Y., rep. Central New York Chapter—Gray Iron.  
 Wm. M. Ball, Jr., Edna Brass Div. of Magnus Metal Co., Cincinnati, Ohio, rep. Cincinnati Chapter—Brass & Bronze.  
 R. F. Lincoln, Russell F. Lincoln Co., Cleveland, Ohio, rep. Northeastern Ohio Chapter—Equipment & Supplies.  
 Earl M. Strick, Erie Malleable Iron Co., Erie, Pa., rep. North-western Pennsylvania Chapter—Malleable Iron.  
 Henry B. Hanley, American Laundry Machinery Co., Rochester, N.Y., rep. Rochester Chapter—Gray Iron.  
 Robert R. Haley, Advance Aluminum & Brass Co., Los Angeles, rep. Southern California Chapter—Aluminum & Magnesium.  
 There being no further business to come before the Executive Committee, the meeting was declared adjourned.

Respectfully submitted,

WM. W. MALONEY,  
*Secretary-Treasurer*

APPROVED:

MAX KUNIAISKY, *President*

## Minutes Meeting of Executive Committee LaSalle Hotel, Chicago—January 21, 1948

**Present:** President Max Kuniansky, presiding  
 Vice-President W. B. Wallis  
 Director James H. Smith  
 Director John M. Robb, Jr.  
 Secretary-Treasurer Wm. W. Maloney  
 Technical Director S. C. Massari

**Absent:** Director S. C. Wasson  
 Director S. V. Wood

*Reading and Approval of Minutes*

Minutes of the Executive Committee Meeting, held in Chicago, November 3, were read, and on motion by Director Robb, seconded by Director Smith and carried, were approved.

*AFA Office Rental Lease*

Informal discussion developed on renewal of the AFA office lease on May 1, 1948, and the consensus of the Committee was that any lease signed should be for a maximum period of three years in view of present conditions.

(Note: All actions of this Executive Committee meeting, being subject to approval by the Board of Directors, are included in the Minutes of the Board of Directors meeting held Jan. 22, 1948).

Respectfully submitted,

WM. W. MALONEY,  
*Secretary-Treasurer*

APPROVED:

MAX KUNIAISKY, *President*

## Minutes Adjourned Meeting of Executive Committee LaSalle Hotel, Chicago—January 22, 1948

**Present:** President Max Kuniansky, presiding  
 Vice-President W. B. Wallis  
 Director John H. Robb, Jr.  
 Director J. H. Smith  
 Secretary-Treasurer Wm. W. Maloney  
 Technical Director S. C. Massari

**Absent:** Director S. C. Wasson  
 Director S. V. Wood

The meeting was called to order immediately following adjournment of the Board of Directors Meeting, the President stating the purpose of the meeting being to follow through on placement of the Staff Retirement Plan for possible adoption on July 1, 1948.

Following discussion, on motion by Director Smith, seconded by Vice-President Wallis, and unanimously carried, the Executive Committee voted to have all details of the Staff Retirement Plan developed by Marsh & McLennan, Insurance Actuaries, Chicago, and the Secretary was instructed to communicate immediately with that firm for initiation of the Retirement Plan on July 1, 1948.

There being no further business to be considered, the meeting was declared adjourned.

Respectfully submitted,

WM. W. MALONEY,  
*Secretary-Treasurer*

APPROVED:

MAX KUNIAISKY, *President*

## Minutes Meeting of Board of Directors LaSalle Hotel, Chicago—January 22, 1948

**Present:** President Max Kuniansky, presiding  
Vice-President W. B. Wallis

**Directors:** R. Gregg H. G. Lamker  
E. W. Horlebein B. L. Simpson  
H. H. Judson E. H. Delahunt  
J. H. Smith W. J. MacNeill  
F. M. Wittlinger R. H. McCarroll  
H. A. Deane J. M. Robb, Jr.  
J. E. Kolb  
Secretary-Treasurer Wm. W. Maloney  
Technical Director S. C. Massari

**Absent:** Director S. C. Wasson  
Director S. V. Wood  
Director A. C. Ziebell

### *Reading and Approval of Minutes*

Minutes of the First Meeting of the Board of Directors on July 30, 1947, and of the Executive Committee on November 3, 1947 having been forwarded to all Directors previously, motion was duly made, seconded, and carried, approving these Minutes without reading.

Minutes of the Executive Committee Meeting January 21, 1948 were read in digest, with the understanding that they would be forwarded to the Board of Directors for letter ballot approval.

### *Membership Report*

Membership Report was presented showing total membership on December 31, 1947 of 10,172, a net increase of 489 over June 30, 1947. The report showed AMERICAN FOUNDRYMAN circulation of 10,172 members, plus 108 subscriptions, total of 10,280. It was pointed out that never before had the Association membership exceeded 10,000.

Breakdown of the membership showed 9266, or 91 per cent of total membership affiliated with AFA Chapters; 458 non-Chapter members, and 448 Foreign members.

### *Financial Report*

The Treasurer presented a Financial Report for the 6 months ended December 31, 1947, showing Income totalling 49.0% of the year's budget, and Expenses totalling 38.4% of the year's budget. Specific items of the Income and Expense with reference to the Budget were commented on and detailed statements presented to the Directors.

On motion by Director Deane, seconded by Director Delahunt, and carried, the Financial Report was accepted by the Board.

It was pointed out that funds available for general purposes of the Association, including the General Reserve Fund and the TDP Fund, totaled \$224,280.64. Discussion brought out a consensus that effort should be made to increase the funds available for general purposes, and a specific Staff recommendation for use of excessive TDP Funds should be prepared. It was also the consensus that caution should be used in setting up additional specifically earmarked funds within the Association, so as not to deplete those funds available for general purposes and current operating uses.

### *New AFA Chapters*

The Secretary reported briefly on approval and installation of the Eastern New York Chapter at Albany, New York and an Ohio State University Student Chapter at Columbus, Ohio; also Board approval of Student Chapters at University of Minnesota, Minneapolis, and University of Missouri School of Mines and Metallurgy, at Rolla, Missouri.

The Secretary then presented a petition for formation of a fourth Student Chapter at Massachusetts Institute of Technology, Boston, and, in view of the petition meeting the requirements of the AFA Student Chapter Policy, recommended its approval. On motion by Director Horlebein, seconded by Director Robb, and carried, the Board voted unanimously to approve this fourth Student Chapter.

### *Recommendations of Board of Awards*

In accordance with the By-Laws, the recommendations of the Board of Awards were submitted for Board approval:

*1948 Medalists and Honorary Life Memberships.* On motion by Director Robb, seconded by Director Wittlinger, and unanimously carried, the Board of Directors approved recommendations of the Board of Awards for the awarding of three Gold Medals and four Honorary Life Memberships in 1948 as follows:

The William H. McFadden Gold Medal to E. H. Ballard, for his steadfast and sincere efforts on behalf of the Association while serving as President during most difficult times, and his continual contributions to the foundry industry.

The Peter L. Simpson Gold Medal to Peter E. Rentschler, for his outstanding and inspiring work toward the promotion of better safety and housekeeping practices in the foundry industry.

The John A. Penton Gold Medal to R. G. McElwee, for his earnest work and outstanding contributions to the dissemination of foundry knowledge, and especially his efforts on behalf of the Cupola Research Project of AFA. Honorary Life Memberships to Medalists Ballard, McElwee and Rentschler; and to Max Kuniansky on completion of his term as AFA President.

*Apprentice Awards.* Board of Awards recommendation was considered that funds for prizes in the 1948 Apprentice Contest be drawn from General Funds of the Association. Discussion developed on availability of the S. Obermayer Fund for such purposes, and motion was made by Director Simpson, seconded by Director Kolb, and carried, that 1948 Apprentice Contest Funds be drawn from the Association's General Funds, unless moneys in the S. Obermayer Fund might be made available for such purposes following discussion between the donor and the Chairman of the Board of Awards. It was the consensus that no fund in the Association be permitted to lie inactive without investigation as to the possibilities for its use.

*Awards Manual.* Action by the Board of Awards was presented, and approved by the Board of Directors without vote, instructing the Staff to prepare "a Manual for the guidance of committees, Chapters, members, and other qualified persons in nominating to the Board of Awards potential candidates for award consideration."

### *Acceptance of Auditors' Report, June 30, 1947*

Recommendation of the Executive Committee that the Auditors' Report for the fiscal year ended June 30, 1947 be approved by the Board, approval having been deferred at the July Board meeting pending analysis of Association finances by an Auditing Committee, on motion by Director Simpson, seconded by Director McCarroll and carried, recommendation accepted.

### *Employment of Staff Members*

Recommendation of the Executive Committee that the employment of John D. Murray as Advertising Manager of AMERICAN FOUNDRYMAN, A. A. Hilbron as Convention and Exhibits Manager, and Miss E. E. Chesire as Librarian be approved by the Board. On motion by Director Delahunt, seconded by Director Judson, recommendation accepted.

### *Agreement With Despatch Oven Co.*

Recommendation of Executive Committee that agreement reached between the Staff and Despatch Oven Co., for settlement of payment for laboratory oven provided for AFA Sand Research at Cornell University, be approved. On motion by Director Horlebein, seconded by Director Simpson, and carried, recommendation was accepted.

*"American Foundryman" Advertising Rates*

Recommendation of the Executive Committee that advertising rates in AMERICAN FOUNDRYMAN be increased, effective as of January 1, 1948, be approved. On motion by Director Horlebein, seconded by Director McCarroll, and carried, recommendation accepted. The basic single time, black and white, advertising rate in AMERICAN FOUNDRYMAN shall be \$225.00 per page, effective January 1, 1948.

*Steel Division Research Project*

Recommendation of Executive Committee, that maximum expenditure of \$5000 for Steel Division Research Project "A Study of the Influence of Mold Conditions on the Development of Hot Tears in Steel Castings" during 1947-48 be approved. On motion by Director Delahunt, seconded by Director Horlebein, and carried, authorization of funds approved.

*Heat Transfer Project*

Recommendation of Executive Committee, for authorization of \$2000.00 for Heat Transfer Project during 1947-48 be approved. It was pointed out that through oversight these funds had not previously been authorized but would not represent an additional item for the current year's budget. On motion by Director Horlebein, seconded by Director Delahunt, and carried, authorization of funds approved.

*Chicago Museum of Science and Industry*

Recommendation of Executive Committee, for authorization of \$2000 for maintenance of foundry exhibit at Chicago Museum of Science and Industry during 1947-48, be approved. It was pointed out that to this maintenance fund the Chicago Chapter has contributed \$300 in addition to expense of \$3200 by the Museum, and that attendance at the exhibit is far greater than local. On motion by Director Simpson, seconded by Director Judson, and carried, authorization of funds approved. Director Simpson brought out the necessity for rewriting the foundry exhibit's script at the earliest opportunity and, without vote, it was the consensus that the proper educational committee of AFA be asked to develop a new script promptly.

*Staff Retirement Plan*

Recommendation of the Executive Committee was presented for approval of the Staff Retirement Plan as originally presented at the Board meeting in July, 1947, commented on thereafter by all members of the Board, and recommended to the Board of Directors by the Executive Committee on November 3, 1947, with certain revisions. It was pointed out that these revisions had been forwarded to all members of the Board for consideration prior to the current Board meeting. Therefore, motion was made by Director McCarroll and seconded by Director Lamker

THAT the Staff Retirement Plan as recommended by the Executive Committee on November 3, 1947, with revisions, be approved.

Motion was carried by unanimous vote of the Board of Directors.

*Selection of 1949 Convention City*

Recommendation of the Executive Committee that the 53rd Annual Meeting of the Association be held at St. Louis as a Non-Exhibit Convention, on the dates of May 2-5, 1949, be approved. On motion by Director Horlebein, seconded by Director Simpson, and carried, recommendation was accepted.

Director Deane called attention to the desirability of staging a Saturday Open House for local plant men at the next Exhibit Convention, so as to provide opportunity for a greater number to inspect the Exhibits. Without vote, the matter was referred to the 1950 Committee on Exhibits (to be appointed at the proper time) for consideration.

*AFA Membership in ASA*

Recommendation of the Executive Committee that AFA Membership in American Standards Association be renewed for one year with dues of \$500, be approved. On motion by Director Horlebein, seconded by Director Judson, and carried, authorization of expenditure approved.

*1952 International Foundry Congress*

Recommendation of Executive Committee that the AFA be host to an International Foundry Congress in 1952 be approved. It was pointed out that a schedule of International Foundry Congresses, calling for a meeting in the United States in 1952 had been adopted by the International Committee of Foundry Technical Associations at a meeting at Paris in September, 1947, with the AFA European representative Vincent Delpont of London attending.

On motion by Director Simpson, seconded by Director Delahunt, and unanimously carried, recommendation of the Executive Committee was approved.

*Author of AFA College Textbook*

Recommendation of Executive Committee that Prof. P. E. Kyle of Cornell University be retained as the author of the projected AFA college textbook be approved. On motion by Director Horlebein, seconded by Director Judson, and carried, recommendation accepted.

It was stated that expenditure for this purpose would be made following completion of the book, and therefore would be included in the budget for the 1948-49 fiscal year.

*AFA Representative to 1948 International Foundry Congress*

Recommendation of Executive Committee that L. B. Knight of Chicago be asked to serve as the official AFA representative at the International Foundry Congress in Czechoslovakia in October, 1948 be approved. On motion by Director Robb, seconded by Director Wittlinger, and carried, recommendation was accepted.

*AFA By-Laws Revision*

The recommendations of the Executive Committee for approval of a general revision of the Association By-Laws was presented and discussed article by article, and the following additional revisions made by the Board:

Art. I (Name and Object) Sec. 1. Proposed change in name to "American Society of Foundrymen" changed to "American Foundrymen's Society"; on motion by Director Judson, seconded by Director Horlebein, and carried.

Art. II (Membership) Sec. 1 (b). To definition of Company Membership a third sentence added reading as follows: "No Company or Sustaining Membership will qualify individuals for Personal Membership at minimum annual dues, where such individuals reside outside the territory of the Chapter in which such membership is held." Change adopted on motion by Director MacNeill, seconded by Director Simpson, and carried.

Art. XVII (Chapters) Sec. 6. To this section regulating the distribution of technical publications "over the name of the Society," an additional sentence added, reading as follows: "Presentation of papers at a Chapter Meeting or Regional Conference, or distribution of same to the Chapters or membership, shall not be construed as technical publication over the name of the Society." This change adopted on motion by Director Simpson, seconded by Director MacNeill, and carried. Following the above revisions, motion was made by Director Simpson, and seconded by Director MacNeill

THAT the proposed general revision of Association By-Laws recommended by the By-Laws Committee on December 19, 1947 and the Executive Committee on January 21, 1948 be accepted as further revised by the Board of Directors, and promulgated in accordance with the Association By-Laws for letter ballot by the membership, for possible adoption as of July 1, 1948.

In consideration of the work by the By-Laws Committee under chairmanship of Past-President H. Bornstein, the following resolution was adopted unanimously by the Board of Directors on motion by Director Horlebein, seconded by Director MacNeill, and carried:

RESOLVED that the Board of Directors of the Association extends to the By-Laws Committee, through its Chairman, H. Bornstein, the congratulations of the Board for its conscientious and well-considered action in developing a general and up-to-date revision of the Association's By-Laws.

By unanimous action of the Board of Directors in meeting January 22, 1948.

APPROVED:

MAX KUNIAISKY, President



*National Castings Council*

Director Smith reported on his attendance at a meeting of the National Castings Council, August 4, at which time proposed By-Laws for the Council were presented. He pointed out that it was necessary for each individual representative body on the Council to approve the proposed By-Laws, copies of which were sent all AFA Directors on September 5. Following brief discussion, motion was made by Director Wood, seconded by Vice-President Wallis, recommending to the Board of Directors

THAT AFA approve the proposed By-Laws for the National Castings Council as adopted August 4, 1947.

Motion was carried unanimously.

Recommendation of Executive Committee that the sum of \$100 be appropriated as the Association's part in establishing a petty cash fund for the National Castings Council, be approved. President Kuniansky, in reporting on meeting of the National Castings Council January 15, 1948 announced the election of F. G. Steinhach of Cleveland as Secretary of the Council, and formal organization of the Council as a going body. On motion by Director Simpson, seconded by Director MacNeill, and carried, authorization for the expenditure of \$100 for this purpose was approved. President Kuniansky also reported that the National Castings

Council was favorably considering the recommendation of AFA representatives that a strong Safety and Hygiene Program be undertaken, and stated that the matter would be further considered at a meeting of the Council during the Philadelphia Convention.

*1948 Convention and Exhibit*

The Secretary and Technical Director reported on progress of the technical and general program and the foundry exhibit for the Philadelphia Convention, May 3-7.

The Secretary stated that the Chairman of the General Host Chapter Committee, Director Robb had developed a program of plant visitations and ladies' entertainment as well as an operating exhibit by the Murrell Dobbins Vocational-Technical School of Philadelphia, and that these events would add materially to the interest of the Convention.

There being no further business, the meeting was declared adjourned.

Respectfully submitted,

WM. W. MALONEY,  
Secretary-Treasurer

APPROVED:  
MAX KUNIAISKY, President

## Minutes Executive Committee Meeting LaSalle Hotel, Chicago—March 16, 1948

Present: President Max Kuniansky, presiding  
Vice-President W. B. Wallis

Directors: Director S. V. Wood  
Director S. C. Wasson  
Director J. M. Robb, Jr.  
Secretary-Treasurer Wm. W. Maloney  
Technical Director S. C. Massari

Absent: Director J. H. Smith

*Reading and Approval of Minutes*

The reading of Minutes of the Executive Committee Meeting held January 21, 1948 was dispensed with since those minutes were approved at the Board Meeting of January 22.

The reading of the Minutes of the Board of Directors Meeting held January 22, 1948 was dispensed with since those minutes had previously been approved by letter ballot.

*Membership Report*

The Secretary reported that membership as of February 29, 1948 stood at 9904 as against 9683 on June 30, 1947. He pointed out that from the high of 10,246 on January 31, the removal of 480 delinquents during February had reduced the total to 9904, a net increase of 221 for the fiscal year to date, but expressed confidence that the total would be above 10,000 by March 31.

The Secretary reported Total Dues Income of \$165,420.65, exclusive of \$6,591.43 Advance Dues for 1948-49, as of February 29, 1948. This figure compares with the Budget of \$175,000.00 for the current fiscal year.

*Financial Report*

The Secretary-Treasurer presented a Financial Report as of February 29, 1948 showing Actual Income of \$367,523.59, or 77.2 per cent of the Budget \$476,250.00; Actual Expense, \$265,544.06, or 55.8 per cent of the Budget.

Balance Sheet as of February 29 showed an increase in the Fund Principal (Surplus) for the General Fund of \$115,656.72, compared with \$15,594.00 on June 30, 1947, the difference being accounted for in the main by improvement in cash position due to Exhibit revenues.

The Executive Committee examined major items of Income and Expense, with the view to estimates for the balance of the fiscal year, indications being given that the Association in all probability would exceed the Income Budget and remain within the Expense Budget for the fiscal year.

*Staff Retirement Plan*

In advancement of the previously approved Staff Retirement Plan, a letter dated February 13, 1948 from Marsh & McLennan was read, setting up the procedure of establishment. The following actions resulted:

- (a) On motion by Director Wood, seconded by Director Wasson and carried, a Retirement Committee of three members was set up, consisting of one Board Member and two non-Board members for purposes of continuity, personnel as follows:

B. L. Simpson, Chairman  
R. L. Lee, Grede Foundries, Inc., Milwaukee  
F. W. Shipley, Caterpillar Tractor Co., Peoria, Ill.

- (b) Without motion, it was agreed that rates of interest on vested funds be fixed, and details of the Retirement Plan documents be established, by the Retirement Committee. It was pointed out that the final plan would be subject to approval by the Treasury Department in Washington.

*Educational Activities*

The Association's Educational activities were discussed at length, hinging on the Secretary's announcement that H. F. Scobie, formerly Educational Director of AFA, had been appointed Acting Editor of AMERICAN FOUNDRYMAN. President Kuniansky reported on conversations held March 11 in Cleveland with G. K. Dreher, Executive Director of the Foundry Educational Foundation, relative to the respective educational fields to be covered by AFA and the Foundation. The consensus of the Committee was that steps should be taken to replace Mr. Scobie as head of the Association's educational work, under the Technical Director.

It was agreed that a plan of Educational Activities should be drawn up by the Secretary and submitted for discussion at the Board of Directors' meeting to be held May 4 in Philadelphia. It was also agreed that President Kuniansky during Convention week would further advance the Educational Program, in conference with F. G. Seifing, Chairman of the AFA Educational Division, in view of the fact that Mr. Seifing had found it impossible to accept an invitation to attend the current Executive Committee Meeting. The consensus was that the AFA Program in the future should be more concentrated on a limited number of definite objectives.

There being no further business to come before the Executive Committee, the meeting was declared adjourned.

Respectfully submitted,

WM. W. MALONEY,  
Secretary-Treasurer

APPROVED:  
MAX KUNIAISKY, President

## Minutes Board of Directors Meeting

Benjamin Franklin Hotel, Philadelphia—May 4, 1948

**Present:** President Max Kuniansky, presiding  
Vice-President W. B. Wallis

**Directors:** R. Gregg J. E. Kolb  
E. W. Horlebein H. G. Lamker  
H. H. Judson B. L. Simpson  
J. H. Smith S. C. Wasson  
F. M. Wittlinger E. N. Delahunt  
S. V. Wood W. J. MacNeill  
H. A. Deane J. M. Robb, Jr.  
Secretary-Treasurer Wm. W. Maloney  
Technical Director S. C. Massari

**Absent:** Director A. C. Ziebell

### *Reading and Approval of Minutes*

Minutes of the Board of Directors Meeting held January 22, 1948 were dispensed with, it being announced that these minutes had been approved by letter ballot of the Directors.

Minutes of the Executive Committee Meeting held March 16, 1948 were read, and on motion, duly seconded, were approved.

### *Membership Report*

Membership report for the period July 1, 1947 to March 31, 1948 was presented, showing a total of 10,174 members, a net increase of 491 over the total of 9,683 at the beginning of the current fiscal year. The report showed a total circulation of AMERICAN FOUNDRYMAN (including 108 subscriptions) 10,282.

Receipts of member dues as of March 31 were given as \$165,107.57, excluding advance dues of \$12,735.49 applicable to 1948-49.

### *Financial Report*

The Financial Report covering the fiscal year to March 31, 1948 showed Actual Income of \$426,185.54, or 89.5 per cent of the year's budget of \$476,250.00; Actual Expense \$297,842.11, or 62.5 per cent of the budget. Unbudgeted Expense items not included total \$2,587.09, bringing total Expense incurred for the current Fiscal year to \$300,429.20.

Balance Sheet presented as of March 31 showed Fund Principal in the General Funds of \$141,350.34, compared with \$115,656.00 on February 29 and \$15,594.00 on June 30, 1947.

### *Report of Technical Director*

The Technical Director presented a report covering Technical Activities of the Association during 1947-48, emphasizing the initiation of research projects for the various divisions. At present eight fundamental Research projects are being sponsored by AFA, of which five are projects for the the following divisions: Aluminum and Magnesium (2), Brass and Bronze, Malleable, and Steel. Other Research Projects include the Cupola Research Program, work on Heat Transfer and Sand Research.

The Technical Director called attention to publication of the new book *Analysis of Castings Defects*, and to the work of the Committees in arranging the Technical Program for the Philadelphia Convention.

### *Safety and Hygiene*

President Kuniansky reported on a meeting of the National Castings Council, held the morning of May 4, at which a Safety and Hygiene Program was discussed at length. He stated that the Council expressed itself favorable to undertaking such a program, with the intention of reviving such a Committee within each membership body, each body to appoint one representative on a committee to bring present Safety and Hygiene Codes for the Foundry Industry up-to-date, it being agreed that funds for such a program should be raised by the Council.

Following discussion, motion was made by Director Smith, seconded by Director Deane and unanimously carried, appointing Past-President L. C. Wilson as the AFA representative on the Council's Safety and Hygiene Committee, with the understanding that Mr. Wilson's expenses incurred in this activity would be reimbursed.

### *Death of Director R. H. McCarroll*

President Kuniansky expressed the sorrow of the Association in the recent death of Director R. H. McCarroll, of the Ford Motor Co., and stated that expressions of sympathy have been extended to Mrs. McCarroll for the Board of Directors. The following resolution was read, and on motion duly approved and seconded, was ordered spread upon the Minutes of the Association:

WHEREAS the untimely death of R. H. McCarroll brings to an end a term of great promise as a member of the Board of Directors of American Foundrymen's Association, and

WHEREAS Hudson McCarroll gained for himself a host of lasting friendships in the Castings Industry by his conscientious and earnest efforts toward the development of foundry practice, and

WHEREAS he earned the permanent respect of engineering and technical societies everywhere by unflagging interest in their activities, and his steadfast belief in the principal of cooperative dissemination of technical information, and

WHEREAS Hudson McCarroll endeared himself to all who knew him for his personal attributes, now, therefore,

BE IT RESOLVED that the Board of Directors of the American Foundrymen's Association expresses by this resolution its deep sorrow in the loss of its distinguished associate and friend, and extends to the family of R. H. McCarroll and his business associates, this expression of sincere sympathy. This Resolution unanimously approved by action of the Board of Directors in meeting May 4, 1948 at Philadelphia.

MAX KUNIAISKY, President

### *Election of New Director*

President Kuniansky called attention to Art. VI, Sec. 2, of the By-Laws which provides that in the event of a vacancy on the Board of Directors, the Board shall appoint a member of the Association for the unexpired term. On motion by Director Smith, seconded by Vice-President Wallis, and unanimously carried, Frank L. Riecks, Manager, Ford Motor Co., Detroit, was elected to fill the vacancy left by the death of Director McCarroll.

### *1948 Convention and Exhibit*

The Secretary reported on the Convention and Exhibit currently in session and called attention to the splendid displays arranged by 270 exhibiting companies, and to the fine program of technical sessions arranged by the Technical Director in cooperation with the Committees of the Association. The President read a telegram received from the President of the Institute of British Foundrymen congratulating the Association on its 52nd Annual Meeting and Exhibit, and expressed the appreciation of AFA for this token of friendship and remembrance.

President Kuniansky called attention to the special services rendered by former Convention and Exhibits Manager, C. E. Hoyt, in connection with the 1948 Exhibit, and suggested that the Association express its appreciation for such services. On motion by Director Simpson, seconded by Director Wittlinger, and carried, a consideration was voted unanimously for C. E. Hoyt as a token of appreciation for his helpfulness.

### *High School Textbook*

Recommendation of the Executive Committee was presented, requesting authority for retaining Dewey F. Barich, Head of Industrial Education at Kent State University, Kent, Ohio, as authority of AFA high school textbook, in accordance with recommendation of the Educational Division. On motion by Director Simpson, seconded by Director Judson and carried, the recommendation was accepted.

### *Staff Personnel*

Recommendations of the Finance Committee for employment of Phillip D. Johnson as Educational Assistant for AFA, and of Robert N. Sheets as Assistant Editor for AMERICAN FOUNDRYMAN were presented. On motion by Director Smith, seconded by Director Wood and carried, the appointments were confirmed.

*New Chapters*

The Secretary announced installation of Student Chapters at Massachusetts Institute of Technology, Missouri School of Mines and Metallurgy, and University of Minnesota. He also stated that petition had been received for the formation of a student Chapter at Oregon State College, and that this petition had been approved by the Board of Directors on letter ballot.

The Secretary also announced organization and Board approval of a new Tennessee Chapter at Chattanooga.

*Research Projects*

The Technical Director requested an appropriation of \$6,500 during the year 1948-49 for continuation of the Research Project for the Aluminum and Magnesium Division now being conducted at Battelle Memorial Institute, Columbus, Ohio. He stated that the customary \$5,000 annual appropriation, as a result of the past year's experience, was considered somewhat inadequate for completing this work. Following discussion, approval of the requested funds was deferred for consideration by the Finance Committee in connection with the Budget for the coming year.

*Publications Distribution*

The Secretary requested discussion on the issuance of free publications of AFA for various purposes, such as local apprentice contests, local educational course attendance, and school and college libraries. He stated that a number of requests for free publications have been received, and suggested the establishment of a firm policy. On motion by Director Simpson, seconded by Director Kolb and carried, the Board agreed that the distribution

of free publications for miscellaneous use be discontinued indefinitely.

*AFA Employment Service*

A proposal to institute a more formalized policy of employment service for AFA members was discussed at length, including a statement of policy prepared by the Secretary. It was the consensus that recommendations of any individual by the Staff should be definitely avoided, as in the past. On motion by Director Horlebein, seconded by Director Judson and carried, the Employment Service Plan, as presented, was approved.

*By-Laws Amendments*

The Secretary reported that the proposed amendments to the AFA By-Laws, approved for letter ballot at the Board Meeting on January 22, 1948, went into the mails on April 15, and that ballots received would be counted on May 15 as provided for in the By-Laws. The ballot tellers appointed by the President were announced as follows: James Thomson, Continental Foundry & Machine Co., East Chicago, Ind., Chairman; A. S. Klopff, Western Foundry Co., Chicago; R. W. Schroeder, Foundry Instructor, University of Illinois, Chicago.

There being no further business to be considered, the meeting was declared adjourned.

Respectfully submitted,

WM. W. MALONEY,  
Secretary-Treasurer

## APPROVED:

MAX KUNIAISKY, President  
June 8, 1948

## Minutes Executive Committee Meeting Hot Springs, Virginia—June 8, 1948

Present: President Max Kuniansky, presiding  
Vice-President W. B. Wallis  
Director James H. Smith  
Director John M. Robb, Jr.  
Secretary-Treasurer Wm. W. Maloney

Absent: Director S. C. Wasson  
Director S. V. Wood

*Reading of Minutes*

Reading of the Minutes of the May 4, 1948 meeting of the Board of Directors at Philadelphia was dispensed with, the Minutes being forwarded to all Members of the Board for letter ballot approval.

*Membership Report*

The Secretary presented a Membership Report as of May 31, 1948 showing total of 10,318 members in good standing as of that date, representing a net increase of 635 over the total of 9683 on June 30, 1947. Total circulation of AMERICAN FOUNDRYMAN was given as 10,444, including 10,318 members and 126 subscriptions as of May 31.

Membership dues collections during the fiscal year to May 31, 1948 were given as \$191,124.73, including \$25,098.60 advance dues applicable to 1948-49, a net of \$166,026.13 applicable to the current fiscal year.

*Report on Finances*

The Secretary-Treasurer presented a Financial Report as of April 30, 1948 showing Total Income of \$478,818.70, or 100.5 per cent of the Budget \$476,250.00 for the fiscal year; also Total Expense of \$359,246.28 or 75.4 per cent of the year's Budget. It was pointed out that the Convention Expenses were not included in this report as of April 30, but for the most part would be included in the report as of May 31. Balance Sheet submitted

showed the Fund Principal of the General Fund as \$131,689.76 on April 30, 1948, as against \$15,594.00 on June 30, 1947.

*Greetings to IBF Annual Meeting*

At the request of the President, the Secretary was instructed to send the following message to P. H. Wilson, the President of the Institute of British Foundrymen, as greetings from AFA to the 45th Annual General Meeting of IBF in London, June 9, 1948.

"Foundrymen of America extend cordial greetings and best wishes for a completely successful 45th general meeting of the Institute of British Foundrymen."

*Staff Retirement Plan*

The Secretary reported on progress of the Staff Retirement Plan as follows:

- (a) It was pointed out that the Retirement Committee, including B. L. Simpson, Chairman; R. L. Lee and F. W. Shipley, met on May 12, and made certain recommendations. Following discussion, on motion by Director Robb, seconded by Vice-President Wallis and carried, it was recommended to the Board of Directors:

THAT the Retirement Committee as appointed be disbanded, and that the following be appointed Trustees for the AFA Staff Retirement Plan:

B. L. Simpson  
R. L. Lee  
F. W. Shipley  
H. Bornstein

- (b) On motion by Director Robb, seconded by Vice-President Wallis and carried, the Executive Committee recommended to the Board of Directors

That all Trustees for the Staff Retirement Plan be indemnified by surety bonds in appropriate amounts, the cost of such bonds to be borne as an expense of the Retirement Trust.



- (c) On motion by Director Smith, seconded by Vice-President Wallis and carried, it was recommended to the Board of Directors

THAT all Retirement Funds be set aside by the Association in a special account at the Harris Trust and Savings Bank, Chicago, and that signatures of the Trustees be properly registered for purpose of withdrawals from such funds on behalf of the Retirement Plan.

- (d) The Executive Committee considered recommendations of the Retirement Committee that possible service credits for service prior to July 1, 1948, the effective date of the Retirement Plan, be allowed for Wm. W. Maloney, T. B. Koeller, C. R. McNeill and H. Hanson, service credits being as compiled by Marsh & McLennan as insurance consultants. On motion by Director Robb, seconded by Director Smith and carried, the proposed past service credits were recommended to the Board of Directors for acceptance and adoption.
- (e) Following discussion, it was agreed to accept the recommendation of the Retirement Committee that the Retirement Trust be administered by the Trustees appointed, without engaging a financial organization as Trustees to the plan, in view of the fact that investment of surplus funds is not involved since it is intended that Retirement Plan funds be used for the purchase of retirement annuity policies.
- (f) The Secretary presented the following resolution necessary for establishment of a Retirement Fund, the execution of all necessary documents to establishment of such fund and the appointment of Trustees to carry out the Retirement Trust agreements:

"IT IS HEREBY RESOLVED that there shall be, and there is hereby established a Retirement Fund for salaried employees of the Society, which shall be designated and known as the AMERICAN FOUNDRY-MEN'S SOCIETY STAFF RETIREMENT PLAN," the plan for which is set forth in a form of Agreement attached to and made a part of these minutes as Exhibit 1.

"IT IS FURTHER HEREBY RESOLVED that the President and Secretary of the Society be, and they are hereby, duly authorized and directed to execute and enter into an Agreement in substantially the form of said Exhibit 1, and that they and other officers of the Corporation be, and they are hereby, authorized and directed to do any and all other acts necessary to the establishment of said Retirement Fund and the consummation of the Plan therefor, including the payment on behalf of the Society of such sums as it shall become necessary for the Society to pay to and deposit with the Trustees in accordance with the provision of said Plan.

"IT IS FURTHER HEREBY RESOLVED that B. L. Simpson, R. L. Lee, F. W. Shipley and H. Bornstein be, and they are hereby, designated and appointed as the Trustees provided for by said Plan; and said B. L. Simpson, R. L. Lee, F. W. Shipley and H. Bornstein

upon acceptance of said Trusteeship, are, and any successor-trustee upon becoming appointed and qualified as provided in the Plan is, hereby vested with all of the powers, privileges, estate, interest and direction and charged with all of the duties and obligations with which the Trustees are vested and charged as set forth in said plan."

On motion by Director Smith, seconded by Director Robb and carried, the resolution was, by unanimous action, recommended to the Board of Directors for adoption.

- (g) The Secretary presented for consideration and approval a complete Trust Agreement for the "American Foundrymen's Society Retirement Trust," including all details of the Staff Retirement Plan to become effective July 1, 1948, such Plan having been prepared in conjunction with Marsh & McLennan as insurance consultants to the Plan, and the AFA Attorney, W. H. Alexander of Chicago. Acceptance of the Trust Agreement and Plan, it was pointed out, were both incorporated in the above resolution adopted by the Executive Committee.
- (h) The Secretary presented for consideration copy of an informative booklet covering the Staff Retirement Plan, as intended for distribution to all Staff Employees. Without vote, the Committee accepted the form and manner of presentation and instructed the Secretary to proceed with its publication to the Staff Employees.

#### Membership Renewal Dates

The Secretary described a proposal to adjust all membership renewal dates so that all dues of all members in any one Chapter area might become renewable on the same date . . . e.g., November 1, January 1, May 1, etc. He stated that such a plan would (a) avoid the heavy peak load of renewals during the short staffed summer vacation months, (b) eliminate the need for temporary employees during summer months, (c) facilitate the work of Chapter Membership Committee Chairmen, and (d) enable more efficient handling of memberships throughout the year.

Following discussion, it was the consensus of the Committee that the plan as outlined should be tested for acceptance by the membership at the Chapter Chairman Conference June 28-30.

#### Foundry Short Courses

The Secretary proposed that AFS initiate in 1949 a summer short course in foundry practice at one of the national engineering universities, as a part of the AFS educational program. Without vote, it was the consensus of the Committee that the plan should be further developed before Board approval could be obtained, and suggested discussion with the Technical Correlations Committee at the latter Committee's Annual Meeting in June.

Respectfully submitted,

WM. W. MALONEY,  
Secretary-Treasurer

APPROVED:

MAX KUNIAISKY, President  
July 16, 1948

## Minutes Annual Meeting 1947-48 Board of Directors

Palmer House, Chicago—July 29, 1948

**Present:** President Max Kuniansky, presiding  
Vice-President W. B. Wallis  
**Directors:** Robert Gregg B. L. Simpson  
E. W. Horlebein S. C. Wasson  
H. H. Judson E. H. Delahunt  
J. H. Smith W. J. MacNeill  
F. M. Wittlinger F. C. Riecks  
S. V. Wood J. M. Robb, Jr.  
J. E. Kolb  
Secretary-Treasurer Wm. W. Maloney  
Technical Director S. C. Massari

**Absent:** H. A. Deane  
H. G. Lamker  
A. C. Ziebell

Directors-elect (as observers)  
T. H. Benners, Jr. A. M. Fulton  
N. J. Dunbeck V. E. Zang

### Reading and Approval of Minutes

Reading of the Minutes of the meeting of the Board of Directors held in Philadelphia May 4, 1948 were dispensed with, these Minutes having been approved previously by letter ballot.

Minutes of the Meeting of the Executive Committee held in Hot Springs, Va., June 8, 1948 were read, and on motion by Director Delahunt, seconded by Director Horlebein were approved as read.

### Reports of Officers

**Report of Secretary**—Secretary Wm. W. Maloney presented the Report of the Secretary covering activities of the Society for the fiscal year ended June 30, 1948, which report is attached and made a part of these Minutes. On motion by Vice-President-elect Horlebein, seconded by Director Wittlinger, the Secretary's Report was accepted.

**Report of the Treasurer**—The treasurer presented the official Auditor's Report of the Society's finances for the fiscal year ended June 30, 1948 as prepared by Lybrand, Ross Bros. & Montgomery, Chicago, Auditors approved by the President. On motion by Director Simpson, seconded by Director Robb, the Auditor's Report was accepted.

**Report of Technical Director**—Technical Director S. C. Massari reported on the Society's technical activities for the fiscal year ended June 30, 1948, a copy of this report being made a part of these Minutes. On motion by Director Wood, seconded by Director Riecks, the report was accepted.

Director Riecks discussed briefly extensive research on sand reclamation by the Ford Motor Company, and offered the Society any data emanating from this work. The Board expressed to Director Riecks its appreciation for his company's spirit of cooperation with the processes of the Society.

President Kuniansky thanked the Executive Committee for its cooperation during the past fiscal year, and the willingness of Committee Members to give considerable time to Committee deliberations.

President Kuniansky reported on progress of the National Castings Council, particularly in connection with the proposed Safety & Hygiene Program, and read a report by AFS representative L. C. Wilson on a meeting of the Safety & Hygiene Committee of the Council on June 30, 1948. It was the consensus that the raising of funds by individual members of the Council was not fully practicable and that AFS could not enter into a fund raising program for this purpose. President-elect Wallis indicated, with Board approval, that he would approach the National Castings Council toward the end of obtaining a more definite statement of policy on progressing the Safety & Hygiene Program.

### Resolution on Amending Articles of Incorporation

It being pointed out that the resolution on amending the Society's Articles of Incorporation required action by the Board for completion of the record, the following resolution was offered:

RESOLVED that the Articles of Incorporation of American Foundrymen's Association be amended to accomplish a change in name to American Foundrymen's Society effective July 1, 1948 as called for by the ballot of the eligible voting members returned to the Society May 15, 1948 in approval of the proposed change in name.

On motion by Director Horlebein, seconded by Director Robb, the resolution was unanimously approved.

The Secretary indicated that the original Articles of Incorporation of American Foundrymen's Association of 1916 required additional changes. The Secretary was instructed to suggest such further amendments at the next meeting of the Board of Directors.

### AFS Retirement Trust

The Secretary reported developments in establishing the American Foundrymen's Society Retirement Trust and presented for approval on official resolution, as recommended by the Executive Committee, for initiation of the plan July 1, 1948.

On motion by Director Horlebein, seconded by Director Delahunt and carried, the resolution was unanimously approved.

In view of the provisions for the allowance of Service Credits under the Retirement Plan for employees having continuous service in AFS for 3 years prior to July 1, 1948, Board approval was requested on the recommendation of the Executive Committee. On motion by Director Simpson, seconded by Director Smith and carried, the Retirement Service Credits as presented by Marsh & McLennan, Insurance Consultants of AFS, were approved.

### Approval of Research Projects

Technical Director Massari presented the following program of Research Projects for the fiscal year 1948-49 as recommended by the Executive Committee on June 8:

Aluminum and Magnesium Research	\$ 6,500.00
Sand Research	6,000.00
Heat Transfer Research	2,000.00
Brass and Bronze Research	5,000.00
Malleable Research	5,000.00
Steel Research	5,000.00

Total \$29,500.00

On motion by Director Horlebein, seconded by Director Robb and carried, AFS proposed Research Projects totalling \$29,500.00 as recommended by the Technical Director were approved for the fiscal year 1948-49.

### Member Dues Rates

Recommendation of the Executive Committee to increase dues of Personal members for revenue purposes was discussed at length, it being pointed out that the new dues structure, effective July 1, 1945, had increased the dues of Company and Sustaining members without increasing the dues of Personal members. Figures were presented showing average income and expense per member as follows:

### Average Income vs. Average Expense Per Member

1937-38 to 1946-47, inclusive

	Average Income	Average Expense		Average Income	Average Expense
1937-38	\$39.83	\$35.97	1942-43	\$20.78	\$23.16
1938-39	16.10	26.81	1943-44	32.09	23.46
1939-40	48.11	36.26	1944-45	24.03	24.03
1940-41	16.07	23.20	1945-46	48.86	\$4.92
1941-42	30.80	25.74	1946-47	27.86	\$7.87

Average 10 years—Income \$30.45, Expense \$29.44.

### Increased Member Dues

Director Simpson pointed out that since expected revenue from such an increase would be a definite part of the 1948-49 Budget, that the matter should be referred to the new Board for action. It was pointed out, however, that action of the Executive Committee required continuous action by the retiring Board of



Directors. On motion by Director Horlebein, seconded by Director Wittlinger, the recommendation to increase dues of Personal Members from \$8.00 to \$10.00 per year was carried, Director Simpson dissenting.

#### *American Foundryman Advertising Rates*

Recommendation of the Executive Committee to increase advertising rates in AMERICAN FOUNDRYMAN from the present \$225 basic page rate to \$250, effective January 1, 1949 was presented, it being pointed out that the proposed rates were those approved by the Board in 1947 but deferred on Staff recommendation. On motion by Director Horlebein, seconded by Director Simpson and carried, the proposed rate increase was unanimously approved.

#### *American Foundryman Subscription Rates*

Recommendation of the Executive Committee to revise subscription rates of AMERICAN FOUNDRYMAN was presented proposing a rate of \$3.00 domestic and \$6.00 foreign, instead of the present \$4.00 Member and \$6.00 non-member, subscription rates being required by postal regulations. On motion by Director Horlebein, seconded by Director Simpson and carried, the recommendation of the subscription rates was unanimously approved.

#### *Convention Registration Fees*

Recommendation of the Executive Committee to increase Registration Fees at the next Non-Exhibit Convention to \$5.00 for Members and \$7.50 for Non-Members, instead of the 1947 rate of \$2.00 for Non-Members with Members admitted free, was discussed. It was pointed out that Registration Fees for Exhibit Conventions should be discussed with Exhibitors before taking action. On motion by Director Simpson, seconded by Director Gregg and carried, the recommendation to increase Registration Fees at the St. Louis Non-Exhibit Convention in 1949 to \$5.00 Members and \$7.50 Non-Members was approved.

#### *Distribution of Bound Volumes*

Recommendation of the Executive Committee was presented to (a) distribute Paper Bound copies of the Bound Volumes of TRANSACTIONS to Members at a price of \$1.00 each; (b) continue the distribution of cloth bound copies to Company, Sustaining and Honorary Life Members gratis on request, and (c) to price cloth bound copies to Members at a minimum of 10 per cent over cost and to Non-Members at a minimum of 50 per cent over cost. On motion by Director Horlebein, seconded by Director MacNeill and carried, recommendations of the Executive Committee were approved.

#### *Use of Technical Development Program Funds*

Considerable discussion took place on the use of Technical Development Funds for current research and special publication operating expenses. It was pointed out that the solicitation of TDP Funds from the industry had been made with the under-

standing that these funds could be used for these purposes and that the fund would be continuously maintained. Some opinion was expressed that the Society possessed too many special funds and that the TDP Fund should gradually be depreciated and any surplus fund of the Society be maintained in one single Reserve Fund. Other Directors expressed the belief that use of the TDP Funds directly concern the Budget for the new fiscal year and therefore should be considered by the new Board of Directors even though the retiring Executive Committee had recommended use of TDP Funds for financing Research Projects and special publications.

Accordingly, on motion by Director Simpson, seconded by Director Wood and carried, the question of utilizing TDP Funds for recommended purposes was tabled until the First Meeting of the 1948-49 Board of Directors.

#### *By-Laws Ballot*

The Secretary reported for record on the ballot returned by the members on May 15, 1948 approving all recommendations of the Board for revision of the Society's By-Laws, effective July 1, 1948. The official report of the ballot tellers was made a part of these minutes.

#### *New Chapters*

The Secretary reported that a petition had been received for formation of a Student Chapter at the University of Illinois, Urbana, Illinois, signed by the required number of students, and indicating that all Student Chapter requirements had been met. Therefor, it was recommended that the petition be accepted, and on motion by Director Robb, seconded by Director Wittlinger and carried, the petition for formation of a 6th AFS Student Chapter at the University of Illinois was unanimously approved. Ballot was circulated and Board signatures obtained.

The Secretary reported interest in formation of a new Kansas City Chapter, and indicated that an early organizational meeting was being planned.

#### *Recommendations of the President*

President Kuniansky expressed his personal thanks to all Members of the Board of Directors for the cooperation given him during his term of office, and offered several recommendations for continuing the progress of the Society, especially as concerned Chapter assistance and establishment of more definite financial policies.

#### *Adjournment*

There being no further business to come before the meeting, the Annual Board Meeting was declared adjourned and the 1947-48 Board of Directors disbanded.

Respectfully submitted,

WM. W. MALONEY,  
Secretary-Treasurer

APPROVED:

MAX KUNIAISKY, President

# BY-LAWS

## American Foundrymen's Society

(EFFECTIVE JULY 1, 1948)

Organized, Philadelphia, 1896

Incorporated under the laws of the State of Illinois, 1916; amended July 1, 1948.

IN ACCORDANCE with action of the Board of Directors on Jan. 22, 1948, letter ballots were prepared and submitted to the eligible membership of the Association for voting on a proposed general revision of the By-Laws, all ballots to be returned by May 15, 1948. Tellers appointed by the President counted ballots received and reported as follows:

### Tellers' Report

Article	Full Ballots		Split Ballots		Total Ballots	
	For	Against	For	Against	For	Against
I	1946	40	83	143	2029	183
II	1946	40	183	45	2129	85
III	1946	40	163	62	2109	102
IV	1946	40	192	33	2138	73
V	1946	40	197	22	2143	62
VI	(not revised)	—	—	—	—	—
VII	1946	40	213	8	2159	48
VIII	(not revised)	—	—	—	—	—
IX	1946	40	214	9	2160	49
X	1946	40	217	7	2163	47
XI	1946	40	216	9	2162	49
XII	1946	40	210	20	2156	60
XIII	1946	40	215	8	2161	48
XIV	1946	40	216	9	2162	49
XV	1946	40	213	14	2159	54
XVI	1946	40	210	17	2156	57
XVII	1946	40	205	19	2151	59
XVIII	(not revised)	—	—	—	—	—
XIX	1946	40	215	10	2161	50
XX	1946	40	218	5	2164	45
XXI	1946	40	218	4	2164	44

TOTAL BALLOTS CAST 1986 233 2219

Illegal Ballots destroyed 4

Grand Total Ballots cast 2223

ABOVE TOTALS CERTIFIED CORRECT MAY 15, 1948—

### Tellers:

James Thompson, *Chairman*

Roy W. Schroeder

A. S. Klopff

The proposed By-Laws thus were completely approved Association membership, and are shown below:

### ARTICLE I

#### Name and Object

Section 1. This Society shall be known as the American Foundrymen's Society.

Section 2. The object of this Society shall be to advance the arts and sciences relating to the manufacture and utilization of metal castings.

### ARTICLE II

#### Membership

Section 1. The membership of this Society shall consist of Personal Members, Company Members, Sustaining Members, Honorary Members, Honorary Life Members, and Student and Apprentice Members. These members are defined as follows:

(a) *Personal Member*. Any individual interested in advancing the objects of the Society.

(b) *Company Member*. Any firm or organization supporting the objects of the Society. Company membership privileges shall apply only to the plant or office where such membership is held.

(c) *Sustaining Member*. Any individual, firm or organization desiring to make substantial contribution to further the objects of the Society. Sustaining members shall enjoy all privileges of Company members.

No Company or Sustaining membership will qualify individuals for Personal membership at minimum annual dues, where such individuals reside outside the territory of the Chapter in which such Company or Sustaining membership is held.

(d) *Honorary Life Member*. Any individual whose services in connection with the objects of the Society, make him pre-eminent among his fellows. Election to Honorary Life membership shall be by three-fourths vote of the Board of Directors acting upon recommendation of the Board of Awards.

(e) *Student and Apprentice Members*.

(1) *Student Member*. An undergraduate in a school of recognized standing.

(2) *Apprentice Member*. A regularly indentured apprentice in the foundry industry.

Student and Apprentice memberships shall expire at termination of the regular scholarship course or indenture period, at which time such individuals may apply for Personal membership.

Section 2. Any individual, firm or organization interested in the objects of this Society may make application for membership. Application shall be made in writing accompanied by payment covering one year's dues. When application is approved, applicant shall be enrolled in the proper classification.

Section 3. The Board of Directors may make provision for special memberships and shall determine the dues and privileges of such memberships.

Section 4. Resignations shall be submitted in writing to the Secretary.

Section 5. Any member of the Society may be suspended or expelled from membership for due cause by a two-thirds vote of the Board of Directors, with simultaneous suspension of all membership privileges, provided said member is notified in writing, at least 30 days in advance, of the Board Meeting at which suspension or expulsion is to be considered.

### ARTICLE III

#### Dues

Section 1. Dues for Members shall be as follows:

(a) Sustaining Members, minimum annual dues \$100.

(b) Company Members, annual dues \$50.

(c) Personal Members, annual dues \$15. Exceptions: (1) Individuals employed by Sustaining or Company Members at plant or office where such membership is held, annual dues \$8. (2) Individuals engaged solely in educational or domestic government work, annual dues \$8.

(d) Student or Apprentice Members, annual dues \$4.

(e) Honorary Life Members shall be exempt from payment of dues.

Section 2. The schedule of dues shown in Section 1 may be amended by three-fourths vote of the Board of Directors.

Section 3. All dues shall be paid to the order of American Foundrymen's Society, and shall become due and payable on the first day of the month following the date on which the membership is entered on the rolls of the Society. A portion of said dues shall apply to cover subscription to *American Foundryman*, official magazine of the Society, in accordance with regulations of the U. S. Post Office Department.

Section 4. Invoice for renewal of a member's dues shall be mailed not less than 30 days prior to the date on which said

dues become due and payable; if not paid within 30 days after due date, the member shall be declared "delinquent," the mailing of Society publications to him shall cease and he shall be so notified. "Final Notice" shall be mailed to the member at that time. If dues remain unpaid 30 days after date of "Final Notice," he shall cease to be a member of the Society.

#### ARTICLE IV

##### *Membership Privileges*

Section 1. Each Sustaining Member and each Company Member shall designate one individual who shall exercise the membership privileges of such membership and shall be entitled to one vote.

Section 2. All Sustaining or Company Member representatives, Personal Members and Honorary Life Members shall have the right to hold office and to vote on all questions submitted to the membership for letter or other ballot.

Section 3. Members of all classes shall be eligible to membership in a Chapter of the Society without payment of additional dues.

Section 4. All members shall be entitled to receive the monthly publication *American Foundryman* and such other publications as may be designated for distribution by the Board of Directors.

#### ARTICLE V

##### *Board of Directors*

Section 1. The general control of this Society shall be vested in a Board of Directors which shall manage the affairs of the Society in conformity with the laws under which the Society is incorporated and the provisions of these by-laws.

Section 2. The Board of Directors shall consist of the President, the Vice-President, the immediate Past President and 15 other members, five of whom shall be elected at the annual meeting each year to serve terms of three years each, their terms to begin at the time of the annual meeting of the Board of Directors.

Section 3. The Annual Meeting of the Board of Directors shall be held within 30 days after the close of the fiscal year, June 30. At least one other meeting of the Board of Directors shall be held during the fiscal year. All meetings of the Board of Directors shall be held at a time and place designated by the President. Nine members shall constitute a quorum.

Section 4. Special meetings of the Board of Directors may be called at any time by the President, and shall be called by him on written request of any five members of said Board. Written notice shall be forwarded to each Director not less than five days prior to any meeting.

Section 5. The Board of Directors shall make provision for the salaries of officers and compensation of all persons whose services to the Society justify such action.

Section 6. The Board of Directors may establish separate funds for specific purposes of the Society and designate the manner in which and purposes for which said funds shall be administered and employed. In order to vest and maintain such funds, the Board of Directors may allocate thereto current revenues of the Society, or may solicit additional revenues for such specific allocation.

#### ARTICLE VI

##### *Officers and Directors*

Section 1. The officers of the Society shall consist of a President, a Vice-President, a Secretary and a Treasurer. The President and Vice-President shall be elected from and by the members at the annual business meeting of the Society; the Secretary and Treasurer shall be elected by the Board of Directors at their annual meeting. Each officer so elected shall serve for one year or until his successor is elected and qualified. The Board of Directors may combine the offices of Secretary and Treasurer.

Section 2. Should a vacancy occur in the Board of Directors or in any of the elected offices, except the office of President, through death, resignation or other cause, the Board of Directors may select a member of the Society to fill the vacancy until the next annual election. Should a vacancy occur in the office of President, the Vice-President shall assume the duties of the President until the next annual election.

#### ARTICLE VII

##### *Standing Committees*

Section 1. The Board of Directors at its first meeting each fiscal year shall elect four of its members who, together with the President and Vice-President, shall constitute an Executive Committee with power to act for the Directors in the interim between meetings of the Board.

Section 2. The Executive Committee shall meet upon the call of the President or upon written request of any three members of the Committee. Four members of the Executive Committee shall constitute a quorum.

Section 3. The President-elect shall appoint, prior to the first meeting of the Board of Directors each fiscal year, two members of the Board who shall, together with the President, form a Finance Committee of three members, which shall prepare and present to the Board a Budget of Estimated Income and Expense for the fiscal year. The Finance Committee shall have control over Society expenditures, subject to Board approval. The Finance Committee shall direct, subject to Board approval, the investment of surplus Society funds.

The Finance Committee shall recommend salaries of officers and compensation of all persons whose services to the Society justify such action.

Section 4. The Technical Correlations Committee shall consist of a Chairman, appointed by the President, the President as Vice-Chairman, and all chairmen of the several technical Divisions and General Interest Committees of the Society. The Technical Correlations Committee shall meet at least once a year, upon call of the Chairman, at which time it shall receive the reports of the various technical Divisions and Committees for the purpose of coordinating their activities. The actions and recommendations of the Technical Correlations Committee shall be subject to approval by the Board of Directors.

Section 5. The President shall appoint the following Standing Committees, designating the Chairman in each case:

- National Membership Committee
- Chapter Contacts Committee
- By-Laws Committee
- Annual Lecture Committee
- International Relations Committee
- Auditing Committee
- Publication Committee

The duties of the Standing Committees shall be as follows:

(a) The National Membership Committee shall include the Chairman of each Chapter Membership Committee and shall stimulate and encourage the work of such committees.

(b) The Chapter Contacts Committee shall encourage close contacts between the Board of Directors and the Chapters through frequent visits to Chapters by Officers and Directors of the Society.

(c) The By-Laws Committee shall study suggested revisions of the by-laws and may make recommendations to the Board of Directors in accordance with provisions in Article XX.

(d) The Annual Lecture Committee shall select, plan and arrange for (a) The Charles Edgar Hoyt Annual Lecture, and (b) the Lecture Courses for presentation at Annual Meetings of the Society. The Committee membership shall include representation of the Society's several technical interests.

(e) The International Relations Committee shall cooperate with the various technical societies recognized by the International Committee of Foundry Technical Associations in arranging for (a) exchange of technical papers, (b) holding of international Foundry Congresses, and (c) cooperation on technical matters of international interest.

(f) The Auditing Committee shall examine and study the annual audit of the Society's books and shall make reports and recommendations to the Board of Directors.

#### ARTICLE VIII

##### *Duties of Officers*

Section 1. The President shall be the executive head of the Society. He shall preside at all Annual and Special Meetings of the members of the Society, and at all meetings of the Board of Directors and of the Executive Committee.

Section 2. The Vice-President shall perform the duties of the President in the absence of the President or the inability of the President to act.

Section 3. The Secretary shall, under the direction of the President and the Board of Directors, be the administrative officer of the Society. He shall keep, or cause to be kept, full and complete records of the membership of the Society, and shall check, or cause to be checked, the records of the National Office with the records of the secretaries of the Chapters. He shall act as secretary at all meetings of the members of the Society, of the Board of Directors, and of the Executive Committee, and perform such other duties as shall be delegated to him by the Board of Directors. The office of Secretary shall be at the principal office of the Society.

Section 4. The Treasurer shall be the financial officer of the Society. He shall keep, or cause to be kept, complete records of all moneys received and disbursed on behalf of the Society. He shall deposit the funds of the Society in such bank or banks as



shall be approved by the Board of Directors. He shall report to the Board, whenever requested, the financial condition of the Society, and perform such other duties as shall be prescribed from time to time by the Board of Directors. He shall give a bond, the amount of which shall be fixed by the Board of Directors, the premium on said bond to be paid by the Society.

Section 5. The duties of all officers or employees of the Society as defined in these by-laws or prescribed by the Board of Directors shall be determined by the Secretary.

Section 6. As an order of business at the Annual Meeting of the Board of Directors there shall be submitted reports of the President, the Secretary, the Treasurer and all committees of the Board. The Treasurer's Annual Report for the fiscal year ending June 30 shall be accompanied by an audit made by certified public accountants named by the President.

#### ARTICLE IX

##### *Annual Meetings*

Section 1. There shall be an Annual Meeting of this Society, the date and location of which shall be announced by the Board of Directors at least three months in advance. The fiscal year of the Society shall begin July 1 and end June 30.

Section 2. The annual Business Meeting of the Society shall be held as one of the sessions at the Annual Meeting, and shall be so designated.

Section 3. The registration fees for members and non-members at Annual or Special Meetings of the Society shall be as determined by the Board of Directors.

#### ARTICLE X

##### *Nominations of Officers and Directors*

Section 1. The Board of Directors of each Chapter eligible to have a member on the Nominating Committee, as herein provided, shall annually select two candidates for the Nominating Committee from the Chapter membership, preferably representing different branches or divisions of the industry within the membership. The names of the candidates shall be forwarded to the President on or before July 1 of each year.

Section 2. On or before October 1 of each year the Executive Committee of the Board of Directors shall appoint a Nominating Committee of seven members, six of whom shall be from the list of eligible candidates submitted by the various Chapters, and one of whom may be from the list of members residing outside Chapter territories, who, together with the last two living Past Presidents, shall constitute a Nominating Committee of nine members.

In appointing the Nominating Committee the Executive Committee shall select members thereof in such a manner as to provide equitable, regional representation, and shall consider proportional representation for the several branches and divisions of the membership.

Section 3. Any Chapter having a member on the Nominating Committee in any given year shall not be eligible to have a member on the Nominating Committee during the two succeeding years.

Section 4. The immediate Past President serving on the Nominating Committee shall be the Chairman. In the absence of the Chairman, the Nominating Committee shall select a Chairman. Five members of the Nominating Committee shall constitute a quorum.

Section 5. The names of the members of the Nominating Committee shall be published to the members of the Society not later than November 15 of each year, following their appointment.

Section 6. If, for any reason, there shall be a vacancy or vacancies on the Nominating Committee, said vacancy shall be filled by the President from the remaining list of eligible candidates proposed by the Chapters.

Section 7. On any day at least 90 days prior to the annual Business Meeting, the Nominating Committee shall meet at a time and place designated by the Chairman and acting in accordance with procedure approved by the Board of Directors, shall name candidates for the office of President and the office of Vice-President and for each directorship that shall become vacant at the annual meeting of the Board of Directors held in accordance with the provisions of these by-laws.

Section 8. Immediately after the candidates are thus nominated, the Chairman shall report the names of the nominees to the Secretary, who shall publish said report either by mail or in the *American Foundryman* to all members of the Society at least 60 days before the Annual Business Meeting.

Section 9. After the report of the Nominating Committee has been published, and at any time 45 days prior to the date of

the annual Business Meeting, additional nominations may be made by written petition filed with the Secretary and signed by 35 members in good standing.

#### ARTICLE XI

##### *Elections*

Section 1. Should no candidates for officers and directors be nominated in the manner provided in Article X, Section 9, the Secretary shall, at the annual Business Meeting, cast the unanimous ballot of all members for the election of the candidates named in the report of the Nominating Committee, and as published to the membership.

Section 2. In the event additional candidates are nominated for any office or directorship in the manner provided in Article X, Section 9, then the election of officers and directors shall be by letter ballot of the members.

Section 3. In the case of election by letter ballot, it shall be the duty of the President to appoint three judges of election who shall have sole charge of the election. The judges shall direct the Secretary to have printed and mailed to all members in good standing, ballots bearing the names of all candidates nominated in accordance with provisions of these by-laws. The judges shall name the final date and place for receiving of ballots.

Section 4. When polls are closed the judges shall examine all ballots cast and shall certify the results of the election to the President at least 24 hours prior to the annual Business Meeting of the Society.

Section 5. The newly elected officers and directors shall assume office at the annual meeting of the Board of Directors, which shall be held within 30 days after the close of the fiscal year, June 30.

#### ARTICLE XII

##### *Advisory Board*

Section 1. All Past Presidents of this Society shall constitute an Advisory Board, and may be called upon for advice by the Board of Directors.

#### ARTICLE XIII

##### *Board of Awards*

Section 1. The Board of Awards shall consist of the last seven living Past Presidents of the Society, which Board shall be self-perpetuating as each retiring President becomes a member. The Senior Past President shall be the Chairman.

Section 2. The Board of Awards shall make awards subject to the terms and conditions set forth in award agreements. All awards made in the name of the Society shall be determined by the Board of Awards, subject to approval by the Board of Directors.

Section 3. The Board of Awards may recommend for approval by the Board of Directors, the names of individuals for consideration as Honorary Life Members.

#### ARTICLE XIV

##### *Awards*

Section 1. The Board of Directors may, under regulations which it may adopt, offer medals or awards for exceptionally meritorious investigations or developments which benefit the foundry industry. The cost of such medals or awards may be defrayed from funds of the Society or from any funds which may be given to the Society for such purposes.

#### ARTICLE XV

##### *Divisions and Committees*

Section 1. Technical committee groups representing the several major technical interests of the castings industry, to be known as Divisions of the Society, may be authorized by the Board of Directors. A Manual for the conduct of the Divisions, approved by the Board of Directors, shall govern the activities and procedure of all Divisions.

Section 2. Technical committee activities of interest to two or more of the Divisions of the Society, to be known as General Interest Committees, may be authorized by the Board of Directors.

Section 3. Special technical interest activities or groups, to be known as "Projects" of the Society, may be authorized by the Board of Directors, and rules for the conduct of such Projects shall be approved by the Board. Funds raised or allocated to such Projects, when administered as an activity of the Society, shall be controlled by or expended with the approval of the Board of Directors.



## ARTICLE XVI

*Research*

Section 1. The Board of Directors may authorize research on problems of value to the castings industry, on recommendation of either the Technical Correlations Committee, the several technical Divisions, or the several General Interest Committees of the Society. The Board may make provisions for financing such research activities, by the allocation of existing Society revenues or the raising of additional revenues.

Section 2. Regulations for the origination, approval, placement, conduct and reporting of Society-sponsored research projects shall be determined by the Board of Directors.

## ARTICLE XVII

*Chapters*

Section 1. Organization of Chapters may be authorized by the Board of Directors when a sufficient number of members of the Society residing within a given territory file written evidence of their desire to organize and hold meetings for advancing the objects of the Society. The approval of a Chapter petition shall be by three-fourths vote of the Board of Directors.

Section 2. An identifying name shall be chosen by the membership of the Chapter organizing group at its first meeting subject to the approval by the Secretary of the Society and thereafter such Chapter shall be so designated.

Section 3. The territorial boundaries of the Chapter shall be subject to the approval of the Secretary of the Society and thereafter shall be kept on file at the Society headquarters. The Board of Directors shall have the right to readjust the boundaries of any Chapter.

Section 4. Any member of the Society in good standing shall be eligible for Chapter membership as provided in Art. IV, Sec. 3 of these by-laws. He may indicate the Chapter in which such membership is preferred. No member of the Society shall be recorded as holding membership in more than one Chapter. Only members of the Society in good standing shall be eligible to hold office in any Chapter.

Section 5. Each Chapter shall be governed by a set of by-laws, which must be approved by the Board of Directors of the Society, following which they shall be submitted to and adopted by the Chapter membership. No action or obligation incurred by a Chapter shall be considered an action or obligation of the Society without prior approval by the Board of Directors of the Society. The by-laws of the Society and the Manual for the Conduct of Chapters, as approved by the Board of Directors of the Society, shall govern the procedure of all Chapters.

Section 6. No Chapter shall issue or distribute any technical publication over the name of the Society without prior approval by the Publication Committee of the Society. Presentation of papers at a Chapter Meeting or Regional Conference, or distribution of same to the Chapters or membership, shall not be construed as technical publication over the name of the Society.

Section 7. The Board of Directors of the Society, by three-fourths vote, shall have the right to suspend or dissolve a Chapter of the Society for good and sufficient reasons, after serving 60 days' notice of intention to do so. A Chapter thus suspended may be reinstated by the Board of Directors. A Chapter thus dissolved may resume Chapter status only in accordance with the foregoing provisions of this Article.

Section 8. For financing Chapters the Board of Directors of the Society may authorize appropriations which shall not exceed twenty per cent (20%) of the annual dues received from members of a Chapter.

Section 9. If the records of the Secretary show that any Chapter has held not less than six regular monthly meetings during

the fiscal year and the aggregate remittance of dues by the Society to such Chapter during the fiscal year has been less than two hundred (\$200) dollars, the Board of Directors may authorize payment to such Chapter of the difference between two hundred (\$200) dollars and the dues actually remitted to the Chapter. Additional financial support of a Chapter may be authorized by the Board of Directors of the Society, on recommendation of the Finance Committee (Art. VII, Secs. 3-5) following submission by the Chapter of detailed financial report indicating such need.

Section 10. The Chapters shall submit not later than July 15 to the Treasurer of the Society an annual financial report. This report shall be in accordance with the requirements of the Manual for the Conduct of Chapters.

## ARTICLE XVIII

*Seal and Emblem*

Section 1. The emblem of the Society shall be of circular form with the initials and date of organization of the Society occupying the cardinal points and connected by double rings. Crossing said rings at right angles, and joined in the center by a vertically placed trowel, a rammer and shovel shall be represented.

Section 2. Pins or other articles bearing the emblem of the Society may, at the discretion of the Board of Directors, be issued or sold to members.

Section 3. The seal of the Society shall be a reproduction of the emblem, with sufficient additional space on the circumference to show these words surrounding the emblem proper: "Seal of the American Foundrymen's Society."

## ARTICLE XIX

*General Provisions*

Section 1. No detail of foundry practice or design of foundry appliances shall be adopted as a standard of the Society or recommended by it except subject to the approval of a three-fourths vote of the directors.

Section 2. The Society shall not be responsible for statements or opinions advanced in papers or in discussions at meetings of the Society or of its Divisions, or Sections, or printed in its publications.

## ARTICLE XX

*Amendments*

Section 1. These by-laws may be amended only by majority of votes cast by letter ballots which shall be submitted by the Secretary to all members in good standing, residing in the United States, Canada and Mexico. Said letter ballots must bear signature of all those voting and they must be received at the Secretary's office within 30 days after being mailed therefrom to the members, in order to be included in the canvass of votes. Preliminary to the submission to the membership of said letter ballots, action by a two-thirds vote of the Board of Directors of any meeting thereof or by letter ballot shall take place, favoring any amendment to be submitted. The Secretary shall mail letter ballots to members within 30 days after adjournment of the meeting or vote authorizing their submission.

Section 2. In the case of ballot on amendments, the President shall appoint judges of election who shall count the ballots at the time and place designated by the President, and submit their report to him.

## ARTICLE XXI

*Rules of Order*

Section 1. Roberts' Parliamentary Rules of Order shall be recognized as authority by this Society and its Chapters and shall govern the deliberations in all cases not covered by these by-laws.

## A.F.S. Staff Retirement Plan

ON JULY 1, 1948, a Staff Retirement Plan was inaugurated for salaried employees of American Foundrymen's Society, following approval by the Board of Directors after several years of discussion and investigation. The plan enables the building up of a retirement income payable, under normal conditions, at age 65, with provisions in keeping with those of similar plans in effect in the foundry and other industries.

The plan is known as the American Foundrymen's Society Retirement Trust and will be administered by four Trustees, appointed by the Board of Directors to serve "until successors are appointed." Only one of the Trustees may be a member of the Board of Directors. Trustees appointed as of July 1, 1948 are: Chairman, B. L. Simpson, National Engineering Co., Chicago, member of Board of Directors A.F.S. (term expires 1950); F. W. Shipley, Caterpillar Tractor Co., Peoria, Ill.; R. L. Lee, Grede Foundries, Inc., Milwaukee, Wis.; H. Bornstein, Deere & Co., Moline, Ill.

Some provisions of the Staff Retirement Plan are as follows:

1. Employees 30 years of age, having served A.F.S. for three consecutive years, are eligible to join the Plan.
2. Normal retirement date is age 65. Retirement prior to age 65, with approval of the Society, is possible but with consequent reduction in the amount of retirement income.
3. Employee contributions, based on earnings during the previous calendar year, consist of 3 per cent of the first \$3,000 of earnings, plus 6 per cent of earnings in excess of \$3,000.
4. Society contributions exceed employee contributions and consist of the balance, annually, necessary to provide the retirement benefits as determined actuarially.
5. Annual retirement income, commencing at normal retirement date and continuing for life, provides 1 per cent of the first \$3,000 of earnings plus 2 per cent of earnings in excess of \$3,000, for each year of participation in the plan. Minimum re-

tirement income \$300.00 per year, maximum \$6,000 per year. Retirement income is, in any event, guaranteed for a period of ten years after retirement.

6. On termination of service during the period of the plan, by death or otherwise, all contributions of the employee will be refunded to him or his beneficiary, plus 2 per cent compound interest.

7. Contributions of employees and the Society are used for the purchase of retirement annuity insurance policies annually. Borrowing against the value of these policies is not permitted.

8. While participation in the plan is entirely voluntary, withdrawal from participation while still an employee of A.F.S. is not permissible, since effectiveness of the plan depends upon continuous participation.

All matters pertaining to the Retirement Trust are under the control of the Trustees, who serve without compensation. Prior to each anniversary date, the Society certifies to the Trustees all data on employees becoming eligible to participation. The Trustees then inform the Society of the amount required for the purchase of retirement annuity policies for newly eligible participants. The Society then deposits in the special Premium Account the amount required, from which the Trustees make the necessary withdrawals of funds. The Trustees are the owners of all policies acquired by them under the plan, although policies will be held by A.F.S. Since the Society is incorporated in the State of Illinois, the laws of that state shall govern all questions arising with respect to the Trust agreement.

While the Society hopes and expects to continue the Retirement Plan indefinitely, the Society necessarily reserves the right to discontinue or amend the Plan by action of the Board of Directors. In the event of termination of the Plan, each participant will receive in cash or paid-up benefits the values that may have accrued from both his own and the Society's contributions.

## Fifth Annual Chapter Chairman Conference

June 28-30, 1948—Stevens Hotel, Chicago

### PROGRAM

#### Monday, June 28

##### 10:30 AM MORNING PROGRAM

Presiding Chairman, E. W. Horlebein, *Introductory Remarks* on purpose of Chapter Chairman Conference.

*Welcome to Chapter Officers*..... Pres.-Elect W. B. Wallis

*'Round-the-Table Introductions* Chairman E. W. Horlebein

*Know Your AFA* ..... Pres.-Elect W. B. Wallis

An outline of the development and organization of AFA, including its origin, basic policies and purposes, and something of its position in the Foundry Industry.

*Technical Activities of AFA* ..... Tech. Dir. S. C. Massari

An outline of AFA service as a Technical Society. Organization and functions of Technical Activities. Research Projects . . . initiation . . . approval operation. Technical Publications available and forthcoming.

##### 2:00 PM AFTERNOON PROGRAM

Chairman Horlebein, presiding.

*Serving the Membership* ..... Promotion Mgr. T. B. Koeller

A presentation of newly developed National Office methods, set up specially for handling promptly and efficiently new memberships, membership renewals and membership changes. Charts to visualize the processing of membership. National Office records in relation to Chapter records. Cooperation in holding delinquent members.

*The New AFA By-Laws* ..... Secretary Wm. W. Maloney

A comparison of the main By-Laws revisions, effective July 1, 1948, especially as they affect the name of the Association, membership requirements and dues, nominations and elections, and Chapter operations. An orientation and interpretation session.

*National Castings Council* ..... Pres.-Elect W. B. Wallis

A brief outline of the purposes, functions and program of the National Castings Council . . . its make-up as an over-all body of the Foundry Industry . . . Projected Activities.

##### 6:30 PM DINNER FOR CHAPTER CHAIRMEN

Presiding—Chairman E. W. Horlebein.

Announcement—Secretary W. W. Maloney.

Guest Speaker—E. A. McFaul, Northwestern University, Evanston, Illinois.

Subject: "How's Your Sense of Humor?"

#### Tuesday, June 29

##### 9:30 AM MORNING PROGRAM

*Chapter Operations Panel* Chairman Horlebein, Conducting

(1) *Chapter Program Building*

(2) *Management Meetings*

(3) *National Officer Visitations*

(4) *Membership Activities*

(5) *Nominations and Elections*

(6) *Chapter By-Laws*

(7) *Chapter Finances*

##### 2:00 PM AFTERNOON PROGRAM

Chairman Horlebein, presiding.

*Educational Activities of AFA* ..... F. G. Seifing

"The Shake Out" ..... Chairman E. W. Horlebein

#### Wednesday, June 30

##### 9:00 AM MORNING PROGRAM

Chairman Horlebein, presiding.

Entire program was devoted to bringing out fine points on how to arrange and conduct meetings, plus an audience participation session.

(1) *Preface to Morning Program*

General explanation of Chairman's job and intent of program being presented.

(2) *Courtesies to the Guest Speaker*

Some points frequently overlooked in making the Guest Speaker feel he is appreciated, thus creating an obligation to make a good talk.

(3) *The Meeting Room and Speakers' Table Setup*

Importance of a properly arranged meeting room, and various types of setups. Arranging the Speakers' Table for maximum importance and efficiency. Some reminders on ticket handling.

(4) *Arranging the Program*

Importance of looking after details of time, arrangements, facilities in the interest of smoothly run meetings. Arranging the technical program for maximum value to the audience and satisfaction of the speaker.

(5) *Proper Microphone Technique*

Do's and Don'ts in using a microphone, and how to get the most out of it.

(6) *How to Induce Technical Discussion*

Some helpful hints on this perplexing problem, and some methods of creating more discussion at your meetings.

(7) *The Speech of Introduction*

Its importance, how to make a good one . . . points to avoid . . . when to be funny and when to avoid humor.

(8) *Getting Up on Your Feet*

##### 2:00 PM FINAL ADJOURNMENT OF 5TH ANNUAL CHAPTER CHAIRMAN CONFERENCE.

### ATTENDANCE

CONFERENCE CHAIRMAN—A.F.S. Vice-President-Elect E. W. Horlebein, president, Gibson & Kirk Co., Baltimore, Md.

PRESIDENT-ELECT—W. B. Wallis, president, Pittsburgh Lectromelt Furnace Corp., Pittsburgh.

GUEST SPEAKERS—F. G. Seifing, metallurgist, International Nickel Co., New York; Professor E. A. McFaul, Northwestern University; and Harold F. W. Hauslein, Harold F. W. Hauslein & Associates, engineers and contractors, Chicago.

GUESTS—National Director Bruce L. Simpson, president, National Engineering Co., Chicago; G. K. Dreher, executive director, Foundry Educational Foundation, Cleveland; A.F.S. Secretary Emeritus R. E. Kennedy; and F. G. Steinebach, Penton Publishing Co., Cleveland.

REPRESENTATIVES OF PROPOSED KANSAS CITY CHAPTER—C. W. Culbertson, sales engineer, M. W. Warren Coke Co., chairman of the steering committee; and J. T. Westwood, president



- Blue Valley Foundry Co., both of Kansas City, Mo.
- BIRMINGHAM—Director J. A. Bowers, melting supt., American East Iron Pipe Co., Birmingham, Ala.
- EAST COLUMBIA—Chairman T. Cowden, gen. mgr., Wm. McPhail & Sons (Can.) Ltd., Vancouver, B.C., and Vice-Chairman J. A. Dickson, owner, Dickson Fdry. Co., Vancouver, B.C.
- EASTERN DISTRICT—Chairman E. H. Taylor, plant engr., F. E. Myers & Bros. Co., Ashland, Ohio.
- CENTRAL ILLINOIS—Chairman F. W. Shipley, fdry. mgr., Caterpillar Tractor Co., Peoria, Ill., and Vice-Chairman, C. W. Bucklar, supt., Superior Fdry. Co., E. Peoria, Ill.
- CENTRAL INDIANA—Chairman R. Langenkamp, secy., Langenkamp-Wheeler Brass Works Indianapolis and Vice-Chairman H. L. Creps, res. sales mgr., Frank Fdry. Corp., Muncie, Ind.
- CENTRAL MICHIGAN—Chairman C. C. Sigerfoos, asst. prof. mech. eng., Michigan State College, E. Lansing, Mich., and Vice-Chairman F. Coghlin, met., Albion Malleable Iron Co.
- CENTRAL NEW YORK—Chairman C. M. Fletcher, supt., Fairbanks Co., Endicott, N.Y., and Vice-Chairman J. F. Livingston, asst. plant mgr., Crouse-Hinds Co., Syracuse, N.Y.
- CENTRAL OHIO—Chairman F. W. Fuller, field engr., National Engineering Co., Westerville, Ohio, and Vice-Chairman W. L. Deutsch, sales engr., Columbus Malleable Iron Co.
- CHESAPEAKE—Chairman B. M. Loring, head, non-ferrous sec., U. S. Naval Research Lab., Bellevue, Washington, D.C., and Vice-Chairman J. B. Mentzer, secy.-treas., Wood Embley Brass Co., Waynesboro, Pa.
- CHICAGO—C. K. Faunt, works mgr., Christensen & Olsen Co., Chicago, and Vice-Chairman W. D. McMillan, met., International Harvester Co., Chicago.
- CINCINNATI—Chairman A. D. Barczak, vice-pres., The Bardes Forge & Fdry. Co., Cincinnati, and Vice-Chairman A. J. Smith, met., Lunkenheimer Co., Cincinnati.
- DETROIT—Chairman A. W. Stolzenburg, Aluminum Co. of America, Detroit, and Secretary J. N. Phelps, Vanadium Corp. of America, Detroit.
- EASTERN CANADA & NEWFOUNDLAND—Chairman O. L. Voisard, gen. supt., The Robert Michell Co., Ltd., St. Laurent, Que., Canada, and Vice-Chairman J. H. Newman, mgr., Newman Fdry. Supply Co., Ltd., Montreal, Que., Canada.
- EASTERN NEW YORK—Chairman A. C. Andrew, fdry. foreman, American Locomotive Co., Schnectady, N.Y., and Vice-Chairman K. F. Echard, met., Eddy Valve Co., Waterford.
- METROPOLITAN—Chairman J. A. Bukowski, met., Worthington Pump & Machinery Corp., Harrison, N.J.
- MICHIANA—Chairman K. A. Nelson, br. mgr., Chicago Hardware Fdry. Co., Elkhart, Ind.
- NORTHEASTERN OHIO—Chairman E. C. Zirzow, core room foreman, National Malleable & Steel Castings Co., Cleveland, and Vice-Chairman W. E. Sicha, met., Aluminum Co. of America, Cleveland.
- NORTHERN CALIFORNIA—Chairman G. D. McDonald, patt. shop foreman, H. C. Macaulay Fdry. Co., Berkeley, Calif., and Director J. R. Russo, vice-pres., General Fdry. Service Corp.
- NORTHERN ILLINOIS & SOUTHERN WISCONSIN—Chairman H. J. Bauman, fdry. supt., Ebaloy Foundries, Inc., Rockford, Ill., and Vice-Chairman L. C. Fill, fdry. supt., Geo. D. Roper Corp.
- NORTHWESTERN PENNSYLVANIA—Chairman J. S. Hornstein, secy., Meadville Malleable Iron Co., Meadville, Pa., and Vice-Chairman J. A. Shuffstall, asst. plant mgr., National Erie Co., Erie, Pa.
- ONTARIO—Chairman R. A. Woods, mgr., Geo. F. Pettinos, Ltd., Hamilton, Ont., Canada, and Vice-Chairman J. H. King, salesman, Werner G. Smith Co., Toronto, Ont., Canada.
- OREGON—Secretary-Treasurer G. C. Vann, asst. mgr., N. W. Fdry. & Furnace Co., Portland, Oregon.
- PHILADELPHIA—C. L. Lane, met., Florence Pipe Fdry. & Machine Co., Florence, N.J.
- QUAD CITY—Vice-Chairman E. P. Closen, fdry. foreman, John Deere Planter Works, Moline, Ill., and secretary-treasurer C. R. Marthens, owner, Marthens Co., Moline, Ill.
- ROCHESTER—Chairman M. T. Ganzauge, fdry. supt., General Railway Signal Co., Rochester, N.Y.
- SAGINAW VALLEY—Chairman A. E. Sunstedt, vice-pres., & gen. mgr., General Fdry. Co., Flint, Mich., and Vice-Chairman L. L. Clark, plant met., Buick Motor Div., Flint, Mich.
- ST. LOUIS DISTRICT—Chairman A. L. Hunt, plant supt., National Bearing Div., American Brake Shoe Co., St. Louis, and Vice-Chairman G. W. Shepherd, supt., Duncan Fdry. & Machine Co., Alton, Ill.
- SOUTHERN CALIFORNIA—Chairman L. O. Hofstetter, Brumley-Donaldson Co., Los Angeles, Calif., and Vice-Chairman E. D. Shomaker, patt. shop foreman, Kay-Brunner St. Products, Inc., Alhambra, Calif.
- TENNESSEE—Chairman G. Frank Anderson, vice-pres. & gen. mgr., Chattanooga Implement & Mfg. Co., Chattanooga, and Director K. L. Landgrebe, supt. Fdry. Div., The Wheland Co.
- TEXAS—Chairman Jake Dee, owner, Dee Brass Fdry., Houston, Texas, and Vice-Chairman C. W. Williamson, vice-pres., Trinity Valley Iron & Steel Co., Ft. Worth, Texas.
- TIMBERLINE—Chairman J. L. Higson, supt., Western Fdry. Co.
- TRI-STATE—Chairman Dale Hall, met., Oklahoma Steel Castings Co., Tulsa, Okla.
- TWIN-CITY—Chairman C. C. Hitchcock, vice-pres., R. C. Hitchcock & Sons, Minneapolis.
- WASHINGTON—Chairman G. Rauen, Olympic Steel Works, Seattle, and Secretary F. R. Young, sales engr., E. A. Wilcox Co.
- WESTERN MICHIGAN—Chairman W. A. Hallberg, fdry. engr., Lakey Fdry. & Machinery Co., Muskegon, Mich., and Vice-Chairman C. N. Jacobson, prod. mgr., Dake Engine Co., Grand Haven, Mich.
- WESTERN NEW YORK—Chairman M. J. O'Brien, asst. work mgr., Symington-Gould Corp., Depew, N.Y., and Vice-Chairman J. R. Wark, sales dept., Queen City Sand & Supply Co.
- WISCONSIN—Chairman R. C. Woodward, fdry. supt., Bucyrus-Erie Co., So. Milwaukee, and Vice-Chairman A. C. Haack, 2413 N. 65th St., Wauwatosa, Wis.

*Chapters not represented: Mexico City, Toledo.*



# GRAIN SIZE BEHAVIOR IN MAGNESIUM CASTING ALLOYS

By

Charles E. Nelson \*

IT IS THE AIM in this paper to give the reader a better appreciation of this subject by discussing its various aspects including the meaning of grain size; the manner in which it is expressed; the importance to serviceability and foundry behavior; the effect of compositional and foundry variables; the most important techniques for grain size control, e.g., superheating, carbon inoculation, agitation, and ferric chloride treatment; and the fundamental mechanism of the grain refining action.

Experimental results and conclusions from the literature or the work of the author and associates will be presented as they are pertinent to the discussion.

## Significance of Grain Size

Grain, as used in connection with metals, represents the individual metal crystals. In the case of cast metals, it represents a crystal growing from the molten metal during solidification. The size of these individual crystals or grains may vary greatly in cast metals depending upon the alloy system, the rate of solidification, and many other factors to be discussed later. In comparing one piece of metal with another it is desirable to be able to refer to the grain size. Unfortunately, there is as yet no universally accepted standard for measuring or expressing grain size.

Usually, the grain size is determined by microscopic examination and measurement on a finely polished and etched section of the metal. Grain size may be given as the number of grains per unit area, e.g., grains/mm<sup>2</sup>, as the number of grains intersected by a line of given length, or it may be assigned an average grain diameter by a calculation from one of the above measurements. In magnesium alloys it is most common to express grain size as the average grain diameter (A.G.D.) in inches. All discussion henceforth will be based on this method of expression.

There is a great deal of discussion of the grain size of magnesium alloys in the literature and in industry. This is not because the normal grain size of these alloys

is coarser than most engineering metals, indeed quite the converse is true, nor is it because these alloys require a finer grain size relative to other metals. The reasons for the interest are two-fold; first, magnesium alloys are extensively used in the aircraft industry where the maximum strength-weight ratio is required; second, magnesium alloys possess a remarkable ability to attain ultra-fine grain sizes through the application of special simple foundry techniques.

A comparison of the grain sizes of similar cross sections of commercial sand cast alloys comprising brass, aluminum, magnesium, and cast iron, indicates that magnesium was the finest with 0.004 in. average grain diameter. In Fig. 1 are shown photomicrographs of two of the most common magnesium casting alloys, AZ63 (Mg-6Al-0.2Mn-3Zn) at the top and AZ92 (Mg-9Al-0.2Mn-2Zn) at the bottom. The picture on the left in each case shows the typical structure of the alloy before any grain refining treatment, and on the right, the same alloy after a normal foundry grain refining treatment.

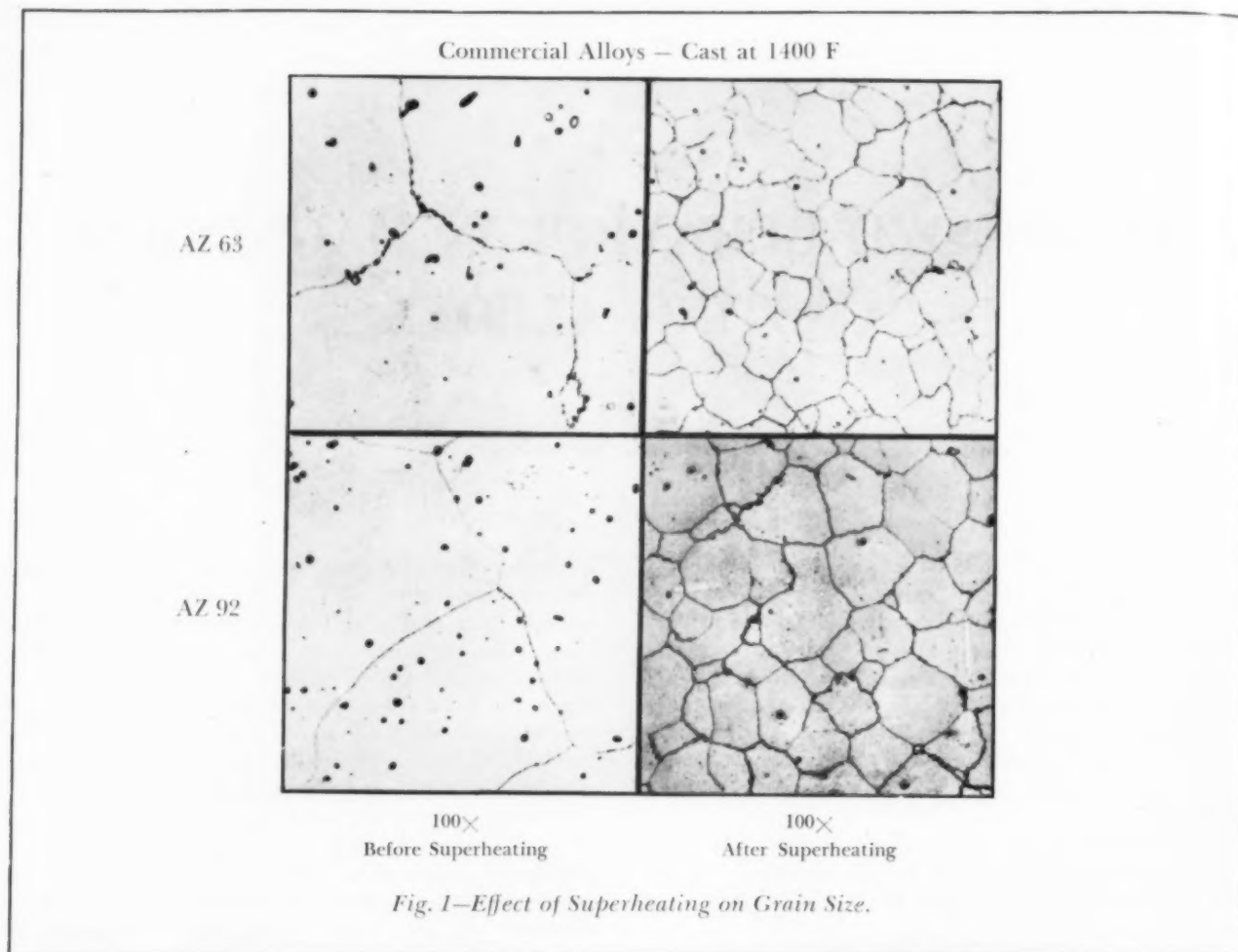
## Effect on Mechanical Properties

Returning now to the importance of grain size, its effect on mechanical properties and other factors that might affect serviceability will be discussed first. In general, finer grain size in all metals leads to higher mechanical properties such as tensile and yield strength, per cent elongation, and fatigue strength. The paper by Busk and Phillips<sup>1</sup> shows properties as a function of grain size. The tensile, ultimate, yield, and fatigue strengths are improved by 10 to 15 per cent, whereas the elongation and impact toughness of the ductile cast alloys may be improved up to 50 per cent in fine-grained alloys (0.004 in. A.G.D.) as compared to alloys with grain size of 0.020 in. A.G.D. or over. Even though these effects are significant, much more potent factors such as stress raisers (notches, poor fillets, high thin ribs, etc.) due to bad design or machining are far more likely to be the seat of any service trouble.

Before leaving the mechanical property effects of grain size, it should be pointed out that there are some commercial uses in which a moderately coarse grain size may be advantageous. Such a case would be a part

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operating at a fairly high temperature where creep may be the limiting factor. Creep resistance of some materials has been shown to increase with grain size within limits, and the usual practice in these special applications is to choose an alloy of intermediate grain size thereby retaining a good balance between fatigue strength and creep.

The secondary effects of fine grain size, at least in magnesium castings, are more significant and are of most consequence to the foundryman. The phenomenon of "hot cracking" during solidification in magnesium alloy castings can be practically eliminated by the use of good grain refining practices<sup>2</sup>. These practices appear to improve fluidity with consequent better feeding from the risers. Grain refining gives a finer and more uniform distribution of the intermetallic phases in the as-cast metal with the result that subsequent homogenizing heat treatments may be shortened or at least become more effective. One further practical advantage of fine grain is in machining, as coarse grains tend to deform under the cutting tool producing a bumpy surface, and in the worst cases chip out when heavy cuts are taken.

#### Effect of Foundry Variables on Grain Size

Before going to the discussion of grain refining treatments, some consideration will be given to common foundry variables that affect grain size. Magnesium,

like most other pure metals, is coarse grained and consequently has low mechanical properties. The grain size is reduced markedly by the addition of alloying elements such as aluminum, zinc, calcium, zirconium, and cerium. Of these, only zirconium produces the ultra fine grain size comparable to that obtained in the common casting alloys which have been given a grain refining treatment. Manganese has only a slight effect on the grain size of binary Mg-Mn alloys, but as will be seen later, it plays an important part in the control of grain size in the common casting alloys. Beryllium, even in amounts as low as 0.0005 per cent, has a unique action in that it coarsens the grain size of the Mg-Al-Mn or Mg-Al-Mn-Zn alloys.

Another factor that basically affects cast grain size is the degree of chill which the mold imparts to the poured metal. By way of example, pressure die cast magnesium alloys have such a fine grain size due to the thin sections of the casting and the chilling capacity of the mold that no grain refining treatment is effective or necessary. Permanent mold castings also show finer grain than sand castings. Due to the same chill action, thin sections of any casting will usually tend to show finer grain size than heavy sections. Heavy sections are a problem in some metals, but here, magnesium casting alloys, particularly if given a grain refining treatment, are unique in that there will be only a slight difference in grain size in thin and moderately thick sections.

One other interesting grain size behavior that needs mentioning, even though rarely occurring in practice, is the phenomenon known as germination.<sup>3</sup> When this occurs, grain growth takes place during the latter part of the homogenization heat treatment and results in large grains and reduced mechanical properties. Its exact cause is not known but it occurs only if all the following conditions are met:

1. The alloys have a fine as-cast grain size.
2. They contain aluminum greater than about 8 per cent.
3. The casting has been subject to drastic chilling in the germinating area.
4. The homogenizing heat treatment is substantially complete.
5. Manganese content is below or on the low side of specifications.

This phenomenon does not occur in the AZ63 alloy and it can be avoided in the other alloys by making certain that chills are coated with an insulating wash and by avoiding flash. To make certain it does not occur, a special non-germinating homogenizing heat treatment is used for susceptible alloys. Now that factors causing germination are known and remedies found, it is no longer a problem commercially.

### Methods of Grain Refinement

So far, the common factors which have a bearing on grain size have been discussed. Now, some of the methods that have been advocated for treating the alloys to produce fine grain will be considered. The process referred to as superheating consists of heating the molten alloy some 300 to 500 F above its melting point for a short time, then cooling back to the required pouring temperature. This method has been discussed by Beck, Gann, Menking, and others and was covered by early United States and foreign patents<sup>4, 5, 6, 7, 8, 9</sup>. Because of its simplicity and dependability, it has been widely used and is just now gradually being displaced by some of the newer inoculation methods such as carbon and carbonaceous addition agents, techniques for which were developed just prior to and during World War II. These latter techniques consist essentially in adding carbon (lampblack), carbides, and carbonaceous gases or solids to the molten metal in certain effective temperature ranges, usually between 1350 and 1550 F.

Another method of grain refining which has received attention consists in agitating the metal, as by mechanical stirring, at moderate temperatures such as 1350 to 1500 F.<sup>10, 11</sup>

The Germans have developed a treatment known as the Elfinal process wherein grain refining is obtained by plunging an inverted ladle containing anhydrous iron chloride into the molten alloy.<sup>12, 13</sup> All of these methods will be discussed in more detail.

The obvious advantage of these newer treatments is the saving in time, fuel, and more important, in the life of the melting equipment.

The literature of the last 15 years, and particularly since 1940, contains a considerable amount of good work on magnesium grain refining. It is rather unfortunate that much of this work was done under war pressure with a view to finding quick and more eco-

nomical processes of grain refining rather than to determine the basic reasons for the refining action. That this latter problem is not simple can be attested to by the author who has been associated with a group of people working on both phases of the problem for at least ten years and still the answer is not, by any means, clear. There have been evolved a number of theories, but rather than take the time to outline them at this point, they will be introduced as the experimental observations are presented. For the most part, these experimental observations have been chosen as the most significant from the work of the author and associates, but others will be presented from the literature and credited to the proper source.

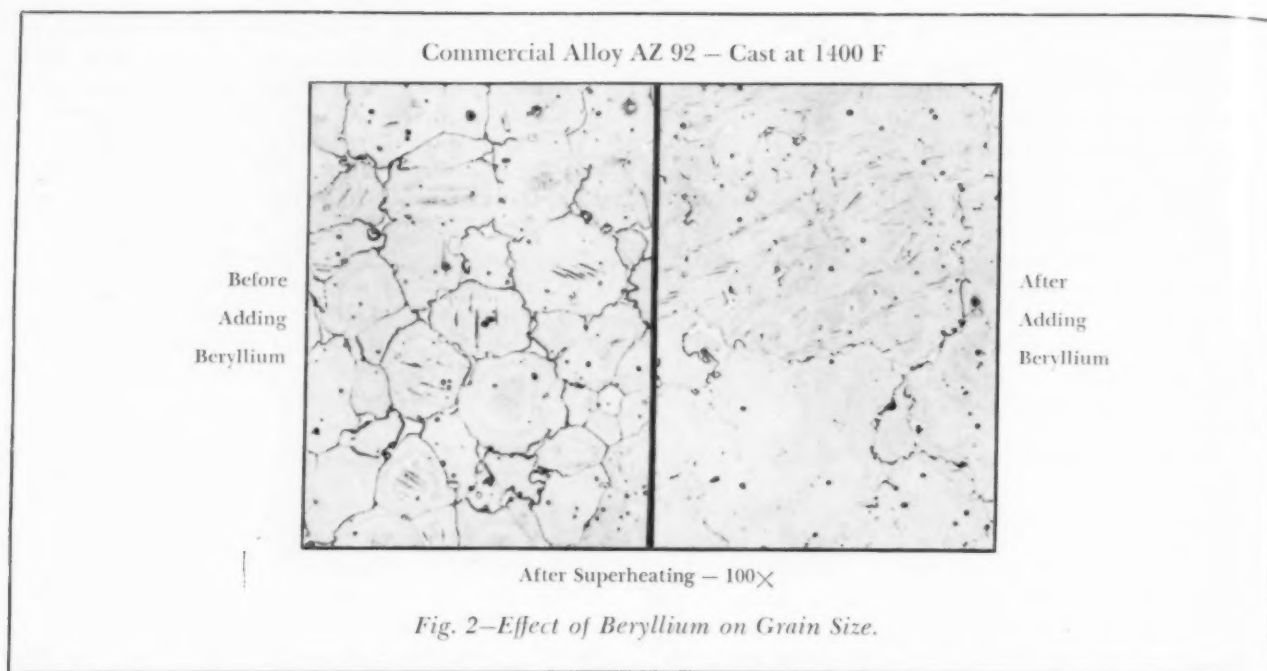
### Superheating Method of Grain Refinement

**Effect of Alloy Composition**—First, the variables of consequence in the *Superheating* method of grain refinement will be presented. Alloy composition is, of course, one of the prominent variables. It has already been pointed out that good grain refining can be obtained in the standard casting alloys which comprise Mg-8 to 12 Al-Mn and Mg-6 to 10 Al-Mn-1 to 4 Zn. Of these, the compositions with the higher aluminum contents have a natural tendency to show slightly finer grain size and are usually more rapidly refined than the lower aluminum content alloys. It has further been found that binary alloys of magnesium with zinc, silver, tin, aluminum, copper, and Mischmetal all respond to superheating to yield fine grain size.

Pure magnesium and the binary Mg-Mn alloys are difficult to investigate inasmuch as the refining effect of superheating is lost so rapidly that a sample must be poured immediately after the treatment to note the maximum refinement. It may be observed that while the grain size is refined perhaps five fold, the absolute grain size is still quite coarse. The Mg-1.2 Mn (MI) alloy with a 0.1 per cent or more calcium addition responds more readily to grain refining treatments, and retains the finer grain size long enough to allow commercial use of the process in the casting of extrusion ingots and rolling slabs. Similarly, the addition of as little as 1 per cent aluminum to the MI alloy renders the alloy even more susceptible to superheating to yield fine grain size.

Binary alloys with bismuth, chromium, and thallium do not respond to superheating.

Beryllium in amounts as low as 0.0005 per cent causes instantaneous grain coarsening in commercial alloys even though they have been given the best superheating treatment<sup>14</sup>. This is strikingly illustrated in Fig. 2 in which the left hand picture shows AZ92 alloy cast at 1400 F after superheating to 1650 F. It has an A.G.D. of 0.005 in. The right hand picture shows a sample poured only a few minutes later from the same melt after adding 0.0034 per cent beryllium. It has an A.G.D. of 0.05 in. Calcium, in amounts from 0.05 to 0.1 per cent, when added to commercial AZ63 or AZ92 alloys will produce this same type of coarsening if the sample is taken immediately after the calcium addition, but if, as would be the normal practice, the metal is not poured for five to ten minutes, the grain will be fine. The coarsening effect of beryllium, however, is



retained and will persist even after repeated remelting. This behavior of beryllium is unfortunate in view of its marked effect of reducing oxidation. It has been claimed<sup>35,50</sup> that zirconium will counteract the coarsening action of the beryllium, but our tests have indicated zirconium is only partially effective in this respect.

The effect of alloy purity, especially in the common casting alloys, on reaction to superheating is of special interest. The most widely accepted theory for explaining the superheating effect is that some elements or compounds are present in the molten alloys that are not effective as nucleating agents at the low temperatures and then reprecipitate on cooling providing many effective nuclei for crystallization and fine grain. The further evolution of this theory would explain that superheating effects are lost again upon holding the alloy at low temperatures after superheating due to a coalescence of these nuclei to an ineffective particle size or state. Credit for this theory, incidentally, should go in a large measure to the German investigators, Achenbach, Nipper, and Piwowarsky<sup>9</sup>, who have carried out important fundamental investigations on the nature of the grain refining effect of superheating on magnesium alloys.

#### Effect of Iron on Grain Size

Since iron is a common impurity in the range from 0.001 to 0.030 per cent, AZ63 and AZ92 alloys were made with <0.001 per cent iron using distilled magnesium with high purity alloying materials, and also by starting with commercial materials and using the conventional iron precipitation method. All of these alloys responded readily to the superheating treatment and gave normal fine grain. This was true even though the iron content at no time exceeded 0.001 per cent and remained <0.001 per cent in the final alloy. It was observed, however, in additional tests that the low iron

alloys could be grain refined at superheating temperatures ranging from 50 to 100 F lower (for the same treating time) than the higher iron alloys. The foregoing experiments with low iron alloys should be kept in mind in view of the theory back of the Elfinall process, namely, that iron is the element which accounts for superheating phenomena. In the high purity alloys just described, typical analyses for those made from all pure materials was as indicated in the following table:

Element	Alloy	
	AZ92	AZ63
Aluminum	8.6	6.0
Cadmium	<.01	<.01
Copper	<.01	0.014
Iron	<.001	<.001
Manganese	0.35	0.36
Nickel	<.001	<.001
Lead	<.001	0.006
Silicon	<.01	<.01
Tin	<.001	<.001
Zinc	2.0	2.7
Silver	—	0.003

Thus it can be seen that other metallic impurities, if effective, must act when present in trace amounts.

In addition to the foregoing observations, it is of interest that with few exceptions the initial grain size of the high purity alloys was fine when the alloy was made, and further, the rate of grain coarsening at low temperatures, 1200 to 1350 F, was comparatively slow. The rate and extent of grain coarsening at low temperatures after superheating was also low. On the basis of these observations one might hypothesize that the basic characteristic of these pure alloys is to have a fine grain size and that some impurity normally present in commercial alloys causes the coarsening tendency which in turn is counteracted in the course of the special refining treatments to yield a fine grain.



### Effect of Manganese

Manganese is present as a minor alloying constituent of nearly all commercial magnesium alloys and its importance with respect to grain size is becoming appreciated rapidly. N. Tiner<sup>15</sup>, in a fine published work, pointed out some of the roles that manganese may play with respect to grain refining behavior. He concludes:

1. That binary Mg-Al alloys which are susceptible to grain refining contain either iron or manganese in excess of their solubilities at the freezing point of the alloy.

2. That whereas approximately 0.1 per cent manganese is soluble at the freezing point in magnesium alloys containing 8 per cent aluminum, 0.02 per cent manganese containing alloys are not susceptible to superheating, while alloys containing 0.16 per cent manganese and over show increasing grain refining tendencies after superheating as manganese content increases.

3. That although the manganese and iron constituents within the grain structure could be recognized by microscopic examination, there appeared to be no correlation between the number of such particles and grain size.

4. That the Mg-Al alloys which can be refined by superheating show a reduction in the amount of undercooling after the superheating treatment, whereas pure magnesium and binary Mg-4Zn do not show such a consistent reduction in undercooling nor do they show refinement.

The work on undercooling by Tiner<sup>15</sup> agreed with and added to the work previously reported by Achenbach, Nipper, and Piwowarsky<sup>9</sup>, who also noted that superheating reduced the amount of undercooling. We will refer to the significance of these undercooling tests again.

Our work also shows that manganese has a particularly important although perhaps not essential connection with grain size. A significant observation is that when a Mg-Al casting alloy is treated with an excess of manganese (i.e., 1 per cent or over), as is common for settling iron from commercial alloys, it is impossible to refine the grain in that alloy during the same thermal cycle by any of the common techniques. To be more explicit, it is the usual practice when treating out iron to raise the temperature of the alloy to around 1400 to 1600 F, dissolve as much excess manganese as possible, and then let the melt cool down to a lower pouring temperature.

One might normally expect a grain refining action from this thermal treatment, but such is not the case. Neither would grain refinement be obtained if the original temperature cycle went as high as 1700 F or over. However, if the excess iron and manganese are allowed to precipitate out at lower holding temperatures, e.g., 1300 F, and are removed by sludging the pot, then it is possible to reheat the melt and superheat or grain refine by other means and get good fine grain as well as retain good purity.

The same interference of excess manganese with grain refining is noted in an experiment in which a melt of AZ92 alloy was treated with a one-third addition of molten Mg-1.2 Mn alloy. The resultant alloy, though within the composition range normally respond-

ing to grain refining, cannot now be refined until the excess manganese is precipitated out by holding at low temperatures (1200 to 1300 F) and removed from the melt by sludging out. After such treatment, the alloy can be refined by thermal superheating or inoculation methods.

The effects just described have also been observed on large scale operations in a production foundry where for a short period the manganese content was permitted, through deliberate additions, to run as high as 0.5 to 0.6 per cent in the alloy. The result was that even though good grain refining practices had been followed the casting grain size was high for refined metal (0.006 in. to 0.015 in. A.G.D.). The behavior just discussed is, of course, of more interest in understanding the grain refining mechanism than as a foundry problem since manganese content would rarely, if ever, go so high accidentally.

The need to place a lower limit on manganese in magnesium casting alloys containing over 8 per cent aluminum to prevent germination (grain growth during heat treatment), has been described in the paper by Peters, Busk, and Elliott<sup>3</sup>, and again shows the important bearing manganese has on grain size behavior in both the liquid and solid states. In the work just mentioned, it was discovered that if manganese content in the AZ92 alloy was 0.2 per cent or higher it was difficult to get germination. On the other hand, if manganese was left out of the alloy, it was almost impossible to avoid germination during the heat treatment. This latter point must be kept in mind when considering the work of Tiner and others who examined their alloys for grain size only after the homogenizing heat treatment. This may explain, for example, why Tiner did not get good superheating results on the Mg-Al alloys containing no manganese<sup>15</sup> whereas our work indicates good response to superheating on these alloys.

Returning again to the effect of manganese on germination, we have observed that AZ92 alloy with manganese content from about 0.09 to 0.15 per cent has considerable susceptibility to germination and might require the special non-germinating heat treatment to prevent it. It is this behavior that led to the short lived epidemic of coarse grained castings during the war. While the base alloys contained plenty of manganese, it was the practice in some foundries to hold the metal in pre-melters during which time the temperatures often got low enough to allow the manganese to precipitate out to values of 0.09 to 0.15 per cent. The remedy, of course, is to keep the pre-melting units at a higher temperature or to raise their temperature to 1350 to 1400 F after any low period, with stirring, so as to redissolve the manganese.

### Effect of Iron and Manganese

It has already been stated that we find that high purity magnesium casting alloys, even with <0.001 per cent iron and manganese have a natural fine grain. This may vary from 0.003 in. to 0.010 in. A.G.D. The interesting thing is that with both manganese and iron low these alloys respond sluggishly to either grain refining or coarsening treatments. Figure 3, showing a thermal treating cycle for the regular commercial purity AZ92 alloys as compared to high purity AZ92 alloys,

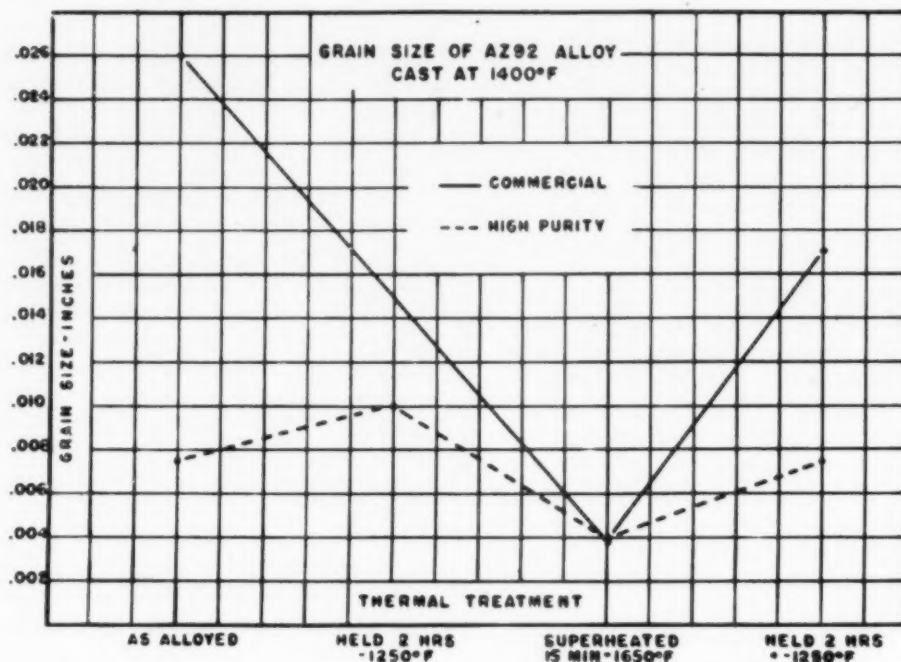


Fig. 3—Effect of Alloy Purity on Grain Size Behavior.

will illustrate this behavior. If iron is present (0.02 to 0.03 per cent) and manganese kept low, the basic grain size may be relatively coarse (0.01 to 0.03 in. A.G.D.), but the alloy responds fully and quickly to both grain refining and coarsening treatments. If the iron is kept low with manganese present (0.09 to 0.35 per cent), the basic grain size is fine (0.004 to 0.009 in. A.G.D.) and the response to grain coarsening treatments is less than with alloys of normal iron and manganese content. Due to the initial fine grain it is hard to evaluate the reaction of the high purity alloys to grain refining, but indications are that the response is definite.

#### The Elfinal Process

Another observation is that if pure AZ92 alloy with iron present but no manganese is treated with ferric chloride (Elfinal process), no refining will take place at the normal treating temperature of 1400 F even though the iron content is increased. However, the resulting alloy does respond fully to superheating and coarsening treatments. If manganese is present in this alloy (0.05 to 0.18 per cent), then the ferric chloride treatment refines the grain fully at 1400 F without any other treatment. Under the same conditions, the addition of iron as 95-5 or 90-10 Al-Fe alloy does not produce the refining effect. To complete the picture, it is well known that regular commercial alloys which contain both iron and manganese respond to both grain refining and coarsening treatments. Furthermore, in the foregoing behavior it can be stated that, except for the germination effects, AZ63 alloy reacts similarly to AZ92.

In connection with the effect of alloy composition on the superheating method of grain refining, the action of silicon cannot be overlooked. It is normally present

in commercial alloys in concentrations ranging from 0.005 up to 0.2 per cent. In binary combinations with magnesium its refining effect is as striking as that of aluminum, zinc, or calcium. A good example of this is the high purity magnesium melted in a clay-graphite crucible shown in Fig. 4. The temperature of the melt was raised slowly with samples taken for analysis and grain size at the beginning and at each range of temperature. The grain refinement, while only to a value of 0.0065 in. A.G.D., indicated that either pure magnesium (essentially free of iron and manganese) responded to superheating, or the silicon picked up from the crucible (0.02 to 0.03 per cent) at the higher temperatures caused the marked grain refinement. This metal did not coarsen appreciably on holding later at low temperatures.

A parallel experiment, Fig. 5, starting with commercial magnesium already containing 0.030 per cent iron and 0.02 per cent silicon, did not show the same grain refinement response to thermal treatment even though the initial grain size was finer.

A third experiment, duplicating the first, with high purity magnesium showing no silicon pick-up also failed to show an effect of the thermal treatment to refine the grain. A trace of aluminum was picked up with the silicon in the first experiment so two more parallel experiments were made, one with high purity and one with commercial magnesium with 0.1 per cent aluminum added to each. There was no silicon pick-up and no grain refinement in either case. The conclusions from these experiments would seem to be:

1. That the small amount of silicon produced a marked grain refinement.
2. That silicon was not effective if high iron was also

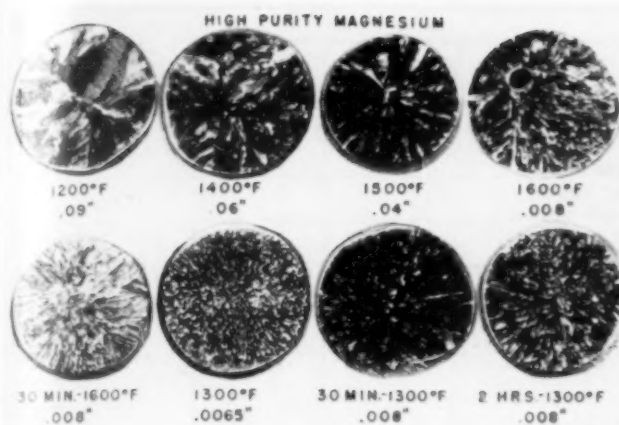


Fig. 4—Effect of Superheating High Purity Magnesium—0.02% Si.

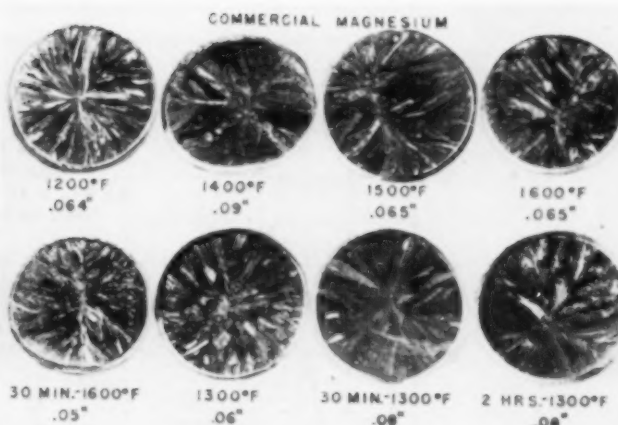


Fig. 5—Effect of Superheating Commercial Magnesium.

present in the casting alloy.

3. That under the conditions of these experiments high purity magnesium (low in iron, manganese, and silicon) did not respond to thermal treatments.

**Effect of Temperature and Time**—Effect of temperature and time on grain size has already been touched upon and has been well reported in the literature by many investigators. The work of Tiner<sup>15</sup>; Mahoney, Tarr, and LeGrand<sup>16</sup>, and the NDRC\* reports of Battelle<sup>17,18</sup> are notable. It is pretty well agreed that for the common casting alloys:

1. The grain size may be quite variable "as melted." There is a coarsening tendency in the temperature ranges from the melting point up to about 1350 to 1400 F. The rate of coarsening increases as the temperature increases within this approximate range.

2. The grain size begins to show refinement at 1450 to 1500 F and increases in rate and extent of refinement with time and temperature up to about 1700 F, at which temperature refinement is already completed. At 1500 F it might require up to an hour or two to get equivalent refining.

3. Once the grain is refined at the higher temperatures, the metal may be cooled to any lower temperature for pouring and still retain fine grain, providing the holding time at and while reaching the lower temperature is kept under control.

4. If the elapsed time at lower temperatures after superheating and before pouring is excessive, the resultant grain size tends to coarsen again in proportion to the time of holding with the rate and extent of coarsening increasing as the holding temperature is lowered. By way of example, superheated metal may be held almost indefinitely at around 1475 F or above without re-coarsening whereas it may be held only about 1/2 hr at 1400 F and perhaps only 15 min at 1300 F. Holding time, as used here, refers to the total elapsed time after cooling below temperatures of about 1500 F to the time of pouring.

5. Providing adequate temperature and time is given to attain fine grain size (<0.005 in. A.G.D.), there is no further refining effect to be gained through increas-

ing the temperature or time or repeating the refining process.

6. The grain size seeks to attain a certain equilibrium size dependent on temperature and it will therefore tend to coarsen or refine in order to reach that equilibrium size whether the metal is cooled to or heated to that temperature.

The conclusions just outlined apply *in toto* only to the thermal grain refining processes, and some of the statements would have to be modified for treatments involving inoculation with addition agents which are to be discussed later. In the latter treatments the general equilibrium behavior would still hold true but the absolute temperatures and times would be different.

One other point in connection with thermal refining is the much discussed question of "over superheating" by treating at too high a temperature<sup>19,20</sup>. Our work shows that a coarser grain size is obtained when commercial casting alloys are treated at 1800 to 1900 F and will be detected if the metal is poured immediately after this treatment. If, however, the metal cools through the regular treating ranges and is poured after the usual holding time, the grain size will be fine. This elapsed time effect is believed to be the reason why some investigators have observed the overtreatment effect while others have not. This behavior may have an important bearing on the theory of grain refining since one might estimate that if the metal is poured quickly from the exceptionally high treating temperatures sufficient unsaturation of the melt with respect to the nucleating phase exists enabling the metal to freeze before enough nuclei are precipitated to get fine grain and it would tend to behave like a pure metal.

**Effect of Metal History**—Generally, if one starts with coarse grain and fine grain ingot or scrap metal and melts and treats them by the regular thermal superheating process taking grain size samples at various stages of the process, the initially fine-grained metal will remain finer than the other until the superheating temperatures are reached after which time the melts will become equally fine. If too much time at low temperatures (1200 to 1350 F) before superheating is allowed, equilibrium will tend to equalize the grain size. This is illustrated by Fig. 6 wherein curves for grain size ver-

\* National Defense Research Committee.



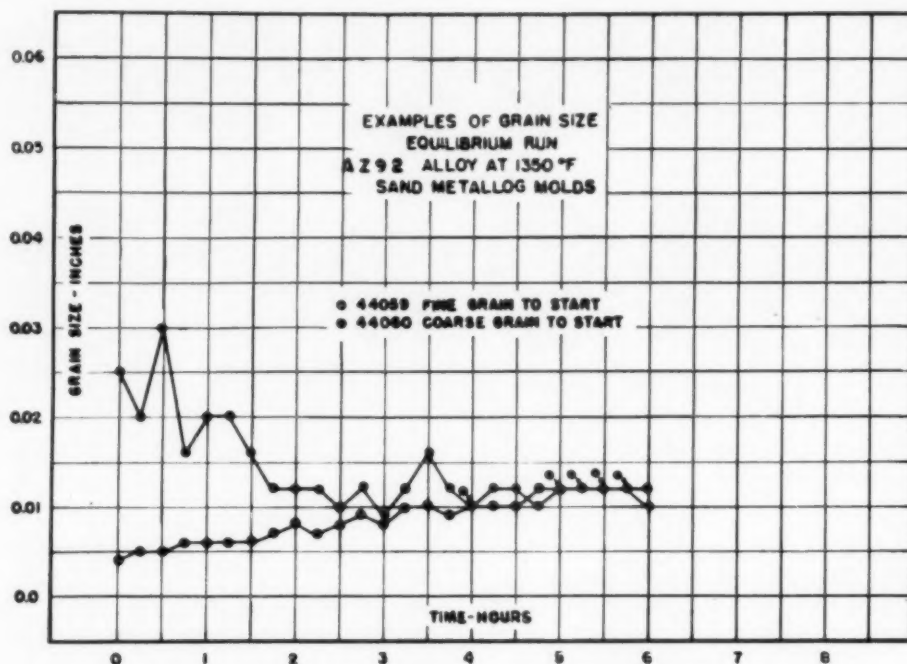


Fig. 6—Examples of Grain Size Equilibrium on AZ92 Alloy at 1350 F.

time at 1350 F are shown for alloys of coarse and fine original grain sizes. It has already been stated that there is no advantage to repeated superheating providing the first treatment was adequate or unless the refinement was lost between treatments. It has been found possible, however, to take advantage of an initially fine-grained metal by several expedients. The first and most commonly applied is simply to resuperheat the metal before casting, but limit the top temperature to that just sufficient to produce a consistent fine grain size. For AZ92 this would be 1500 to 1600 F and for AZ63 somewhat higher, 1550 to 1650 F. Since many sand foundry pouring temperatures are in these ranges anyway this involves no great heating penalty. Another expedient is to melt the fine-grained metal in such a way that the melt temperature is rapidly brought to 1400 F or above so that only a very short time is spent in the grain coarsening range (1200 to 1400 F).

#### Inoculating Coarse-Grained Metal

Experiments on "seeding" or inoculating coarse-grained metal with fine-grained metal have not led to any practical utility. When fine-grained ingot, scrap, or molten metal that has been grain refined is added to molten metal that has not been refined, an impractical proportion of the seeding metal is required. The minimum ratio of fine-grained to coarse-grained metal was of the order of one to one. A pot of metal maintained continuously at 1300 F must be seeded with its whole replacement volume of fine-grained metal every hour in order to remain fine. Obviously, the amount of seeding required depends on the holding temperature and a lower ratio would probably suffice at 1400 to 1450 F. Inasmuch as refining techniques are simple and more reliable, seeding is not to be recommended.

**Effect of Melt Size**—The size of the melt being treated has an interesting effect on the efficiency of the grain refining process. A great deal of experimental work shows conclusively that it is much easier to obtain grain refinement in small melts. This does not mean that the grain size "as melted" is different depending on the melt size. Specifically, there are a host of ways to treat small melts of 1 to 50 lb and obtain good refining, yet a large proportion of these treatments either do not work at all or are unreliable for refining melts of 500 to 2,000 lb. It can be seen, therefore, that it is important to evaluate any promising method by testing on a large scale if it is proposed to use the method for production-sized batches.

The transition of this melt size effect is gradual and apparently quite continuous up to 4,000-lb and even 20,000-lb batches. While all grain refining methods tried appear to be sensitive to this melt size effect, fortunately there are plenty of refining methods that are fully effective up to the 20,000-lb size. Superheating is one of these methods and there are many combinations of inoculating methods to be described later that are equally effective. The reason for this behavior is not clear although it has been considered that differences in the motion of the metal due to convection currents may be part of the answer.

Sufficient experimental work has been done to indicate that the effect is not simply one of time and temperature due to the bulk of metal, although, of course, this has an effect, particularly on cooling after superheating since the time required to cool to pouring temperature may be such that the grain refining effect would be lost unless artificial cooling were used or metal were ladled from the pot. It has already been stated that superheating is effective regardless of melt



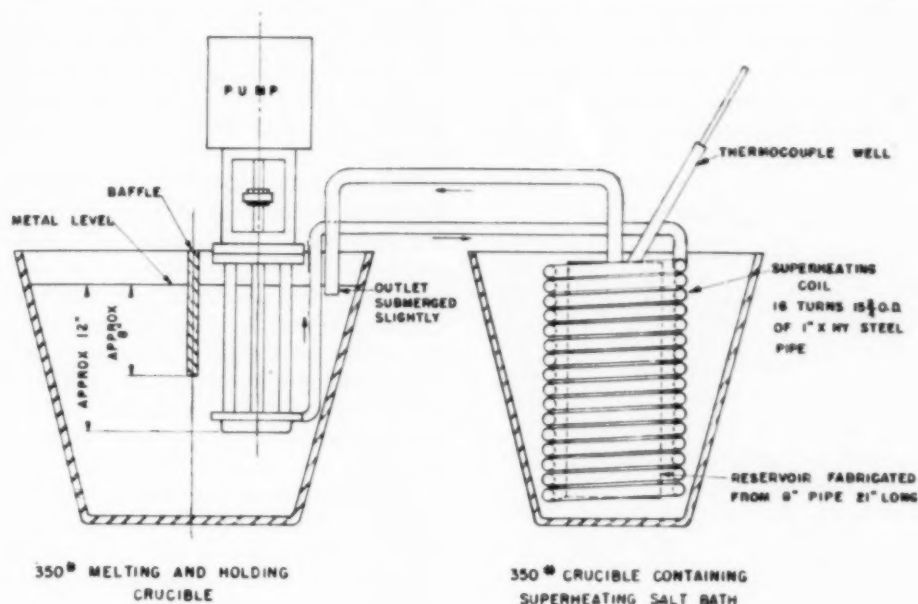


Fig. 7—Equipment Layout for Pipeline Superheating.

size, but other methods, such as chlorination alone or agitation alone, even though refining the grain in a small melt will not adequately refine the grain in a 1,000-lb melt. This is true even if excesses of chlorine or agitation or time are used.

**Effect of Gases, Atmosphere, Fluxes**—The type of atmosphere under which the metal is held has a marked effect on grain size behavior. If metal is superheated and held under dried air, carbon dioxide, argon, or chlorine, the rate of grain coarsening after superheating is greatly reduced even at temperatures as low as 1300 F. If the flux is specially dried, for example by separate treatment with calcium carbide, and dried air is held over the melt, grain refining is spontaneous at temperatures of 1400 F or over. Furthermore, the grain will remain fine on holding under the dried air atmosphere at 1400 F. The condition of a melt with respect to dissolved hydrogen does not seem to have much effect on grain size directly, or on the action of other treatments to produce good grain refinement. Our experiments included the range from metal degassed with chlorine to metal that had been over-treated with hydrogen to the point of sponginess. This agrees with the results reported by Tiner<sup>15</sup> and others.

In order to determine further whether magnesium alloys could be superheated out of contact with air, several types of experiment were carried out. The first experiment consisted of completely filling a special closeable ladle with molten metal by immersing the ladle in a crucible full of molten AZ63 alloy at 1250 F. The ladle was then sealed. After sampling the crucible metal for initial grain size, the entire crucible of metal containing the submerged sealed ladle was given a superheating treatment of 15 min at 1650 F. After cooling to 1400 F, grain size samples were poured from the metal in the open crucible and also from the sealed ladle. Both gave uniformly and equally fine grain sizes

of 0.004 in. A.G.D. A control experiment, the same as the foregoing, but without superheating, gave coarse grain in both samples.

A great many experiments were made on a method referred to as pipeline superheating. In this set up, Fig. 7, molten alloy (AZ63) at 1250 to 1300 F was continuously circulated with a centrifugal pump from one pot through a closed-pipe coil immersed in a second pot of metal held at 1650 F and then returned to the original pot. In experiments at various temperatures and rates of circulation it was determined that at low speeds of circulation, 10 lb/min, good grain refining was obtained when temperatures in the superheating coil were around 1600 F or above. At higher rates of circulation, grain refinement was not attained presumably because the metal was not in the hot superheating coil long enough. (About 7 min was required.) That this was actually the case was shown by attaching a sealed reservoir capable of holding about 40 lb of metal in series with the superheating coil and immersed in the same pot. Through the use of this inventory in the hot zone it was possible to get grain refinement at higher speeds of circulation.

The practical implication of these latter experiments is that a pipeline superheating unit can thus be operated so as to take a molten metal supply at low temperature and pump a superheated and re-cooled metal at predetermined pouring temperature direct to molds. The experiments are important technically in that they demonstrate that metal can be grain refined by superheating in a closed system away from atmosphere or flux. Separate pot experiments indicate that it is easier to superheat metal when flux is absent and that the grain does not recoarsen as rapidly at low temperatures.

Very good experiments reported by Tiner<sup>15</sup> on superheating in vacuum add to the strength of the conclu-

sion that air, atmosphere, or flux are not necessary to the grain refining action. These experiments have a strong bearing on the so-called oxide theory which hypothesizes that the superheating effect may be due to increased reactivity between metal and atmospheres at the high temperature with resultant formation of fine oxide or non-metallic nuclei.

### Carbonaceous Additions for Grain Refining

Carbon or carbonaceous addition methods, because of their effectiveness at lower temperatures than those required for superheating, show great promise of becoming the predominant treatments of the future<sup>21</sup>. It is possible that their adoption has actually been retarded by the fact that there are so many variations of the method and materials which may be used that no single agent has received sufficiently widespread usage to get the practical background of experience necessary to make a method popular. This is a temporary situation, however, and eventually several of the methods will be developed to fully evaluated practical treatments by industry itself.

It seems fitting to discuss these methods in the chronological order in which they appeared in the field. Our first observation of grain refining by carbonaceous action was in the course of experimental melting of magnesium alloys (AZ63) in a graphite resistor type Detroit rocking furnace in 1940. It was observed that grain sizes were 0.002 in. to 0.004 in. with excellent mechanical properties even though the temperature of the molten bath never exceeded 1350 to 1400 F. Actually, superheating this metal in the same furnace at 1650 F did not produce further refinement.

Experimental work was started at once to determine whether the refinement obtained was due, first, to the carbon monoxide and carbon dioxide which by analysis were shown to exist in the atmosphere over the metal; second, to the particles of carbon dust that were dropping off the oxidizing electrodes; third, to the agitation caused by the rocking of the furnace; or fourth, to the local superheating of small portions of the melt that were exposed to direct radiation or to contact with the hot furnace walls. The individual experiments that were set up to delineate these effects showed that within certain limits each of the above agents would produce grain refinement.

### Effect of CO and CO<sub>2</sub> Atmospheres

Atmospheres of carbon monoxide or carbon dioxide over a pot of metal would give refinement of small melts but it was found necessary actually to bubble one of these gases through the melt to get good refinement of larger melts<sup>22</sup>. Temperatures of about 1400 F are necessary for good results. An unusual behavior was also noted at this time in that if too much of the gas were passed through the metal the mechanical properties of the cast metal would drop. Observations indicated that the metal became more viscous even though the grain size remained fine. The exact cause of this action has not yet been determined. Experiments with other carbon containing gases and vapors such as acetylene, methane, natural gas, butylene, carbon tetrachlor-

ide, carbon disulphide, alcohol, and many others were carried out and all gave good grain refining effects on small melts. Some of the more promising ones, including carbon dioxide and natural gas, were tried alone and in combination with agitation in melts up to 4,000 lb with good grain refinement at temperatures of 1400 F or over.

Attempts were made in the early years of World War II to put these processes into the production foundry, but because of the difficulty of treating each of the small superheating crucibles and the fact that the larger pre-melting crucibles and permanent mold casting pots operated at too low a temperature (1250 to 1350 F) to get refinement the processes were held back. A scheme for degassing with chlorine and grain refining with carbon dioxide simultaneously also failed to do the job adequately on the larger pots operating at low temperatures (less than 1400 F).

Paralleling this work was the testing of solid carbon additions in the form of graphite, carbon, and lampblack. Grain refining in small melts was obtained at 1400 F and over by adding the powdered materials or lumps of carbon or graphite. The most reproducible results were with lampblack. Some difficulty, as has been mentioned by others, was encountered in getting the lampblack in contact with the metal and consequently more attention was given to the gaseous or vaporizable carbon compounds previously described, and also to solid addition agents such as calcium carbide, silicon carbide, inorganic carbonates, magnesium carbonate, zinc carbonate, sodium bicarbonate, etc., (evolving carbon dioxide), hexachlorethane, and hexachlorobenzene. Of these, calcium carbide and hexachlorethane appeared to be the most dependable and safe. As has been discussed previously, the calcium carbide not only refined the grain at temperatures as low as 1350 to 1400 F, but also dried the flux producing a quiet melt that retained fine grain size. This works well in 600 to 2,000-lb pots at 1400 F or over. Additions of about 1 to 2 per cent calcium carbide are required. All of the agents mentioned above gave some refinement. The hexachlorobenzene, while good for refining, reacts violently and would be too dangerous for commercial use.

The third general outgrowth of the rocking furnace observations was the work on agitation which was done simultaneously with the developments described but will be left to later discussion.

At about this time great activity began in the whole field of magnesium grain refining due to the pressure of war production. The excellent work on NDRC projects<sup>23</sup>, particularly those at Battelle<sup>17,18</sup>, and the independent work of Mahoney, Tarr, and LeGrand<sup>19</sup> began to bring to the foreground the importance of carbon additions and the possibilities of more reliable combination treatments for degassing and grain refining simultaneously. The chlorine-carbon tetrachloride treatment developed by Battelle, which consists in entraining carbon tetrachloride vapors in the chlorine gas stream and bubbling the mixed gases into the molten metal at about 1350 to 1400 F, has been found quite effective for both degassing and grain refining. Our work shows that this treatment will give grain refine-

ment on melts as large as 20,000 lb. The treatment also is of interest because it will begin to refine at temperatures as low as 1350 F whereas most of the carbonaceous additions, except perhaps calcium carbide, do not become very effective until 1400 F or above.

With the advent of carbon reduced magnesium it was found that alloys made with this metal appeared to refine somewhat easier than previous commercial alloys. The supposition was that the effect was due to residual carbon carry-over in the metal and led to the independent discovery of the effect of carbon additions by Hultgren and Mitchell<sup>11</sup>; Mahoney, Tarr, and LeGrand<sup>12</sup>; and others. While the carbon may have played a major part, the question may now be raised as to what extent this ease of grain refining may have been due to a higher purity in the alloys as indicated by our previously discussed results on controlled purity alloys.

Krynitsky and Holm<sup>24</sup> have recently reported grain refining of magnesium alloys by the addition of lumps of uncalcined magnesite to the molten metal. They point out that lumps are necessary since finer powders are hard to bring in contact with the metal due to the refining action of the flux. It has been assumed that the magnesite refines due to the fact that carbon dioxide is evolved during the calcination which takes place in contact with the hot metal.

Our further work on carbon inoculation has all been in the direction of a process that will work dependably on large melts and at as low temperatures as possible. It would appear that in this field the best possibilities would be lampblack, hexachlorethane, which, incidentally, are claimed to be used with success commercially by Canadian and English magnesium foundries, calcium carbide, and chlorine-carbon tetrachloride. The latter has a disadvantage from the standpoint of re-

quiring good control to insure pick up of the proper amount of carbon tetrachloride in the chlorine. Our findings, though not backed by production foundry experience to date, would indicate that chlorine-natural gas or methane mixtures would work equally well and be easier to handle. This has worked satisfactorily on melts up to 20,000 lb at 1350 to 1400 F.

The temperatures required for grain refining by carbonaceous additions have been discussed. It will only be necessary to discuss differences as compared to the thermal refining method since, in general, the same overall considerations, such as effect of melt size, chill effects in casting, alloy composition and purity, effect of atmospheres, fluxes, etc., apply to the carbon treatments.

Perhaps the outstanding difference is the lower temperatures at which the carbonaceous treatments operate. The same type of equilibrium grain size with respect to temperature seems to hold in that the grain size of carbon refined metal also tends to recoarsen on holding at low temperatures but the critical temperature limits appear to be displaced downward to some extent. Further, the rate of coarsening of carbon refined metal at 1350 to 1400 F seems to be slower than with superheated metal at the same temperature. Similarly, the carry-over effect of fine grain size on carbon-treated metal seems to be more significant. This last behavior is of considerable interest since it apparently leads to what may be called "over treatment". If, as the grain size is observed, metal is repeatedly treated with carbonaceous agents, it can be shown that at first grain size gets fine and after extended treatment it coarsens. The same effect can be obtained if carbon refined metal is superheated.

Hence it is important to use and adhere to only one refining method and not try to play safe by giving the

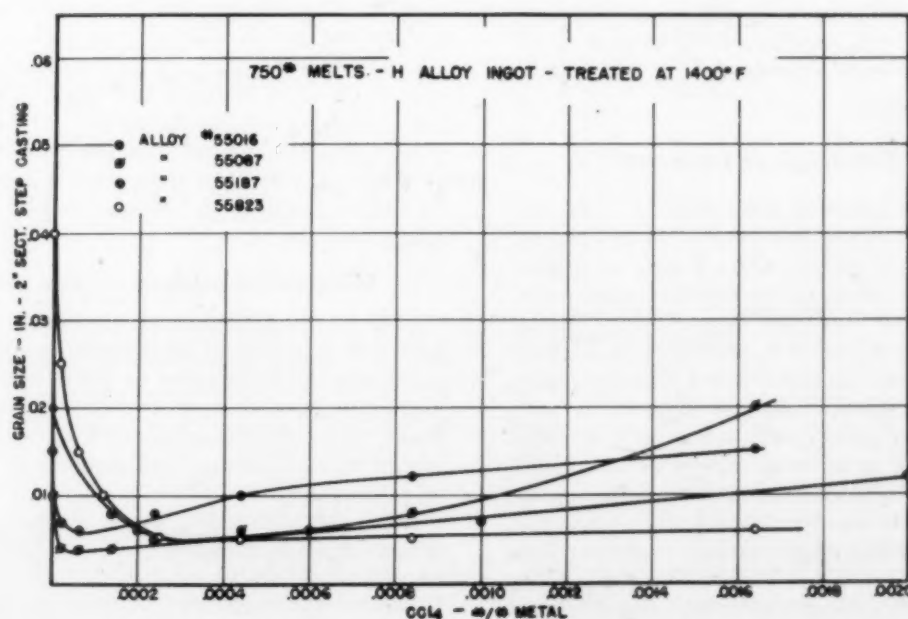


Fig. 8—Effect on Grain Size of Varying Amounts of  $Cl_2-CC_4$  Treatment.



combination treatment. This effect of "over treatment" is of consequence only if castings of large cross section, 2 in. to 4 in., are being made. Castings with small sections will be fine grained and do not show the overtreating effect unless there is a hot zone through which an exceptional amount of the molten metal passes during the filling of the mold. Figure 8 shows the grain size curve for 2-in. sections cast from AZ63 alloy that has been given varying stages of carbon treatment by the chlorine-carbon tetrachloride method at 1400 F. Note that the alloys with coarse initial grain size refined rapidly and then as carbon was added up to about 0.0005 per cent coarsening began. The initial fine-grained alloys began to coarsen at even lower amounts of carbon. Under similar test conditions AZ92 alloy began to show the coarsening effect at about 0.0015 lb of carbon tetrachloride per pound of metal.

This cumulative effect of grain refining has never been observed when only superheating is used even though the process is repeated many times. This overtreating effect is not believed to be cause for alarm since, in general, broad ranges for carbon treatment appear to be available before one gets the cumulative effect. The only uncertain point is to know the pretreatment history of the remelted or scrap metal so as to know how to prescribe an adequate future processing. It would appear that this point will have to be worked out by trial and error in each foundry to find the correct practice in view of the type of metal they are processing and their particular type of casting and melting practice.

It should be stated that the apparent advantages of carbon treating outweigh the points just mentioned and a foundry should find it worthwhile to set up a process suitable to take advantage of the method. For the chlorine-carbon tetrachloride treatment it has been found in our work that 0.02 to 0.1 lb of carbon tetrachloride per 100 lb of metal will usually give satisfactory results in normal processing and in making castings up to 2-in. cross section. This amount is in the same range as that recommended in the Battelle work<sup>17, 18</sup>.

#### Agitation Method of Grain Refinement

The last major method of grain refining to be discussed is agitation. While this process when properly done is satisfactory for smaller melts, it does not appear from our work to be adequate by itself for larger melts. In combination with carbonaceous treatments, special atmospheres, or gas bubbling it does have possibilities for certain operations. In order to get effective results by agitation it is important that the agitation be such as to provide rather violent local action such as could be obtained by high speed small stirrers or better still, an apparatus that will shear the metal rapidly and repeatedly. Large, slow moving stirrers may actually give grain coarsening rather than refining, except at high temperatures.

In summary of the action of the high speed type of agitation, it appears that, in general, agitation acts by permitting normal grain size equilibrium with respect to temperature to be reached much more rapidly. More specifically, agitation will cause grain coarsening more

rapidly at low temperatures, that is, below 1400 F, but will give rapid refining at temperatures of 1450 F or above which would take a long time if temperature alone were acting. One might then say that agitation makes superheating at 1450 F practical. Similarly, agitation in combination with other partially effective carbonaceous additions like, for example, carbon dioxide, or methane gas atmospheres, or bubbling makes this a treatment that is effective even on large melts. As has previously been indicated, these treatments may be effective even down to 1350 to 1400 F as found by Hultgren et al.<sup>11</sup> in their work on magnesium alloys made with carbon reduced magnesium.

There has been some mention in the literature of the possibility that cybotactic groups or liquid crystals<sup>9, 12</sup> might exist in magnesium alloys which might predetermine a coarse grain size on freezing unless these liquid crystals were broken up by thermal or mechanical treatment such as agitation. In view of all our other knowledge on the grain size behavior, it seems more likely that agitation is merely speeding up an unfavorable coalescing action of nuclei at low temperatures leading to grain coarsening and increasing the rate of solution or transformation of the nuclei forming phase at the higher temperatures much like you would stir sugar into your coffee.

#### Elfinal Method of Grain Refinement

The only other method that needs more discussion is the previously mentioned Elfinal process developed and used extensively in Germany in the last three or four years. The method consists in the addition of about 1 per cent anhydrous ferric chloride to the molten alloy at a temperature of 1500 to 1550 F. The ferric chloride must be anhydrous to avoid explosion and is plunged under the surface of the metal by holding it in a perforated iron tube or ladle. After the treatment, the metal temperature is dropped to 1350 to 1450 F and it is claimed that the metal can be held at this temperature for an hour without grain coarsening.

Our experience with this method is limited. However, it has been checked and found to work satisfactorily on standard casting alloys. As has been stated previously, this method did not give refinement on alloys containing no manganese although both superheating and carbon additions will refine the grain with manganese absent.

#### Miscellaneous Methods of Grain Refinement

Chlorine bubbled through the metal at 1400 F will refine the grain size of small melts but by itself will not give adequate refinement to melts over about 50 lb. A chlorine atmosphere over molten metal and flux appeared markedly to increase the length of time superheated metal could be held at 1300 to 1400 F before recoarsening.

Other gases, including argon, helium, hydrogen, hydrogen chloride, nitrogen, oxygen, and dried air, were tested by bubbling through 60 to 130-lb melts at 1400 F without significant or consistent grain refining. Nitrogen, helium, or argon can be used in place of chlorine as the carriers for the carbon tetrachloride vapor and give good grain refining, but chlorine is a more rapid degasser and gives clean metal.



Boron chloride, potassium chlorate, and manganese oxide all give erratic grain size results and in some cases react vigorously with the metal so are not of much interest.

Several special types of experiment were carried out to study the phenomena of superheating and are of considerable interest. Leading to the experiments was the observation that in a number of cases molten alloys at relatively low temperature appeared to show good grain refinement when a part of the same melt was simultaneously held at grain refining temperatures.

In order to extend the foregoing principle further and to develop a production method of providing a continuous source of grain refined metal for pouring at temperatures as low as 1250 to 1300 F the so-called "gradient experiments" were made. Figure 9 shows a schematic drawing of a hearth type furnace with about a 1200-lb metal capacity. The particular furnace used was electrically heated with radiant carbon resistors. It provided a closed atmosphere over the metal and an open metal dipping well at one end. Metal could be charged into the furnace proper through a door on the end opposite the well or directly into the dipping well. In the course of the experiments the temperatures were varied on both the well metal and the metal in the furnace. Grain size samples and production castings were poured under each set of conditions. AZ63 and AZ92 alloys were tested.

It was found that fine grain was obtained in the test pieces and the castings when the well temperature was held at any point above 1225 to 1250 F and the metal in the furnace was above about 1470 F. Well tempera-

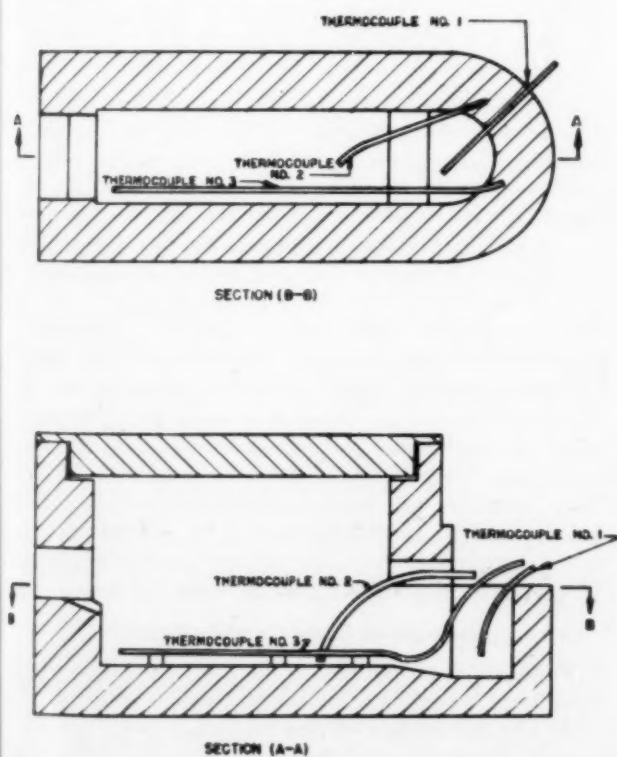
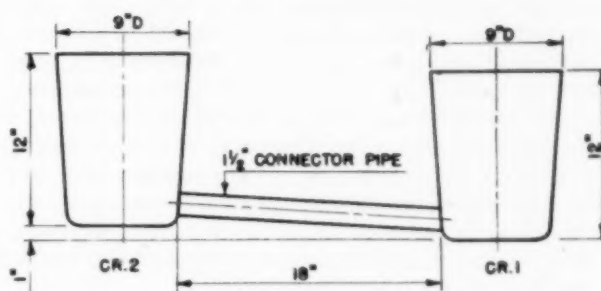
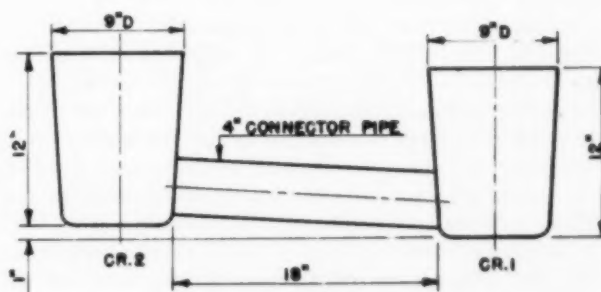


Fig. 9—Schematic Diagram of Hearth Type Furnace.

## DOUBLE CRUCIBLE ARRANGEMENTS



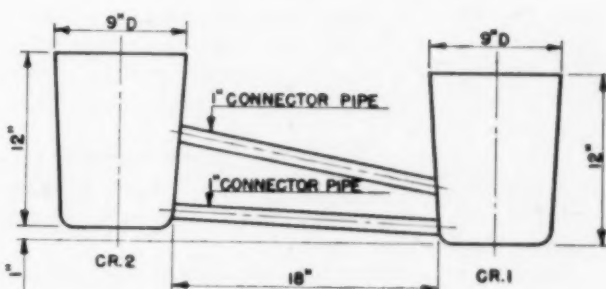
DOUBLE-CRUCIBLE A



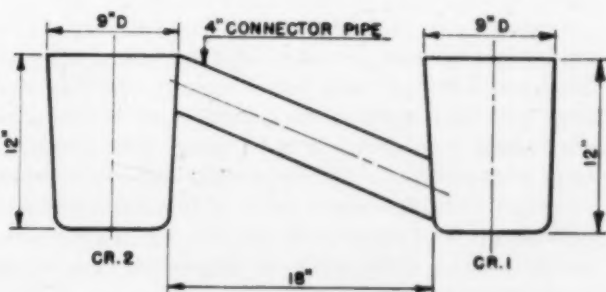
DOUBLE CRUCIBLE B

Fig. 10—Sketch of Double Crucible "Diffusion" Equipment; A and B Arrangement.

## DOUBLE CRUCIBLE ARRANGEMENTS



DOUBLE CRUCIBLE C



DOUBLE CRUCIBLE E

Fig. 11—Sketch of Double Crucible "Diffusion" Equipment; C and E Arrangement.

tures or furnace temperatures substantially below these values gave coarse grain. Higher well and furnace temperatures also gave excellent grain refinement. Good results were obtained whether the added metal during continuous operation was charged into the well or the furnace proper. It was noted that if the temperatures were on the low side, better results were obtained by charging directly into the furnace. The grain refining took place as soon as the correct temperatures were reached. Castings could be poured continuously with new coarse-grained metal also being added continuously at least up to the melting capacity of the unit which was about 200 to 300 lb per hour.

In an effort to isolate more completely the factors that might be causing the grain refining in the above furnace experiment some other tests were made using what is called the "diffusion equipment" illustrated in Figs. 10, 11, and 12. The apparatus consisted of two 20-lb capacity steel crucibles inter-connected with a section of pipe such that when the crucibles were filled with molten metal the level would be the same in each crucible, but the only molten metal connection was through the pipe. Several sizes of connecting pipe and different hook-ups were tried. The crucibles were filled with molten metal at about 1250 F. Initial grain size samples were taken and then one of the crucibles was heated to 1550 to 1700 F to get the superheating effect while the other crucible was maintained at temperatures of 1250 to 1300 F.

It was found that with small interconnecting pipes of 1½-in. diameter good grain refining was not obtained, but with pipe connections of 2½ to 4-in. diameter good grain refining was obtained in the 1300 F crucible as long as the other crucible was at temperatures of 1550 F or over. This refining took place regardless of which crucible was at the high temperature. The refining also took place without regard to whether the connection between the crucibles was favorable for convection currents.

When the ratio of flux in the crucible was made purposely high so the only connection between the two crucibles was through flux, no refining took place in the low temperature crucible.

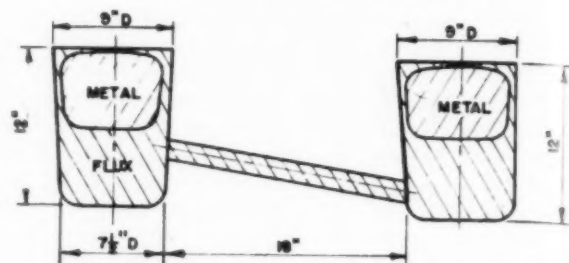
When an atmosphere of carbon dioxide was held over the high temperature crucible and metal contact made between the crucibles it was found possible to get complete grain refinement with the hot crucible as low as 1400 F and the low temperature crucible at 1300 F.

Evidence that diffusion could take place rapidly in the equipment was provided by experiments wherein 0.002 and 0.005 per cent beryllium was added in separate tests to only one of the crucibles under conditions where both were being held at comparable temperatures near 1300 F. Within several minutes samples from each crucible showed identical beryllium contents.

Evidence that convection was not the main reason for the transfer of the grain refining action depends on four considerations:

1. The grain refining obtained seemed independent of whether the mechanical set-up was favorable or unfavorable for convection.
2. Temperature control and measurement on the two crucibles did not indicate any appreciable cooling effect

## DOUBLE CRUCIBLE ARRANGEMENTS



SKETCH SHOWING PROBABLE ARRANGEMENT OF FLUX AND METAL IN RUNS 3A, 6A & 20B.

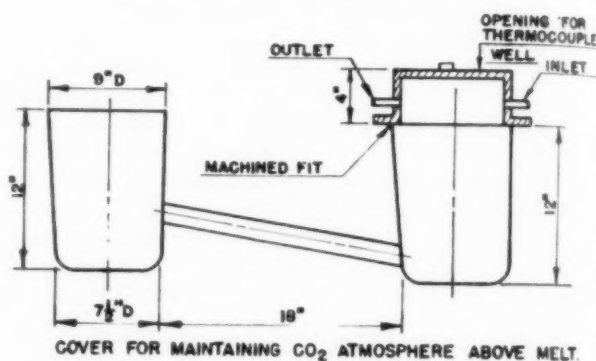


Fig. 12—Sketch of Double Crucible "Diffusion" Equipment Molten Flux Interconnection (Top), CO<sub>2</sub> Atmosphere Equipment (Bottom).

on the hot crucible or heating on the cold crucible that would be associated with any considerable metal interchange.

3. If convection and temperature were the dominant factors then the flux should have been susceptible to the same interchange and should have carried over to the metal in the cases where flux separated the two crucibles.

4. Separate "seeding" tests carried out on two unconnected crucibles of similar size showed that coarse-grained metal at 1300 F must be displaced with its entire volume of high temperature fine-grained metal every hour in order to show continuously a fine grain.

The foregoing experiments and particularly the gradient furnace are believed to have practical value where foundry metal requirements fit in with this kind of set-up. They also have theoretical value in that in these cases we are transmitting a thermal or carbon inoculation grain refining action to moderately cold metal simply by some interchange process when the proper thermal gradients exist in the same melt.

## Summary of Metal Treating Results

Following is a summary of the significant facts which have been brought out by the discussions thus far:

1. The advantages of fine grain size are:
  - a. Improved mechanical properties.
  - b. Reduced "hot cracking" in the foundry.
  - c. A fine-grained alloy is easier to solution heat treat.

The effect of alloying elements is as follows:

- a. An alloy content of 7 per cent or over is more favorable to grain refining.
  - b. Binary alloys of magnesium with zinc, silver, tin, copper, cerium, and aluminum are susceptible to refinement by superheating.
  - c. Binary alloys with bismuth, thallium, chromium, and calcium either coarsened upon superheating or were not affected.
  - d. Beryllium causes a marked grain coarsening that is even further increased by superheating.
  - e. Excesses of manganese in common casting alloys may interfere with grain refining treatments unless the excesses are allowed to precipitate and are removed from the pot. Manganese contents below the normal specification range may permit grain growth during heat treatment (germination) in alloys containing over 8 per cent aluminum.
3. Grain refining is obtained in magnesium alloys by:
- a. Superheating at 300 to 500 F above the melting point of the alloy.
  - b. Treating the metal with carbon or carbonaceous solids, liquids, or gases.
  - c. Violent agitation at 1400 F or above, especially in combination with carbonaceous treatments.
  - d. Treatment with ferric chloride at 1400 F (El-final). This treatment is effective by itself only if manganese is present.
4. Holding at low temperatures, 1200 to 1400 F, tends to coarsen the grain especially after a refining treatment.
5. For each temperature there appears to be an equilibrium grain size which will be reached if sufficient time is given regardless of whether the melt approached from either a higher or lower temperature.
6. There is no cumulative effect of grain refining treatments to give additional refining once a grain size of the order of 0.004 in. has been obtained. There does appear to be a cumulative coarsening effect from using too much of the carbonaceous addition agents if castings of large section size are being made. An "overheating" grain coarsening effect can be detected if the metal is superheated at 1800 to 1900 F and samples are poured immediately from this temperature. If, however, the metal subsequently cools at a normal rate through the regular superheating range the grain will be fine as usual.
7. Grain refining by thermal treatment can take place under vacuum, in a closed pipeline or ladle out of contact with atmosphere or flux.
8. Grain refining can be induced into metal at as low as 1250 F by molten metal contact with a portion of the melt which is at 1470 F or above.
9. Metal can be grain refined at low temperatures by "seeding" it with large quantities (1:1) of fine-grained metal but the process is expensive and critical.
10. It is much easier to obtain grain refinement in small pots than in large ones.
11. Grain refinement is more easily attained and retained under dry flux and dry atmospheres or where flux is absent.

## 12. Common magnesium casting alloys:

- a. If low in iron, appear to have a native grain size considerably finer than high iron alloys. They do not coarsen at as high a rate as alloys of normal iron and manganese content on holding at low temperatures. These alloys respond to superheating and carbon treatments but the net grain size change is smaller due to their initial fine grain.
  - b. When iron is present but no manganese the native grain size is moderately coarse but the alloy can be refined by superheating.
  - c. When neither iron nor manganese are present, the alloy has a native medium-fine grain but is extremely sluggish in response to either grain refining or coarsening. One might conclude from this behavior that either manganese or iron must be present in order to get response to these treatments.
13. Gases such as hydrogen, helium, argon, nitrogen, etc., by themselves do not give consistent grain refining action.
14. Additions of magnesium oxide or aluminum oxide<sup>25</sup> directly, through burning on the surface of the metal, through the use of fluxes like 310 (containing 15 per cent magnesium oxide), or by treating the melt with gaseous oxygen do not produce definite consistent grain refining.
15. Work on undercooling reported in the literature indicates that it is sharply reduced on superheated alloys and that the amount of this difference increases as alloy content increases<sup>9</sup>.

## Discussion

The most interesting experiments in connection with the methods of grain refining have now been covered and the significance of the results will be discussed with a view to developing a better understanding of the mechanism of the grain refinement. It will not be possible to present adequately the fine discussions given in the papers by Tiner<sup>15</sup>, Mahoney, Tarr, and LeGrand<sup>16</sup>, Achenbach, Nipper, and Piwowarsky<sup>9</sup>, and others but many of the thoughts presented will be drawn from those references.

It seems to be a natural tendency, in seeking a theory to explain a number of related phenomena to attempt to find one single mechanism which will fit all the experimental facts. That this is a good approach cannot be denied but if followed too far may lead to oversimplification and failure to recognize the possibility that several inter-related mechanisms offer a more satisfactory explanation. With these points in mind, let us consider in turn several of the most common hypotheses that have been advanced to explain the grain refining behavior.

## Liquid Crystal Theory

It has been found<sup>9,15</sup> that certain solid substances carry over into their molten state a semblance of the crystal structure characteristic of the solid material. These are sometimes referred to as "liquid crystals". If one were to assume that the grain size of magnesium alloys were controlled by such crystals, then one would



expect them to be broken up or dissolved in the process of superheating thus eliminating them as nuclei for crystallization. If such were the case, greater undercooling would take place on freezing, and as pointed out by Tiner<sup>15</sup>, and Achenbach, Nipper, and Piwowarsky<sup>9</sup>, the exact opposite result is obtained in that the amount of undercooling is inversely proportional to the degree of superheating treatment. Further, the fact that a melt may be relatively fine grained as melted but coarsens gradually as time and temperature increase up to a range between 1400 and 1500 F and then begins to refine again with further increases in temperature, does not fit in with the liquid crystal idea.

### Magnesium Oxide Theory

The hypothesis has been advanced that grain refinement may be due to the formation of clouds of magnesium oxide or similar non-metallic inclusions that act as nuclei for crystallization.

It is certainly true that magnesium oxide, aluminum oxide, or nitrides of magnesium or aluminum are almost certain to be introduced into magnesium alloys as they are melted, superheated, and poured. One could also visualize the increased formation of such compounds at the higher temperatures of superheating.

While the possibility of some nucleating or "blocking" action from these constituents cannot be wholly denied, the following points appear to indicate that their effect is minor:

1. Direct additions of magnesium oxide, aluminum oxide, oxygen, nitrogen, etc., to the melt do not produce grain refinement nor does burning of the metal on the pot surface.
2. The grain refining effect can be obtained in a vacuum and in a closed system away from atmospheres or fluxes.
3. It is not easy to see the correlation of this type action with the carbonaceous treating agents.
4. The strong relationship that has been demonstrated between metallic constituents such as iron and manganese and the grain refining phenomena does not tie in with the oxide theory.

### Surface Tension Hypothesis

An interesting hypothesis has been advanced by Bastien<sup>20</sup> to explain the considerable grain refining effect of adding 1 to 10 per cent magnesium to aluminum and conversely when adding similar quantities of aluminum to magnesium. He states that there is an excellent correlation with the surface tension in that it also decreases as the second metal is added. No data is given with respect to definite grain refining practices.

It is interesting from our experience to note that 1 per cent zirconium additions to magnesium produce fine grain sizes and also appear to reduce markedly the surface tension of the alloy. This entire field of viscosity and surface tension as a function of alloys and treatments should receive more study.

### Temperature-Solubility Nucleation Theory

The most widely accepted theory to date concerning the general mechanism of grain refinement was first presented, as far as we know, by Achenbach, Nipper,

and Piwowarsky<sup>9</sup>. In its simplest form it suggests that some material which at normal temperatures is too large in particle size to be effective as nuclei is taken into solution at the higher temperatures and then reprecipitates to form fine nuclei during the cooling process. The fact that after superheating the grain refining effect deteriorates on long standing at temperatures below the superheating range would lend credence to the idea that these nuclei are again coalescing and growing to their original form and relative ineffectiveness. This general theory, while it has some weaknesses that will be discussed later, goes far toward offering a mechanism which will fit the major part of the observed results. Investigations carried out in the last seven or eight years bring out more specifically some of the materials causing grain refining that could fit into the above theory and it should be worth while to examine the case for each of these materials, namely, carbon, iron, and manganese.

### The Carbon Hypothesis

There is adequate evidence from both experimental and production use that carbon and carbonaceous solids and gases can be used as inoculants to produce fine grain in magnesium casting alloys. The significant difference in behavior between the carbon refining methods and thermal superheating seems to be in the lower temperature at which fine grain is obtained with the carbon treatments. In other respects, the grain size seems to be dependent on temperature in the same general way as thermally refined alloys in that grain coarsening occurs if the alloy is held for appreciable time at temperatures below 1400 F either before or after the refining treatment.

The paper by Mahoney, Tarr, and LeGrand<sup>16</sup> concluded that it was probable that the carbon content, even of electrolytically produced magnesium was basically responsible for grain refining regardless of whether thermal or other methods were used. The lower temperature of refining in the case of deliberate carbonaceous additions was then explained on the basis that more carbon was available. No complete explanation was offered as to how the carbon was functioning in the refining mechanism other than the supposition that the carbon or some carbon compound had a temperature dependent solubility which permitted precipitation of cloud-like nuclei after the melt was cast following exposure to temperatures above 1400 F.

The results of the work at Battelle Memorial Institute led to the hypothesis<sup>18</sup> that aluminum carbide,  $Al_4C_3$ , was the compound responsible for the carbon refining effect largely on the basis that both aluminum and carbon had to be present in order to make such treatments effective. No conclusive experimental data has been found in either their work or ours to substantiate this supposition since direct additions of this material gave erratic grain refining results. This does not preclude the possibility that aluminum carbide is important since the particle size or manner in which it is formed may be significant. In connection with the aluminum carbide hypothesis, Mahoney, Tarr, and LeGrand<sup>16</sup> also pointed out that there is no indication in the literature<sup>27</sup> of instability of aluminum carbide

the entire temperature range of carbon additions to magnesium. They found no evidence of the compound in the cast alloys after carbon treatment. Our work has indicated:

1. That thermal superheating, ferric chloride additions, or carbonaceous additions are ineffective for grain refining unless either iron or manganese are present.

2. That casting alloys made with distilled magnesium (low in carbon) can be refined by superheating if iron or manganese or both are present.

In summary of the situation with respect to carbon, it would be concluded that like iron and manganese, it is an important but not an essential factor in grain refining mechanisms.

### The Iron Hypothesis

That iron plays an important role in the grain size behavior of magnesium alloys has long been suspected and for good reason since iron is almost always present in commercial alloys and has a temperature dependent solubility in these alloys. It may be observed that the iron solubility drops markedly with the addition of either aluminum or manganese thus tending to make any precipitation reaction more likely and more critical. It is further known that Al-Fe, Fe-Mn, and Al-Fe-Mn compounds are common and may be found in the metallography of magnesium alloys. The further fact that by the Elfin process of adding anhydrous ferric chloride to the metal a good grain refining action is obtained without superheating seems to clinch the point that iron is one of the elements that play some role in the grain size behavior.

On the other hand, it must be remembered that the ferric chloride process did not work at temperatures below the superheating range unless manganese was also present. Also, experiments have been presented which showed that good grain refining could be obtained both by superheating and carbon refining of casting alloys essentially free of iron ( $<0.001$  per cent). These facts show that iron is not essential to the grain refining action. Furthermore, it has been shown that common casting alloys containing a normal amount of iron (0.01 to 0.02 per cent) show a coarser as-alloyed grain size, 0.015 in. to 0.02 in. A.G.D., and tend to superheat and also coarsen readily at low temperatures. The same alloys with low iron ( $<0.001$  per cent) have a native moderately fine grain size, 0.006 in. to 0.010 in., which is refined to 0.004 in. by superheating and does not coarsen at low temperatures to the same degree as in the higher iron content alloys. These observations were made on several check tests and are believed to be reliable.

Several alloys made with intermediate iron content, 0.003 to 0.004 per cent, gave intermediate grain size response. Therefore, it may be stated that **before superheating** the grain size of these alloys is directly proportional to the iron content. It should be remembered that the alloys mentioned thus far have a normal manganese content, and that even with a high iron content they all reach the same fine grain size after superheating. From the foregoing data one might speculate that before superheating the iron is actually coarsening the

grain size in an alloy that would be fine by nature if iron were absent. If this were true, superheating would then be a mechanism to offset or overcome this coarsening effect.

Since it has been established that iron is a coarsening agent at low temperature, it is necessary to explain why the ferric chloride process works. There are the possibilities that, first, since the process is carried out close to the superheating temperature the heat of the ferric chloride plus magnesium reaction may be great enough to raise the metal temperature locally into the superheating range; or second, the ferric chloride is reduced to give fine iron particles which may act as nuclei for crystallization. Our later discussions may clarify the true action. The iron effect may be summarized by simply stating that:

1. Iron appears to play an important although non-essential part in grain size behavior.
2. At low temperatures, iron actually gives coarser grain size than the same alloy without iron.

### The Manganese Hypothesis

The story regarding manganese will be just like that of iron in many respects. Manganese, also, is present in nearly all commercial magnesium alloys, and it has a temperature dependent solubility in the molten alloys which is greatly affected by the aluminum content.

Tiner<sup>15</sup> indicates the importance of manganese to the effective superheating of Mg-Al and Mg-Al-Zn base alloys in that he consistently obtained finer grain sizes in alloys with greater than 0.15 per cent manganese than with 0.02 per cent or no manganese. It was also observed that superheating could take place when iron was present without manganese. Tiner, therefore, concluded that it was important to have manganese or iron present in excess of the amount soluble at the freezing point of the alloy.

Tiner did not work extensively with casting alloy compositions containing exceptionally high manganese, hence did not report the inability to get superheating effects as found in our work.

It is concluded from these observations that, like carbonaceous additions and iron, manganese takes an important but non-essential part in the grain refining mechanism. Unlike iron, manganese does not appear to have a coarsening action when present in the amounts commonly found in casting alloys.

The effect of manganese in the above amounts in preventing abnormal grain growth (germination) during solution heat treatment has been mentioned in this paper and should be kept in mind with respect to the work of Tiner on alloys containing high aluminum and low manganese<sup>15</sup> inasmuch as he solution-treated all samples before examination and may have had considerable grain growth.

All of the foregoing grain refining mechanisms fall nicely into the general theory of temperature dependent solubility. One other possibility exists and that would be soluble gases. These can be dismissed from consideration in view of the following facts:

1. None of the common gases tried (except carbon compounds) gave significant or dependable refining at least on moderate sized melts. Chlorine, hydrogen

chloride, and certain others may have worked on small melts because of the thermal effect of the considerable amount of heat locally generated.

2. Superheating phenomena were independent of whether the melt was gassed with hydrogen or degassed with chlorine, inert gases, etc. There was no refining unless the thermal, carbon, or other definite grain refining treatment was given.

3. The thermal grain refining treatment was effective even when carried out under vacuum or in a closed system.

The general summary of all the foregoing data on the temperature solubility theory is that there are three materials, carbon, iron, and manganese, no one of which is essential but which have an interdependence with each other and with temperature to bring about grain refining. This theory explains fairly well most of the observed results. However, it has at least one bad weakness in that it does not explain why a normal casting alloy when remelted, first shows a relatively fine grain size at just above the melting point and then coarsens at an increasing rate with time and temperature increase until a critical temperature is reached above which refining starts and continues until a final temperature is reached at which coarsening again begins. Solubility behavior alone would not seem to furnish the answer to these reversals of result.

A better explanation of the observed results was found as a result of some laboratory studies not heretofore reported. The electron microscope was first used to examine and compare unrefined versus superheated alloys. The first observation of interest was that, in general, the fine-grained or superheated alloys showed more prominent grain boundaries, that is, they appeared to be broader or darker. The conclusion was drawn

that grain boundary prominence was inversely proportional to grain size. This correlation was not perfect and thus far no great significance has been attached to the observation.

The next study consisted of making microradiographs at 100 diameters of alloys at various stages of grain refining and coarsening treatments. The samples for radiographing were first machined so that the sample thickness corresponded closely to one grain diameter.

Figure 13 shows commercial AZ92 alloy after three stages of thermal treatment. The test samples were all poured at 1400 F. The "as alloyed" piece had a grain size of 0.028 in. and shows a fine overall cloud background. After superheating for 15 min at 1650 F the grain size had refined to 0.0025 in. and the radiograph shows clearly defined larger white (dense) particles. After holding for 2 hr at 1250 F the grain size had coarsened to 0.010 in. and the structure shows crosses and needles of fairly large size—not too sharply defined.

Figure 14 shows another commercial AZ92 alloy. Again the "as alloyed" sample had a grain size of 0.028 in. and shows the overall cloud background of fine particles. After treating the melt with 0.5 per cent hexachlorethane the grain size refined to 0.005 in. and the structure shows a large number of medium-sized sharply-defined particles. On holding 2 hr at 1250 F the grain size coarsened to 0.009 in. and the structure shows many large fuzzy particles in fine discontinuous cloud-like background.

Figure 15 shows commercial AZ92 alloy. After refining the grain to 0.003 in. by adding 1 per cent ferric chloride the structure shows a large number of moderately well-defined medium-sized particles. On superheating for 15 min at 1650 F and then holding 2 hr at 1250 F, the grain size had coarsened to 0.014 in. and

Commercial AZ 92 Alloy — Cast at 1400 F

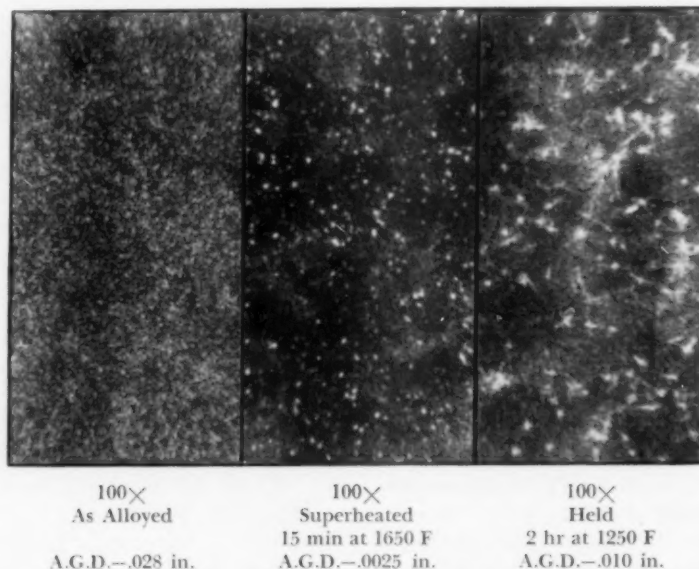
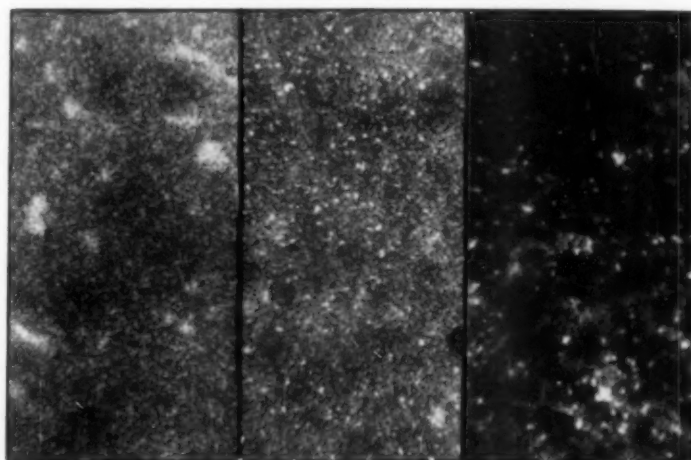


Fig. 13—Microradiographs as Function of Thermal Grain Refining Treatment.



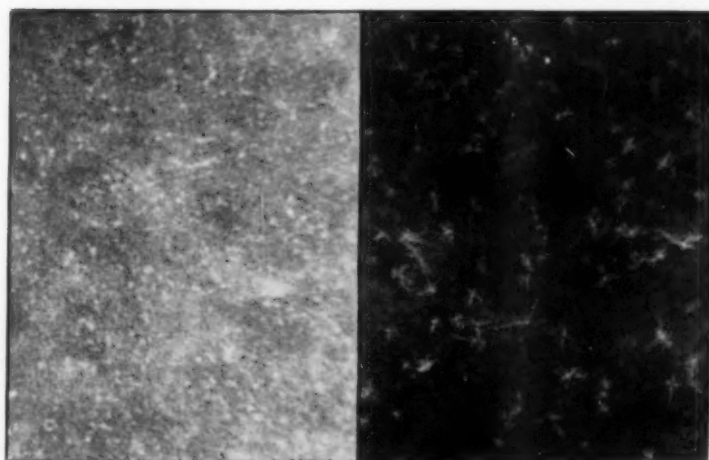
Commercial AZ 92 Alloy — Cast at 1400 F



100×	100×	100×
As Alloyed	0.5% $C_2Cl_6$ Added	Held 2 hr at 1250 F
A.G.D.—.028 in.	A.G.D.—.005 in.	A.G.D.—.009 in.

Fig. 14—Microradiographs as Function of Carbonaceous Grain Refining Treatment.

Commercial AZ 92 Alloy — Cast at 1400 F



100×	100×
1.0% $FeCl_3$ Added	Superheated 15 min at 1650 F. Held 2 hr at 1250 F.
A.G.D.—.003 in.	A.G.D.—.014 in.

Fig. 15—Microradiographs as Function of  $FeCl_3$  Grain Refining Treatment.

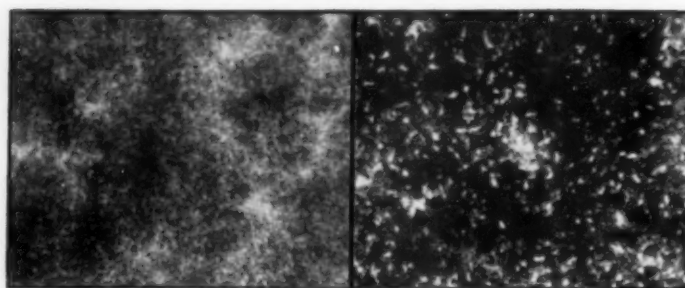
shows fuzzy large needles, stars, and crosses.

Figure 16 presents high purity AZ92 alloy (no iron). "As-alloyed" it had a grain size of 0.008 in. and shows segregated milky clouds of fine particles. On holding at 1250 F for 1 hr the grain coarsened slightly to 0.010 in. and shows coalesced larger irregular particles and tendency for fuzzy crosses and stars. After superheating for 15 min at 1650 F the grain refined to 0.004 in. and the structure is now made up of discrete large equiaxed particles. On holding 2 hr at 1250 F the grain size coarsened to 0.007 in. and the structure shows large

fuzzy needles, stars, and crosses.

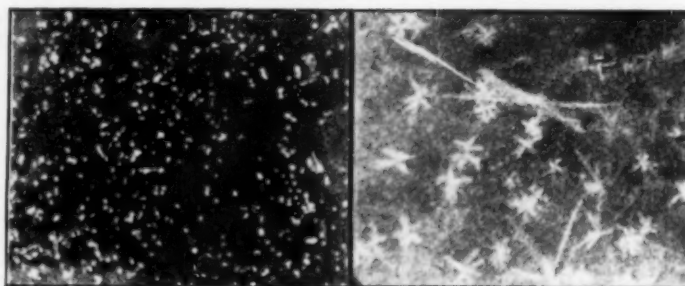
The foregoing microradiographic pictures certainly give factual evidence that definite metallographic changes are actually taking place corresponding to grain refining treatments and ultimate grain size. The general trend in the pictures shown and in others not shown is a general fine particle background in alloys before grain refining treatments, and tendency of the particles to grow in size and approach the characteristic stars and crosses if the alloy is held for appreciable time at temperatures below the superheating range.

## High Purity AZ 92 Alloy (No Fe) Cast at 1400 F



A-100 $\times$   
As Alloyed  
A.G.D.—.008 in.

B-100 $\times$   
Held 1 hr at 1250 F  
A.G.D.—.010 in.



C-100 $\times$   
Superheated 15 min at 1650 F  
A.G.D.—.004 in.

D-100 $\times$   
Held 2 hr at 1250 F  
A.G.D.—.007 in.

Fig. 16—Microradiographs as Function of Thermal Grain Refining Treatment—High Purity Alloy.

This is accompanied by grain coarsening. After superheating or other refining treatments the particles seem to become fewer in number, more equiaxed, and more sharply defined. When the metal is held at low temperatures after the refining treatment, the particles grow in size, assume the form of needles, stars, and crosses and the structure is less sharply defined. This corresponds to grain coarsening. It is interesting also that the rate of formation of the needles, stars, and crosses on low temperature holding after refining is much greater than before refining which correlates well with the relative grain coarsening rates.

The needles, stars, and crosses are characteristic of alloys containing manganese. If iron is present **without manganese** the general trend with respect to size of particles is similar but the structure is free of the needles, stars, and crosses and shows equiaxed large particles when the grain is fine and a cloud-like fine particle structure when the grain is coarse.

In order to identify the phases present in the foregoing samples, characteristic particles of each type were examined in the electron microscope. Figure 17 shows one of the star-shaped particles at a magnification of 5800X. Electron diffraction pictures were taken to determine the probable crystal structure and lattice constants of each phase. Space will not permit including a detailed presentation of this phase of the work but the following interesting results are indicated:



Fig. 17—Electron Microscope Picture of Star-Shaped Phase Believed to be  $MnAl_6$ . Mag.—5800 $\times$ .

In the casting alloys containing manganese:

- a. At temperatures below the superheating range the predominant phase has an orthorhombic crystal structure that appears to be  $MnAl_6$ .
- b. At temperatures in the grain refining range some of the  $MnAl_6$  phase still exists but a large amount of an "X" phase is present which is either in the hexagonal crystal system or has a hexagonal habit plane of crystallization and has lattice constants close to those for magnesium. There is reason to believe that this is  $MnAl_4$ .
- c. At temperatures in the grain coarsening range after superheating it has been determined that the "X" phase has practically disappeared and the  $MnAl_6$  phase is again predominant.

2. In an AZ92 alloy containing iron but no manganese, a similar series of phases exists that are temperature dependent, and a "Y" phase shows up in the grain refining range and disappears in the grain coarsening range. The "Y" phase while not yet identified has a definite hexagonal habit plane.

3. If AZ92 alloy containing manganese and no iron or both manganese and iron is treated with carbon (hexachlorethane), a "Z" phase which has a hexagonal habit plane is again produced within the grain refining range. The alloy with no iron and no manganese has not been examined.

4. In the alloy wherein two portions of AZ92 alloy were mixed with one portion of Mg-2Mn alloy, no hexagonal phase was ever found even in the superheating range and no grain refinement was obtained presumably because of the excess manganese.  $MnAl_6$  was found and also a body-centered cubic Mn-Al phase.

5. AZ92 alloy showed normal phases before and after superheating, but after beryllium addition  $MnAl_6$  was the predominant phase and present in much smaller particles. There was no hexagonal phase.

Microradiographs of this same alloy also show that beryllium additions greatly modify the structures obtained and eliminate the needles, crosses, and stars normally found after holding but still produce extremely coarse grain.

Inasmuch as in all the foregoing cases where grain refinement was obtained a temperature dependent phase of hexagonal crystal habit was also found, some consideration was given to what phases or compounds are known that would have this crystal habit and also have lattice constants close to that for magnesium since it is obvious that the presence of such a solid phase would be ideal for nucleating the crystallization of magnesium. The Mn-Al equilibrium diagram given by Mondolfo<sup>28</sup> shows that:

1.  $MnAl_6$  is stable below 1310 to 1320 F.
2. Above 1310 to 1320 F this phase changes to  $MnAl_4$ .
3. At around 1550 F  $MnAl_4$  goes to  $MnAl_3$ .

Considering that the above phases may be altered somewhat by the presence of magnesium, iron, carbon, etc., the temperature ranges check well with the occurrence of grain refining effects. It is known that the  $MnAl_6$  is not a favorable structure for nucleation, the  $MnAl_4$  would be ideal, and the nature of  $MnAl_3$  is not yet known. If it should turn out to be other than hexagonal it would provide an excellent explanation of the

over-superheating effect since the coarsening only occurs on samples quickly cooled from the high temperature.

While no such complete information is as yet available on the Fe-Al or modified phases it is relatively certain that similar phase changes occur as indicated by the electron diffraction results.

Furthermore, investigation shows that other likely materials that could fit into the hexagonal habit behavior and appropriate lattice constants are: carbides of iron, silicon, aluminum, calcium, and at least one of the carbides of manganese, aluminum nitride, aluminum chloride, metallic zirconium, and zirconium-zinc. There are other possibilities but the foregoing come closest to those likely to be present in magnesium metallurgy.

In summarizing the significance of the foregoing preliminary laboratory studies, it is concluded that there are definite evidences of phase changes in magnesium alloys that correspond to the known thermal and grain refining treatments, and that the phases present when grain refining is known to take place are favorable to nucleation, and conversely, when coarsening is taking place the favorable nucleating phases are disappearing and unfavorable phases are taking their place. It is further concluded that a temperature-phase relationship theory goes much farther in explaining the entire grain refining mechanism since it can take into account the interaction of the known grain and phase modifying elements.

### Summary

As a final summary of this paper it may be stated that research on the grain refining of magnesium alloys shows that:

1. Magnesium casting alloys as commonly processed have a unique fine-grained structure that gives high mechanical properties and eliminates many foundry difficulties.

2. There are at least four major methods of grain refining that give good results on commercial type operations:

- a. Superheating
- b. Carbon or carbonaceous inoculation of which chlorine plus hydrocarbon gas or vapor, direct addition of carbon, hexachlorethane, or calcium carbide are outstanding.
- c. Violent agitation plus carbonaceous gases or solids.
- d. The German Elfinal or ferric chloride inoculation process.

The superheating process is still widely used but may be displaced by the carbon inoculation methods as experience is gained.

3. The effect of alloy composition variables, time, temperature, rate of chill and many other factors affecting grain size have been studied and the important conclusions have already been summarized (p. 14). Several significant conclusions are:

- a. Grain refining methods are not effective or required for drastically chilled small section castings such as pressure die castings.
- b. The Mg-Al-Mn and Mg-Al-Mn-Zn alloys are



most readily grain refined and are the alloys of most interest to foundrymen.

4. Grain refining can be done under vacuum, in a closed system away from atmospheres and fluxes or with gassed or degassed metal.

5. Fine-grained metal can be made continuously available at low temperatures, 1250 to 1300 F, if this metal is in continuous metal contact with a portion of the melt which is held at temperatures above 1475 F.

With the mechanism of grain refining considered in the light of existing information and theories and in view of metallurgical studies presented in this paper, the hypothesis is offered that grain refining or coarsening is controlled by the presence of temperature dependent phases that have favorable or unfavorable crystal habits for nucleation of magnesium. Microradiography and electron diffraction studies have shown that when grain refining is taking place by any of the common practices, phases can be found that have a hexagonal crystal habit and lattice constants favorable to nucleate magnesium crystallization. Conversely, when coarsening is taking place these hexagonal habit phases are disappearing or absent. In the case of AZ92 alloy which contained the usual amount of manganese, sufficient identification of the actual phases formed and verification of their temperature equilibrium conditions was accomplished to substantiate the hypothesis. The possibilities offered by these techniques and hypotheses have only been touched upon and, no doubt, a far greater understanding of grain refining behavior as well as many other related phenomena will be forthcoming.

### Acknowledgment

I wish to express my appreciation to all those whose published results have been so freely drawn upon. In particular, I wish to pay tribute to the members of the Magnesium Laboratories of The Dow Chemical Co., whose work has been presented here in cross section for the first time. It is contemplated that much of this work will be reported in more detail in future technical papers.

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## DO YOU USE YOUR COST SYSTEM?

Bv

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## ABSTRACT

*The author stresses the importance of having in operation a good cost system to know your costs of production and operation. If you charge too little for your castings the foundry will eventually go out of business. If you charge too much your customer may go out of business.*

Appended to the paper are payroll, production order and standard cost procedures as followed by a large gray iron foundry.

ALMOST ALL of the foundry trade associations in the United Kingdom and the United States have published or have available good cost systems. For the average foundry, it is more a matter of deciding how simple or how complex the cost system should be and then putting it into operation.

The problem in many foundries is that a great deal of time and money is expended in obtaining the correct results of past operations and then these "costs" go into a binder or folder and are "kept" under cover. The writer has been in a number of foundries where the costs are carefully completed and balanced with the ledgers in from two to four months after the finished castings have been shipped and then the cost reports are filed. This method of "keeping costs" may give employment in and be a pleasure to the cost department, and the pay roll and other costs may be distributed to the penny, but it does not help in intelligent and profitable operation of the foundry.

No cost system is of real value unless it can be and is used to judge current operations and accurately fore-

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cast results. How can this be done? The system may be used for both current and future operations.

## Current Operations

Costs should be completed quickly and compared with the estimating sheets to ascertain that the estimate is correct.

Prices on new orders should be checked and if the cost experience is fairly recent this cost may be used as a guide.

If the estimate is correct, then the labor and other costs become "standard" and only those costs which are under or over standard are reported.

Almost all of the overhead items are costed as a percentage of direct labor in the departments. Usually in the preparation of this overhead percentage, *excess* direct costs are included in the direct labor money total on which the overhead percentage for indirect costs is established. And, unless the excess labor cost can be eliminated, then it must become a part of the standard cost used on the estimate sheet for the particular job or that part of the overhead will be lost.

For intelligent operating management, these excess costs must be brought to the attention of the supervisors promptly. What good does it do to tell a superintendent in March that a job cost too much in January? He should know within a day or two in order to make corrections while the job is running.

A daily report of excess charges may be made as shown in Fig. 1.

Department and plant supervisory meetings should be held to discuss these excess charges and proper action should be taken. As stated above, if the excess cannot be eliminated, then it must become a part of the standard and included in the cost on the estimate sheet. These daily reports should then be combined and reported on a weekly summary as shown in Fig. 2.

*Fig. 1—Daily labor loss report records excess charges.*

[illegible]



# SUMMARY OF LABOR LOSSES AND COST OF COMPLETED CASTING SALES

WEEK NO. \_\_\_\_\_  
 DATE \_\_\_\_\_  
 PROJECT NO. \_\_\_\_\_  
 NO. OF WORKING MEN \_\_\_\_\_

## LABOR LOSSES ON EXCESS PRODUCTION

PATTERN NO.	PATTERN REMARKS	MOLD		SLINGER		CORE		SHAPE OUT		BELL		BAND PLAST		SCALE		BRIDGE		GUY		TOWER		TOTAL LABOR LOSS	TOTAL LABOR COST	TOTAL LABOR COST PER LB. OF CASTING
		PAUSE TIME	LOSS TIME	PAUSE TIME	LOSS TIME	PAUSE TIME	LOSS TIME	PAUSE TIME	LOSS TIME	PAUSE TIME	LOSS TIME	PAUSE TIME	LOSS TIME	PAUSE TIME	LOSS TIME	PAUSE TIME	LOSS TIME	PAUSE TIME	LOSS TIME	PAUSE TIME	LOSS TIME			

Having the historical record of past costs on a particular casting and the percentage increases or decreases in labor, metals, etc., it is an easy matter to forecast results on the basis of past experience. These results are checked currently by the method shown under "Current Operations." The estimate sheet is revised by using the current change percentages and then the new estimate becomes the "standard cost."

### Future Operations

Overall average percentages should not be used. The writer has seen cases where a foundry has used averages and compared, say, January 1947 with January 1948, where increases have been as follows:

	Average Per Cent of Total	Per Cent Increase Jan. 1948 over Jan. 1947	Present Cost	Average Increase Per Cent
Metals . . . . .	14	50	21.00	
Direct Labor ..	21	12	23.52	
All other costs ..	65	17	76.05	
	<hr/> 100	<hr/>	<hr/> 120.57	20.57

(It must be emphasized that the percentages shown are for purposes of illustration and are not actual)

Using average figures, this percentage of 20.57 would be used on all castings.

It would be quite a task to get these percentages on the individual castings, but the work in the average foundry can be divided into a number of classifications, the number depending on the complexity of the work.

How do the figures on certain classes of castings compare with the average? A comparison of a flywheel and a cylinder block for a Diesel engine follows.

	FLYWHEEL		
	Per Cent of Total Cost	Per Cent Increase	Present Cost
Metals .....	50	50	75.00
Direct Labor .....	12	12	13.44
All other costs .....	38	17	44.46
	100		132.90
Per cent increase January 1948 over January 1947 — 32.90			

Per cent increase January 1948 over January 1947 — 32.90

	CYLINDER BLOCK		
	Per Cent of Total Cost	Per Cent Increase	Present Cost
Metals .....	19	50	28.50
Direct Labor .....	19	12	21.28
All other costs .....	62	17	72.54
	100		122.32
Per cent increase January 1948 over January 1947 —			22.32

Per cent increase January 1948 over January 1947 — 22.32

The percentage difference here is 12.33 per cent (32.90 per cent minus 20.57 per cent), which on \$1,-000,000 production would be \$123,000 and would probably be more than the entire anticipated profit.

If the average figure is used, under competitive conditions, a foundry using these and quoting against those using cost figures intelligently, would tend to fill

Fig. 2—Weekly summary of labor losses and cost of completed casting sales compiled from daily reports.

up with work similar to the flywheel and would lose about 12½ per cent on each job. This figure might well be 20 per cent and higher on some jobs.

There are many other ways in which a cost system can be used profitably. Competition will be getting keener, the days of a seller's market are on the way out. Reference has been made chiefly to excess costs, but it must not be forgotten that where costs are lower than average the customer should get the benefit. If the casting price is too low the foundry will eventually go out of business. If the price is too high the customer may go out of business.

In the following section of the paper the methods used in applying payroll, production order and standard cost procedures in a foundry are outlined.

### Payroll Procedure—Cleaning Department

*Former Procedure*—Under the procedure formerly in effect in the castings cleaning department, each employee kept a record of his individual production on an operation ticket. He recorded thereon the pattern number, operation, and number of pieces with respect to each pattern worked on during the day. At the end of the day the data shown on the operation ticket was copied on a second operation ticket, so that two identical records of production for each employee would be available. One copy would be retained by the individual employee to furnish him with a record of the work performed during the day. It would also serve as a medium on which he could compute his earnings for the day. The other copy was turned over to the timekeeping department for payroll purposes.

In the timekeeping department the information shown on each operation ticket for each employee was completely transcribed daily onto a summary of operations. This set forth a record of the work performed by each employee and the related earnings classified as to direct and indirect payroll, based on piecework, and the hours of daywork. The daily piecework earnings and hours of daywork for each employee were posted daily on a card, from the summary above referred to. The earnings for the payroll week were totaled and recorded thereon after the summary card had been matched with the related attendance clock card. The payroll check was issued on the basis of the information appearing on the summary card.

The foregoing method of compiling the payroll data involved a certain amount of duplication of clerical effort. In addition, it did not provide the basis for the accumulation of cost data for each lot of castings.

The revised procedure provides that each employee



for use of the office in computing the payroll and compiling statistics relating to the foundry production.

The classification of time and earnings will be made in the timekeeping department, but the clerk in the cleaning department will, at the end of the day, record the total number of tickets for each employee including daywork tickets, if any, on the reverse side of one of the production tickets. This will furnish the timekeeping department with a means of determining that all of the tickets for each employee for the day are accounted for. In addition the clerk will record on the

casting or otherwise mark it to indicate to the employee that a re-operation is involved, as the position of the casting in the production line will constitute notice to that effect.

All special tickets reporting re-operations should be reviewed by the foreman before being forwarded to the timekeeping department. The tickets should be noted by him as to the department responsible for the special work, and the reasons therefor, so that proper accounting distribution may be made of the related payroll charges, in the timekeeping department.

Fig. 4—Daywork ticket to report all indirect labor.

same ticket the total hours worked by the employee for the day. The information for this record will be obtained from the attendance clock card.

Regular piecework tickets which are not pre-slotted with the operation number will be made available to each employee. These tickets will be used in instances where the employee is engaged on a piecework operation to which he is ordinarily not assigned, in which case it will not be possible to pre-slot the card for the operation number; the operation number will be slotted after the card has been turned in to the timekeeping department.

In addition to the regular production ticket described above, a special production ticket will be used in the cleaning department as follows:

**Special Production Ticket for Reporting Re-operations**—The special production ticket is also identical in form with the regular production ticket, except that no provision is made thereon to record extra operations as no extra operations should be involved in connection with the work to be reported on this ticket. The special operation ticket is to be used to record re-operations, necessitated by reason of faulty workmanship. Re-operations will usually be required by reason of welds or re-welds. The ticket will be pre-slotted for employee and operation number.

When a casting has been rejected by an inspector and one or more operations must be repeated thereon, because of welds or re-welds, or for other reasons, the work must be reported on the special ticket provided for the purpose. It will not be necessary to tag the

Tickets containing a record of extra operations and re-operations should be summarized daily in the timekeeping department. A daily report of the related excessive labor costs should be furnished to the foundry. This report will contain a record of the excessive labor costs at each operation with respect to each pattern number as well as the total amount for the day. In addition, these excessive labor costs will be set out separately in the general accounts, classified as to the departments responsible.

Labor loss on each pattern by operations, representing the difference between the quantity of castings reported as produced at each operation and the quantity shipped (after allowance for normal scrap) shall be determined when the job is finished and summarized on a weekly report (Fig. 2). In this manner any failure on the part of the foundry to report defective work as such will be subsequently disclosed.

#### Reporting Indirect Labor

**Daywork Ticket**—The daywork ticket (Fig. 4) is to be used to report all indirect labor, whether paid for on a daywork or a piecework basis. These tickets will be pre-slotted for the employee number only.

Each employee in the cleaning department will be furnished with a supply of the three types of tickets previously described. Each type of ticket will carry a different colored border to readily distinguish it from the others. They will all be addressographed in advance



Fig. 5—Two-part production ticket used in mold department to report direct labor operations.

with the employee's name and clock number, and will be pre-slotted with the employee's clock number. In addition, the regular ticket and the special ticket for reporting re-operations will be pre-slotted with the operation number.

#### Payroll Procedure—Molding Department

The revised payroll procedure requires that the mold checker prepare a separate production ticket covering the production of each employee on each separate pattern worked on during the day. The production ticket is designed to facilitate preparation of the payroll and tabulation of labor statistics with reference to castings produced compared with the quantity shipped.

Two separate forms of tickets will be used to report labor operations in the mold department—first, a production ticket for reporting all direct labor operations, and second, a daywork ticket for reporting all indirect labor operations, whether paid for on a daywork or piecework basis.

**Production Ticket**—A sample of the production ticket to be used in the mold department for reporting direct labor operations is shown in Fig. 5.

The production ticket consists of two parts: (a) original, on paper stock, which will be used as a scrap report in the manner hereinafter described and (b) duplicate, which will constitute the record of production on which employee earnings will be computed.

Space is provided on the outer edge of the ticket to record the numerical code designation for the following: Employee number; labor operation number; pattern order number; payroll period—year and week; labor loss.

The employee number and labor operation number will be slotted on the tickets, and the name and clock number of the employee addressographed thereon, in the space provided, before the tickets are distributed for use. The pattern order number, payroll period, and labor loss will be slotted on the tickets after they have

been turned in to the plant timekeeping department.

The reason for providing for a labor loss code designation on the border of the card is to permit ready accumulation in the timekeeping department of the labor cost of broken molds, company broken castings, and company welds. The recommended procedure provides that tickets containing a record of the above operations be summarized daily in the timekeeping department, and a daily report of the related labor costs be furnished by the foundry, segregated as to pattern number. In addition, the amount of these labor losses will be set out separately in the general accounts. Excessive labor costs must be properly segregated and classified in order to be adequately controlled.

The various molders will not record their own production; this will be done by the mold checker in conformity with the practice heretofore followed. The mold checker will prepare a separate ticket covering the production on each pattern by each employee during the day. In the body of the card space is provided for the following data which will be recorded thereon by the employees indicated:

**Mold checker on original and duplicate tickets:** Pattern number; mold time; floor; molds rammed; pieces per mold; molds broken; molds not poured; molds poured by molder; date.

**Timekeeping department on duplicate ticket:** Pattern order number; pieces poured; bad pieces; broken pieces; pieces to weld; credit welds; good pieces; molds paid; weight each; shakeout labor rate and amount; mold and pour rates; mold and pour amounts; pattern change amount; total to pay; reasons for defects.

**Inspector on original ticket:** Reasons for defects.

The mold checker formerly recorded the above information on a piece of scratch paper from which it was later transferred to a form designated as a mold checker's record. The information shown on the mold checker's record as to the production of the various molders for the day was subsequently transferred to a daily summary on which the earnings of the various employees in the molding department were later computed. The computation of earnings for each employee

was made when the scrap report relating to the day's production was turned in by the inspector, which permitted the determination of the number of pieces on which the employee was entitled to receive payments.

The information to be noted on the production ticket by the mold checker, as outlined above, will be written on the original copy of the ticket, but it will also appear on the duplicate by reason of carbon spots affixed to the reverse side of the original.

### Pattern Changes Recorded

The number of pattern changes made by each molder during the day will be accumulated and recorded by the mold checker on the ticket covering the last production for each molder for the day.

Slinger molding is a crew operation. The mold checker will prepare a production ticket covering the production on each pattern by the slinger molder crew. In addition one ticket for the day will be prepared for each man participating in the slinger mold production, on which will be indicated that the employee is a member of the crew, and the total number of hours he has worked in that capacity. The distribution of the crew earnings will be made in the timekeeping department.

With respect to shakeout, which is also a crew operation, a daily ticket will be prepared for each member of the shakeout crew, showing total hours worked in that capacity. The labor cost of the shakeouts relating to each pattern will be computed on each production ticket in the timekeeping department, based on the shakeout rate applicable to each pattern, and the total shakeout labor for the day determined, which will be divided between the individual crew members according to hours worked.

After the production tickets have been completed by the mold checker in the manner indicated above, he will detach the original from each ticket and pass it along to the inspector, which the latter will use to record the scrap relating to the particular production. The duplicate of the production tickets will be given to the timekeeping department.

### Scrap Classified

The inspector will record on each ticket the quantity of scrap relating to the particular production, classified according to the reasons, and indicating the responsibility, as set out on the ticket. Such classification of scrap involves no changes in the procedure formerly followed. He will then forward these scrap reports to the timekeeping department, where they will be matched with the related duplicate tickets, and the information transferred to the latter record. The duplicate ticket will then contain a complete record of the production for which the employee will receive payment, as well as a record of the scrap produced.

When the information shown on the scrap reports has been transferred to the duplicate tickets covering the production for the day, the scrap reports will be returned to the foundry where they will be turned over to the respective employees to furnish them with a record of their production.

The mold checker will, at the end of the day, record the total number of tickets for each employee, including daywork tickets, if any, on the reverse side of one of the production tickets. This will furnish the timekeep-

ing department with a means of determining that all of the tickets for each employee for the day are accounted for. In addition, he will record on the same ticket the total hours worked by the employee, which information will be obtained from the attendance card.

**Daywork ticket**—The daywork ticket is to be used to report all indirect labor, whether paid for on a daywork basis or a piecework basis.

Individual daywork tickets should be prepared for each member of the mold pouring crew.

The daywork tickets will be pre-slotted for the employee number only.

The mold checker will be furnished with a supply of the production and daywork tickets for each employee. The two types of tickets will carry different colored borders to distinguish them from one another. They will be addressographed in advance with the employee's name and clock number, and will be pre-slotted with the employee's clock number. In addition, the production ticket will be pre-slotted with operation number.

### Payroll Procedure—Core Department

Under the procedure formerly in effect in the cleaning department, each core checker kept a record of daily production of cores for employees by pattern number. These reports were turned in to the timekeeping department each day, where they were posted to a daily summary of cores produced by pattern number for each core maker. At the end of the weekly pay period, the total production for each pattern was ascertained and extended at the established piecework rates, and the total piecework earnings for the employee were determined and recorded on his weekly summary of production.

The revised procedure provides that the core checker make out a separate production ticket covering the production of each employee with respect to each pattern worked on during the day. The production ticket is designed to facilitate the preparation of the payroll and the tabulation of labor statistics with reference to each casting produced, compared with the quantity shipped.

Two separate forms of tickets will be used to report labor operations in the core department—first, a production ticket for reporting all direct labor operations, and, second, a daywork ticket for reporting all indirect labor operations, whether paid for on a daywork or piecework basis.

**Production ticket**—A sample of the production ticket to be used in the core department for reporting direct labor operations is shown in Fig. 6.

It will be noted the ticket is designed to provide space on the outer edges thereof for the numerical code designation for the following: Employee number; labor operation number; pattern order number; payroll period—year and week.

The employee number and labor operation number will be slotted on the tickets, and the name and clock number of the employee addressographed thereon, in the space provided, before the tickets are distributed for use. The pattern order number and payroll period will be slotted on the tickets after they have been turned in to the timekeeping department.

Each core maker will prepare a separate ticket covering his production with respect to each pattern num-

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Fig. 6—Production ticket used in core department for reporting departmental direct labor operations.

ber, recording on the ticket the following information: Pattern number; core name; core time; cores made; and date.

In addition he will record the core change, if any, classified as to type, in the space provided on the ticket. The number of core changes made by a core maker during the day will be accumulated in the timekeeping department, from the information shown on the individual tickets, and be recorded on one ticket for the purpose of computing the earnings applicable thereto.

The clerk in the core department will, at the end of the day, record the total number of tickets for each employee, including day-work tickets, if any, on one of the production tickets. This will furnish the timekeeping department with a means of determining that all of the tickets for each employee for the day are accounted for. In addition he will record on the same ticket, in the space provided on the reverse side of the ticket, the total hours worked by the employee, which information will be obtained from the attendance clock card.

**Daywork ticket**—The daywork ticket is to be used to report all indirect labor, whether paid on a daywork or a piecework basis. These tickets will be pre-slotted for the employee number only.

Each employee in the core department will be furnished with a supply of the production and daywork tickets. Each type of ticket will carry a different colored border to readily distinguish it from the other. They will both be addressographed in advance with the employee's name and clock number, and will be pre-slotted with the employee's clock number. In addition, the production ticket will be pre-slotted with the operation number of the work to be performed.

#### Payroll Procedure—Cupola Department

A special payroll ticket (Fig. 7) has been designed for reporting cupola production.

The ticket will be pre-slotted with the operation number and employee clock number, and will be addressographed in advance with the employee's name

and clock number. This refers to the name of the employee responsible for the cupola work. As will be noted, provision is made on the bottom of the ticket for the name of the helper, whose name will be inserted on the ticket by the crew leader.

The tons of metal melted and time elapsed will be recorded on the ticket by the crew leader, as well as the time worked by the helper.

In addition, it will be necessary for the helper to make out a ticket on which will be shown the hours worked, which ticket will be used in the timekeeping department to record his daily earnings, as computed on the cupola ticket.

#### Timekeeping Department Procedure

When the daily production and daywork tickets are turned in to the timekeeping department by the various foundry departments, they will be handled in the following manner:

A. The tickets will be reviewed to see that all of them are on hand according to the number of tickets reported by the foundry timekeeper as covering each employee's work for the day, and also to see that the date is shown on all tickets.

B. The tickets will be gang-punched to record the pay period.

C. Production tickets on which the operation number has not been pre-slotted will be sorted out and slotted for the operation number. Such tickets will consist principally of cleaning department tickets reporting operations on which the employee is not ordinarily engaged, and consequently for which the tickets had not been pre-slotted for operation number.

D. Tickets will be manually sorted as to pattern number, and then taken to the pattern price record, from which the standard pattern order number, and piece-work rates will be posted to the tickets. Having all tickets relating to a particular pattern number grouped together will facilitate the pricing thereof from the pattern record, and effect a saving of time as compared to the method previously employed.

E. Wherever necessary the tickets will then be regrouped according to standard pattern order number, which will in all cases identify the pattern number, and





has been responsible and the pattern numbers on which the loss was incurred. A daily total of the total loss chargeable to each department will be obtained and recorded. This daily total should be added to the accumulated total for the period to date as shown on the previous day's report so that a current total for the period to date is shown in each daily report. At the end of the period, the accumulated total of labor loss chargeable to each department, as shown by the report for the last day of the period, will serve as the basis for posting the schedule entry which will credit the total labor loss amount and charge an appropriate departmental labor loss variance account for that portion of the labor loss for which the department was responsible.

The foregoing daily report of labor losses due to extra operations and re-operations will not include the labor losses of the molding department as a result of scrap molds. A separate report will be prepared for this department.

4. A daily labor distribution will be prepared and posted to the Summary of Payroll and Labor Distribution, under the appropriate headings. To facilitate the preparation of this distribution the tickets will be key-sorted by operations and arranged in operation number sequence. This will permit of a total of the tickets for each operation being readily prepared. The total of all operations, except those of the molding department, posted as above noted, must be equal in exact amount to the control total recorded on the summary of payroll and labor distribution chart immediately upon completion of ticket pricing and extension.

The daily distribution of labor will not include overtime or special allowances as the amounts thereof will not be determinable until the end of the weekly pay period. These allowances will be made, using as a basis the average hourly rate developed for the current weekly payroll, and will be included with the tickets for the last day of the current week.

Before describing further procedures which are to be followed and operations that are to be performed on the labor tickets described in this section, the special procedures relating to labor tickets of the molding department will be outlined to that point at which procedures are identical for all departments.

#### Procedures in Timekeeping Department Relating To Molding Department Labor Tickets

In preceding paragraphs were described the operations that are to be performed on the molding department tickets up to the point of insertion of the piece-work rates on the tickets. At that point the molding department tickets were set aside due to the fact that earnings thereon could not be calculated as good production was not recorded and cannot be recorded until after scrap reports for the day are received from the foundry.

The molding department tickets, that have been segregated as above noted, will be subject to the following procedures:

1. Keysort the tickets so as to place them in numerical order as to employee number. Keep all of the tickets for the day intact pending the completion by the inspector of the related scrap reports.

2. When the scrap reports covering the day's pro-

duction are turned in to the timekeeping department they will be matched with the related production tickets, and the information shown on the scrap reports transferred to the tickets. When this is accomplished the tickets will contain a complete record of gross production, and net production on which earnings are to be computed, as well as a record of scrap produced.

3. After the information shown on the scrap reports has been transferred to the tickets covering the production for the day, the scrap reports will be returned to the foundry where they will be turned over to the respective employees to furnish them with a record of their net production.

It is important that all information appearing on the scrap report be copied on to the production ticket, as otherwise the timekeeping and cost departments will not possess a complete record of production which will be necessary in the event of question on the part of an employee as to the amount of his earnings.

4. The payroll amount should now be extended on each ticket. In addition to the determination of the mold amount, this may also involve a calculation of earnings for molds poured, if the molder has poured his own molds, as well as a calculation of earnings on account of pattern changes, which changes will be accumulated and shown on one ticket for each molder.

In addition to the foregoing, the amount of shakeout labor will be computed on each production ticket, but will not be added to the employee's earnings shown on the ticket.

5. Total amount of shakeout labor earnings due to shakeout crew members, as shown on production tickets, and distribute total earnings to individual shakeout crew tickets.

6. Distribute slinger molding crew earnings as shown on master tickets to individual slinger crew tickets.

Shakeout and slinger molding are crew operations and the total of all earnings for the day as shown on the production tickets will be divided between the individual crew members on the basis of hours worked. An individual ticket will be filed by each employee who performs any shakeout labor or is engaged in slinger molding operations. The time spent by each employee on these operations will have been entered thereon by the foundry timekeeper.

Having obtained the total of earnings due to shakeout crew members the hours shown on the individual shakeout labor crew tickets will be totaled and the total earnings divided by the total hours to arrive at an average hourly rate. This developed rate will be applied to the hours worked by each crew member and the earnings as computed entered on the ticket. For example, if the total shakeout earnings for the day amounted to \$15.00 and was earned by the shakeout crew consisting of five members who worked a total

Employee	Hours Employed	Earnings at Average Developed Rate
A .....	7	\$ 5.25
B .....	5	3.75
C .....	4	3.00
D .....	2	1.50
E .....	2	1.50
Total .....	20	\$15.00

of hours, the computation would be made as follows: total earnings, \$15.00 ÷ total hours, 20 = average hourly rate, \$0.75.

The total earnings of the slinger molding crew will be obtained from the master tickets covering the production on each pattern by the crew. These earnings will then be distributed among the various crew members on the basis of the hours worked by each, as shown on the daily ticket turned in by individual crew member. The manner of computation and distribution of the earnings of each crew member will be identical with that above described for earnings of the shakeout crew members.

7. The tickets will next be added to establish a control on the mold department payroll for the day, and the control total recorded on the summary of payroll and labor distribution chart.

The payroll and labor distribution here referred to is described in a foregoing section devoted to procedures relating to labor tickets for departments other than the molding department. The manner of recording the control total and distribution will be the same as provided for in that section. The total of the molding department tickets as daily recorded on this summary will be equal in exact amount to the amount of earnings posted to the employees' earning cards, from which the payroll will be prepared at the end of each week.

8. Keysort the tickets showing labor losses, which will comprise all tickets which include earnings for broken molds, company's broken molds, and company welds, and prepare a statement showing the total of such labor losses by pattern number under each reason.

In this connection, the labor loss to be shown for company welds is one-half the labor cost thereof and not the entire amount, which will serve to put company welds on the same basis as molder welds for labor statistical purposes.

The foregoing daily report of labor losses in the molding department will be in addition to the daily report of labor losses in the cleaning department, previously described.

9. Prepare a daily labor distribution. This will be posted to the summary of payroll and labor distribution, under the appropriate headings, in the same manner as provided for tickets of departments other than the molding department, fully described previously.

As in the case of the labor distribution relating to the other departments, the molding department labor distribution will also only include overtime and special allowance on a weekly basis, as previously stated.

Labor losses in the molding department will be charged to a special account, to be cleared at the end of the month by posting to appropriate departmental labor loss variance accounts, based on the information shown on the daily reports of labor losses.

#### Additional Procedures on Labor Tickets of All Departments

The following described procedures apply to the labor tickets of all departments, including molding.

After the daily labor distribution has been prepared, the labor tickets will be keysorted by employee number in order to group together all of the tickets for each employee for the day. The time and earnings segre-

gated as to piecework, daywork, and overtime for each employee for the day will then be posted to a daily payroll summary ticket which will be maintained for each employee during each pay period. The foundry timekeeper will have previously recorded the total number of hours worked and the total number of tickets for each day. Therefore the total hours as now posted will balance with the original hours as entered by the foundry timekeeper, and the earnings of all employees will balance with the total earnings as controlled through the entries made in the summary of payroll and labor distribution.

In the cleaning department, where employees are not required to report the time spent on productive operations, the hours spent on piecework will be determined by deducting daywork hours, if any, from the total hours.

After the time and earnings of all employees have been summarized in the above manner, the total earnings shown on each employee's summary will be footed, and agreed with the payroll control for the day previously established.

At the end of the payroll period the daily payroll summary ticket will be footed and balanced to arrive at the total of the employees earnings for the period. A grand total of all employees earnings will be obtained and agreed with the grand total of earnings for the payroll period as shown by the summary of payroll and labor distribution. The tickets will then be forwarded, together with the adding machine tape showing the grand total of the tickets, to the general accounting department for preparation of payroll checks.

After payroll checks have been written and the daily payroll summary tickets returned, they will be matched with the related attendance cards of the employees. The two cards should be clipped together and permanently filed, arranged according to employee number sequence in each payroll period. After the individual employee's labor tickets have been kept intact for a sufficient length of time, they should be keysorted by pattern order number and grouped with like tickets relating to the same pattern order number already developed in prior payroll periods.

Subsequently, where the particular order is completed and shipped, the production tickets relating thereto will be used to accumulate certain labor cost data, as previously described under "summary of labor loss on completed castings."

#### Procedure for Computing Labor Loss on Completed Castings

As already explained the labor tickets will be filed by pattern order number after the payroll for the week has been prepared. This file of production labor tickets will be maintained in the general accounting department, and refers to production tickets covering the direct labor operations shown in Table I.

Operations shown in Table I do not cover all direct labor operations involved in the making of castings, but cover by far the major part of the direct labor expended which can be readily controlled for cost purposes. The direct labor operations which will not be controlled for cost purposes include cupola labor, pouring and shakeout labor of the molding and slinger departments, and inspection and shipping labor of the cleaning depart-



TABLE I—DIRECT LABOR OPERATIONS

Name of Operation	Operation Number
Molding Department—Molding .....	5
Slinger Department—Molding .....	11
Core Department—Makers .....	15
Core Department—Pasters .....	16
Cleaning Department—Mill and tumble .....	21
Cleaning Department—Sand blast .....	22
Cleaning Department—Scale .....	23
Cleaning Department—Grind .....	24
Cleaning Department—Chip .....	25
Cleaning Department—Test .....	27

ment. It will be noted that the direct labor operations not to be controlled for cost purposes are negligible in amount and will in no way interfere with the efficient control of foundry operations by pattern numbers.

The purpose of instituting a cost control procedure is to determine and report to the foundry and general managements the amount of direct labor on major operations incurred in the production of castings in excess of the standard direct labor of those operations included in the cost estimate of the castings. The amount of the excess direct labor expended on the production of castings will be determined exclusive of the molding and cleaning departments' direct labor losses on broken molds and defective castings. As previously commented upon under the caption of "timekeeping department procedures" this latter group of labor losses will be reported daily by the timekeeping department for management control.

The production tickets, covering operations to be controlled for cost purposes, will remain in the timekeeping department files, by employee number, until sufficient time has elapsed to allow foundry employees to complete the computation of earnings for the week covered by those production tickets. When sufficient time has elapsed, the production tickets will be removed from the foundry timekeeping files and forwarded to the general accounting department for the purpose of computing costs.

#### Establish Control Record

The general accounting department will establish a memorandum control record of production tickets for the operations named at the fore part of this section. The amount to be charged to the control record will be the total of the labor distribution for the week covered by the production tickets for operations 5, 11, 15, 16, 21, 22, 23, 24, 25, and 27. Charges to this memo control record will be made weekly—one charge for each weekly payroll distribution.

The production tickets charged to the control record will be filed according to the standard pattern order number. Production tickets for succeeding weeks will be sorted into the file by standard pattern order number but it will not be necessary to sort the tickets in the file by labor operation numbers.

When the work on a particular pattern has been completed in the foundry and the casting shipped to the customer, the production tickets relating thereto will be used to develop the labor loss statistics by operations for the pattern in the manner and for the purpose set forth in the subsequent comments. To repeat, the tick-

ets are already in pattern order number, the only additional operation required being to key sort the tickets relating to the pattern in order to group them by labor operations, according to the various labor operation numbers slotted on the border of the tickets.

When a foundry order has been completed the producing foundry will send a copy of its completed foundry order, on which has been recorded the total number of castings shipped and billed, to the general accounting department. The receipt of the completed foundry order by the general accounting department will be the basis for the removal of all labor production tickets for that particular pattern from the cost file.

The production tickets for each standard pattern order number should be sorted according to labor operation numbers and, in the case of molding operation No. 5 and slinger operation No. 11, the total number of molds paid for shall be determined from the tickets and subtraction from that amount should be made representing the number of defective molds represented by the labor losses noted on the production ticket by the timekeeping department. The net production of the molding and slinger departments and the gross production of the core and cleaning departments shall be recorded on a three-part form designed to accumulate the labor losses of completed patterns by major direct labor operations. The information necessary to record the costing of a completed casting will require a single line entry on this form.

#### Determine Operating Losses

The labor loss by operations will be arrived at by ascertaining the differences between the quantities reported as produced at each operation and the quantities shipped (after giving effect to the scrap allowance provided in the estimate) and extending such differences at the established piecework rates for the various foundry operations involved.

The total labor paid for the operations controlled under this procedure will be determined by computing the quantities reported as produced at each operation and recorded on the form, at the established piecework rates for the related operations. The total of the direct labor for the labor operations being controlled should be recorded on the form. That amount will be the basis for crediting the memorandum control record for production tickets removed from the file and costed. In addition to entering on this record the labor losses as determined through the above mentioned procedures, the total of the labor losses due to defective castings previously reported to the management for daily control purposes should be accumulated and summarized on this report in the following manner. The labor cost of defective work will be shown daily on a special report segregated as to pattern numbers, operations, and cause of defectiveness. However, that daily report will not reflect the accumulated labor loss of defective work for each pattern.

The production tickets for the labor losses in the molding, slinger, and cleaning departments incurred by reason of defective castings, extra operations, and re-operations on castings have been slotted in the timekeeping department in the space provided therefor. Accordingly, these production tickets can be readily removed from the rest of the production tickets by the

patterns. The total of the defective labor losses recorded on those tickets should be recorded on the labor summary in the space which has been provided.

#### Summary

To summarize, the labor summary of completed casting orders will include the following information which will be recorded on the form in a one line entry for each pattern:

Mold	- -	Pieces produced
		Labor losses
Core	- -	Pieces produced
		Labor losses
Mill	- -	Pieces produced
		Labor losses
Sand blast	- -	Pieces produced
		Labor losses
Scale	- -	Pieces produced
		Labor losses
Grind	- -	Pieces produced
		Labor losses
Chip	- -	Pieces produced
		Labor losses
Test	- -	Pieces produced
		Labor losses
Total loss		
Labor loss - Defective castings		
Total labor loss		
Total labor cost		
Pieces shipped.		

The original copy of this summary will be forwarded to the foundry management for its study and guidance in the continued operation of the foundries. Several

situations requiring foundry attention may be disclosed through this report. For instance, a number of castings may be scrapped in the various labor operations for which no report is made. A record of excessive scrap may, upon investigation, be found to report a normal condition in respect of a particular pattern due to the requirements peculiar to such pattern. Also, castings may be paid for at the various labor operations in excess of the number of castings which could possibly have been worked on at those operations. If the labor loss is sufficient in amount and subsequent methods of production will not reduce or eliminate the labor loss, it may be necessary to revise the standard cost estimate of the casting, which may necessitate an upward revision in the selling price of the casting.

The summary forms recording the labor losses on completed casting orders should be numbered and the totals on the forms should be forwarded to obtain the total labor losses by operations for the four-week accounting period.

The information recorded on the summary sheets of completed casting orders should be recorded for permanent historical reference onto a permanent record by patterns, preferably onto the permanent pattern record, recording the following information: Labor loss report number; labor loss; labor loss—defective work; total labor loss; total labor cost.

If this information is recorded on the pattern record, it will provide the means of reviewing foundry experience on previous casting orders which should prove helpful in effectively controlling the labor cost of the current casting order.

# CONTRACTION AND DISTORTION IN GRAY IRON CASTINGS

By

E. Longden\*

THE PROBLEMS INVOLVED with contraction from the point of solidification to ambient temperature are well known to founders of castings of any shape and form. But there are certain classes of castings which present more than the average hazards to counteract. It so often presupposes a compromise metallurgically, technically and practically when the design of the component has left little room for further concessions to suit the founder.

When producing large castings of complicated, or of contrasting sections, considerable knowledge is necessary in respect to the thermal behavior of such castings during the period of cooling down from the freezing point of the metal. Nor can the conditions present during the liquid shrinkage of a metal be left out of serious calculations. But it is the experience gained empirically, as with most progress, that we must look for our principal guidance in overcoming the hazards of contraction stresses generated by the strain set up by the dual and conflicting conditions of transformation temperatures in a cooling metal and distortion due to the contrasting sections and unequal rates of cooling.

With small castings, usually required in large quantities, a trial or pilot casting can be made to ascertain the best manufacturing methods needed to take care of liquid shrinkage and solid contraction influences on the serviceability of the casting. But with large castings this approach is not possible. A large casting has often been made with all evident outward appearance of success, but it has been followed by the production of a second casting, made in a similar manner as with the first, but with much less good fortune. Such a subsequent failure is often due to the very narrow safety margin in the countering of the many pit-falls of manufacture, or in understanding the many variables possible.

A problem understood, or diagnosed, is half way to its solution. It is with this thought in mind that the author has endeavored to examine and follow the thermal behavior of certain difficult classes of castings during cooling with a view to providing standards for correctives or the elimination of errors. Most foundrymen will agree that there is an almost complete absence of published camber standards to counteract distortion in large castings, in particular of beds, tables, slides, trays and so on. In a paper<sup>1</sup> published by the author in 1932, standards were put forward for a certain class of machine tool bed. These standards have been slightly modified and extended in the meantime and are again submitted for guidance of those founders who may feel the need.

## The Subject Outlined

The subject is dealt with by outlining some of the records taken of the thermal behavior of certain large castings, in particular of simple bars and machine tool bed design, from the solidification point of the metal to ambient temperature. The methods employed to follow the movement of the castings cannot be considered as ultra-scientific, but the large-scale patient records taken provided dependable data. Most of the records were not obtained automatically, although an automatic recording instrument was built to the author's design. It was employed to check the efficiency and reliability of the simple system generally used which depends upon frequent and continuous measurements between fixed points on the castings and fixed points on the outside of the mold. The movements of the castings are graphically recorded and some small attempt is made to correlate temperature with certain of the critical points on the curves.

From results of the practical research, there is ample evidence to conclude that with cast iron castings of contrasting section, or of a form which precludes a uniform cooling rate, the strains set up introduce conflicting stresses due to the dual action of expansion and contraction operating in the same casting and the distorting effects of contraction.

It is shown that due to the dual expansion and contraction influences, common to cooling cast iron

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casting of contrasting section (in contradiction to the normal behavior of simple uniform castings), that the thicker sections take up a greater percentage contraction than the thinner members of the same casting forming the single component.

From information resulting from tests made in 1936 and onwards, the author decided to introduce a system of air cooling the slideways of a certain class of machine tool casting (a roll grinding machine bed) during the period of cooling down to room temperature. The effect of accelerating the cooling rate of the central areas of two middle slideways considerably reduced the thermal stresses and the need for camber allowances by 36 per cent and the hazards of fractures were eliminated.

A test casting, 10 ft. in length, is employed to closely study contraction and distortion with varying depths and sections. From these tests, and experience, camber standards are founded for certain classes of machine tool castings and similar castings.

Some observations on steel and cast iron gear blank castings bring out a fundamental difference between the two metals when employed in such a class of casting.

#### Influences on Contraction

The influences which determine the behavior of cooling cast iron sections of simple and regular form are:—

1. Shrinkage of the fluid volume to the point of solidification, the shrinkage varying with the composition and initial temperature of melting and superheat at the point of mold filling.

2. Chemical and physical changes in the volume of the metal from the point of solidification to room temperatures.

3. Reduction in dimensions consequent on the normal diminution in volume when the metal cools from an elevated to ambient temperatures.

In castings of complicated and contrasting section, or sections of unequal cooling rate, there are other influences which hinder and mask those stated in 1, 2 and 3 such as:

4. Designs introducing unequal and contrasting sections and machining additions to the casting.

5. Unequal cooling rates in spite of equal sections, because of the unequal rate of heat transfer through the various zones of the mold.

6. Unequal cooling rates of a casting due to the addition of runner and riser gates and feeder heads.

7. Hindered contraction from cores, enclosed sections of a mold and frictional resistance of the mold.

#### Designing for Sound and Dimensionally Correct Castings

Without due consideration of the complexity of the volume changes in cooling cast iron, the designer will be unable to give much help to relieve the manufacturer of many of the obscure hazards of which the foundryman, himself, is not always too clear about.

With the acceptance of the oft repeated dictum counselling maximum cooperation between the producing levels, the founder is so often compelled to resort to expedient and, frequently, expensive tech-

nique in his endeavor to produce an acceptable casting. Indeed, in spite of all his resourcefulness, he is not always able to give a completely sound or dimensionally correct casting. Designers should fully recognize that designing for a best metal component involves consideration of not just the working or mechanical strength required but, also, that of thermal stresses inherent in the cooling of a metal when poured into castings of contrasting section. The contrasts, he must also remember, are emphasized by the additions made to a casting of machining allowances. The sections of a casting need careful thought if it is to be produced free from defect and true dimensionally. Sections of equal thickness, or mass, are not necessarily correct from a casting point of view since certain area of a mold will not transfer heat from the metal at the same rate as other areas.

#### Foundry Technique to Overcome Contraction Stresses

The amount of contraction and degree of final distortion, it is seen, depends upon the design and its relationship to the physical and mechanical properties of the alloy and the influences inherent in the manufacturing methods such as:

1. Design section and volume of the casting.
2. The class of alloy.
3. Melting influences on the alloy, superheat and pouring temperature.

4. Size, shape, location and distribution of the runner and riser gates and feeder heads.

5. Character of the mold and core materials.

6. The method of making the mold and cores.

7. Stress relief treatment during the cooling period and subsequent to cooling down.

1. Item one has had adequate reference made to it.

2. The class of alloy is so often a compromise because cast iron properties can vary so widely, in a casting of contrasting section. For instance, if one takes as an illustration a heavy machine tool bed, the best class of metal for the massive slideways is by no means suitable for the remaining members of the casting which it may exceed by 300 to 400 per cent. The most suitable alloy to ensure maximum freedom from porosity in the slideway will produce, in many designs, either a mottled or even white iron structure in the, comparatively, thinner walls of the casting. Such a metal would introduce acute contraction stresses and the attendant problems of warping and fracturing. Occasionally, conditions being favorable, it is possible to pour a bed casting with two classes of metal more favorable to the sections. The running arrangements are made to first pour the metal of a close-grained structure in the slideways and then to complete the filling of the mold with a much softer grade of metal through ingates located at a higher level in the mold. This arrangement also favors a reduction in distortion and, consequently in the camber allowances.

3. Metal melted at a high temperature with adequate superheat and poured into a mold at a suitably high enough temperature will allow the mold to reach a more accommodating temperature than if poured at a low temperature. Thus, contraction and distortion tendencies are reduced somewhat.

4. The size, shape and location of the ingates, riser gates and feeder heads are so often a compromise. The admission of metal at a safe point, such as a heavy lower section of the casting, will not be favorable to equalizing cooling rates or stresses. It does, indeed, increase the contrast of a section already overburdened with heat as a result of its section. In contrast, it would be more equalizing if the mold is poured through the thinner sections to even up the temperature gradient. Flow-off risers can be suitably located, but for feed headers there is often no alternative.

5 and 6. The character of the mold and core material and the manner of constructing the mold. Much can be done to reduce contraction stresses if enclosed portions of the mold and cores are constructed to yield to the contracting metal.

Iron reinforcements to the sand should be carefully watched and reduced to the minimum margin. Resort may be made to creating a looseness in certain areas of the core and mold by introducing pockets of cinders, sawdust, wood and even an easily fusible metal. It is often only possible to overcome a contraction hazard by some treatment of the casting at some stage after the metal has solidified after pouring. It may involve the release by breaking away of the enclosed mold or cored area. Or a section of the casting may be bared so that its rate of cooling is accelerated and in line with the remainder of the casting. With a certain class of steel casting the core is released immediately after solidification by injecting a stream of water into the cored area, all other methods not being rapid enough to allow the early contraction which develops. Great care is often exercised to produce a core in green sand which will be less resistant than a dried core.

7. The heat treatment of castings has been developed to accelerate the so-termed aging and stress relieving of castings so that failure from fracture and creep after machining operations is materially reduced. It is known that there are other methods of relieving stresses.

#### Test Procedure

Most of the tests were made without correlating temperature with the thermal behavior of the castings. Various types of castings were examined, but machine tool beds, boring bars and gear blank castings provided the most outstanding examples of greatest clarity.

It will be realized that the movements of a large complicated casting during the cooling down period will need to be examined carefully to avoid the many sources of error. It is not claimed that all possible sources of error have been eliminated, but such errors which may exist are not of a major character and the main conclusions are based on supportable premises.

Figures 1, 2 and 3 indicate the arrangements employed to follow the movements of the castings examined. Figure 1 shows the method first used and it was found to be most universally applicable on account of the number of points of the casting which were checked. The method illustrated in Fig. 2 provided an additional check when used in connection

with Method No. 1. Method No. 3 was designed by the author to provide automatic recording at certain points of a casting and as a check on Method No. 1. Therefore, the use of Methods 1 and 3 are responsible for the records committed to the various graphs to be referred to.

Method 1 (Fig. 1), involves following the movement of the casting by frequent measurements between fixed points outside the mold and rods held by the metal of the casting. Whenever practicable, temperature was associated with the volume changes. A lathe bed is chosen to describe the principles of the controls applied. A cross-sectional elevation, K, and two end-sections of the longitudinal elevation, M, are shown in Fig. 1.

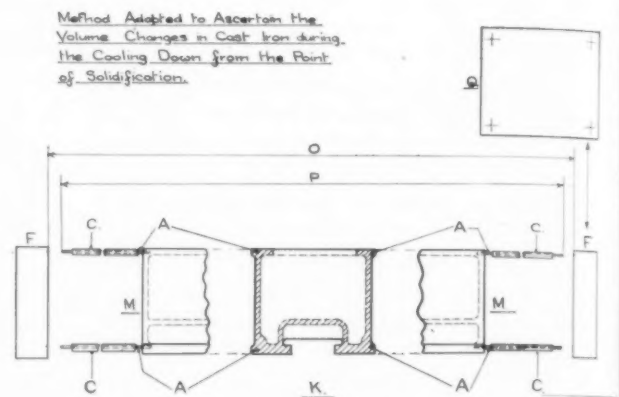


Fig. 1—Method used for ascertaining volume changes.

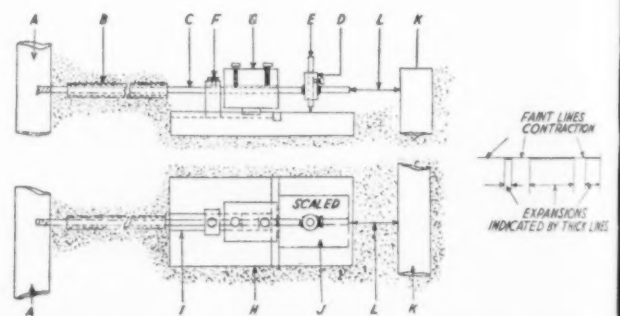


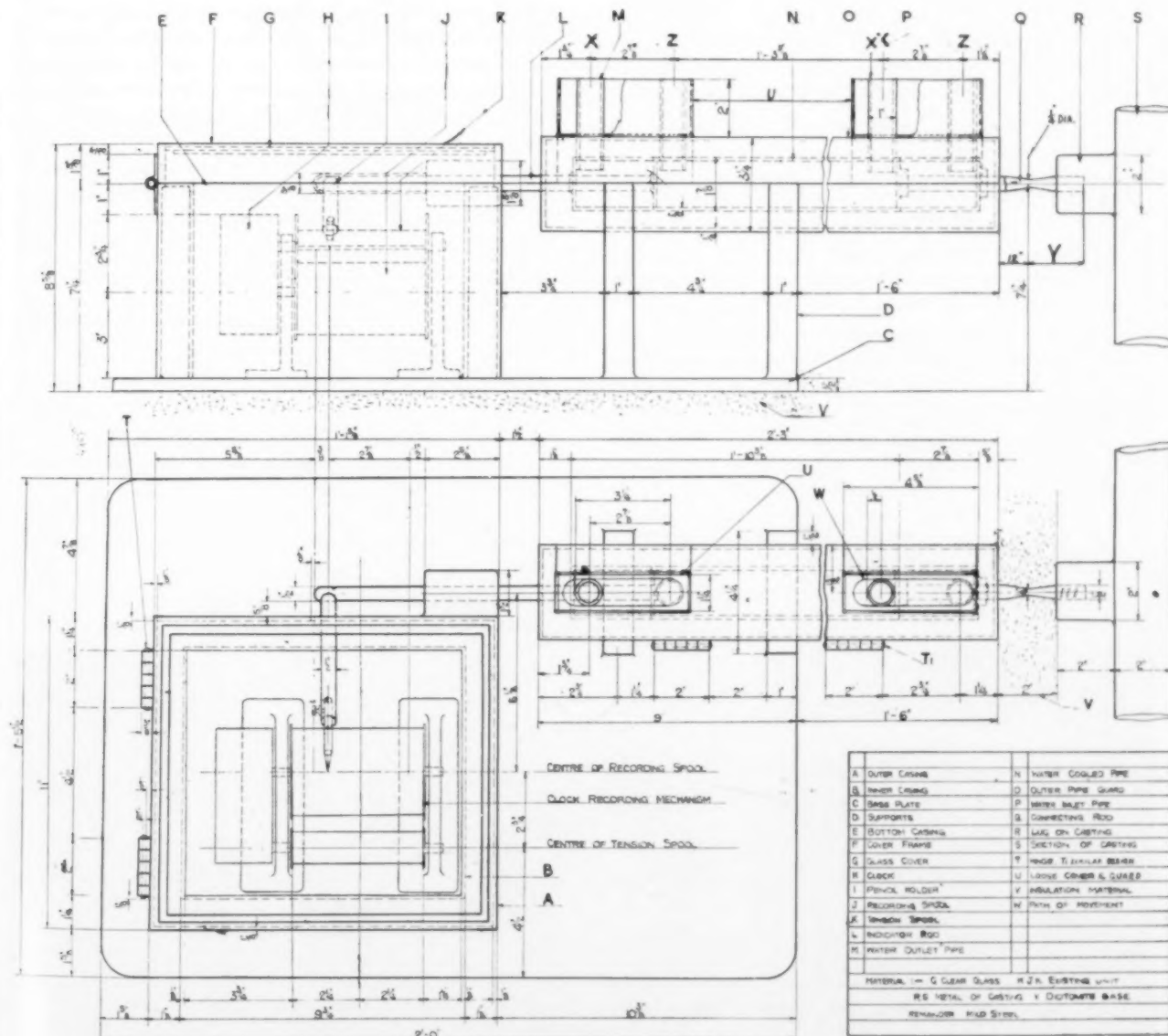
Fig. 2A (left)—Method of obtaining additional automatic record of movement of casting. Fig. 2B (right)—Example of record using method shown in Fig. 2A.

The test procedure is as follows: During the ramming of the molds, heavy weights, F, (Fig. 1) are located soundly below the floor level, outside the end of the pattern, at a suitable distance (in this example 3 ft.). Wrought-iron tubes, C, 2 ft. 8 in. in length, are also positioned and rammed around with the rest of the mold. The tubes are placed with ends within 1 in. of the vertical faces of the casting and the opposite ends about 4 in. away from the machined vertical face of the weights, F. Four tubes are located at each end of the mold, two opposite the heavy slideways and two 2 in. from the top face of the mold. Each tube is fixed opposite the vertical walls of the casting. There are, thus, eight tubes to hold eight rods.

Following the withdrawal of the pattern and the finishing operations, the molds are broken through to the tubes and  $\frac{1}{2}$ -in. wrought-iron rods, A, are pushed forward to within about 3 in. of the face of the heavy weights, F, at each end of the mold. About 1 in. of the rod, A, made jagged, to hold fast to the metal of the casting, penetrates into the mold cavity. The opposite ends, which have been ground perfectly flat, correspond with crossed center-lines scribed on a polished section of the weight opposite (see Fig. 1, F). The spaces between the ends of the tubes and the mold face are well sealed and secured by ramming sand around the rods, now leaving, as before stated, the ends of the rods protruding into the mold cavity, to be gripped by the metal of the casting.

Just before pouring the casting, careful measurements are taken, by inside micrometers, of the gap between the ends of the rods, and the centers marked on the weights. Provision is made to begin taking measurements within 3 min. after pouring the mold.

Fig. 3—Contraction recording mechanism.



In the case of a large casting, similar to the bedplate illustrated in Figs. 6 and 7, repeated measurements proceed at intervals of 3 min. for about 5 hr., when the measurements are slowed down to every 15 min. for a further 5 to 6 hr., then every half hour for about 6 hr., every 3 hr for 24 hr., and checking is continued four times daily until contraction is complete. The total contraction is shown by the final gap distance between the faces, P, of the rods A, and face, O, of the weights, F, less the original gap dimensions before pouring the casting. The sum of the net dimension at each end equals contraction.

The alterations between the points, O, P, indicate, with a closing of the gap, expansion, and, with a widening of the gap, contraction. The records of the movement of the casting, with associated times, are plotted on graphs. The early behavior of the casting is expressed in minutes and the complete movement in hours.

A line, zero, on the graphs identifies the condition of the molten metal in the mold. The penetration of the curves below the zero line indicates expansion beyond the size of the mold. Above the zero line the



curves register contraction and any further volume change, such as expansion or arrests. Note is taken of the expansion of the test rods due to absorption of heat from the casting. The temperature variations of the rods are checked at intervals. Later, the same class and length of rod is heated to a similar temperature as obtained during the cooling of the casting. The length of the rod, cold, as against the rod heated to the particular temperature experienced, gives the amount of extension of the rod, which must be cancelled from the early movements of the casting as recorded. A further adjustment is needed in castings which are cambered. On the total contraction, an allowance is made for the difference in length that a curved line will give when straightened out.

The possibility of employing a metal with a low coefficient of expansion instead of mild steel rods was considered necessary. An alloy of nickel and iron, of similar composition to invar-metal, was tried but it was abandoned because the temperatures to be contended with went behind the range at which the alloy reactions caused a cancellation of volume increase.

It was also found that the alloy test rod became embrittled, resulting in partial fracture at the juncture where the rod entered the metal of the casting, indicating intercrystalline failure. This failure would be due to subjecting the nickel alloy to high temperature of the face of the molten cast iron of the casting.

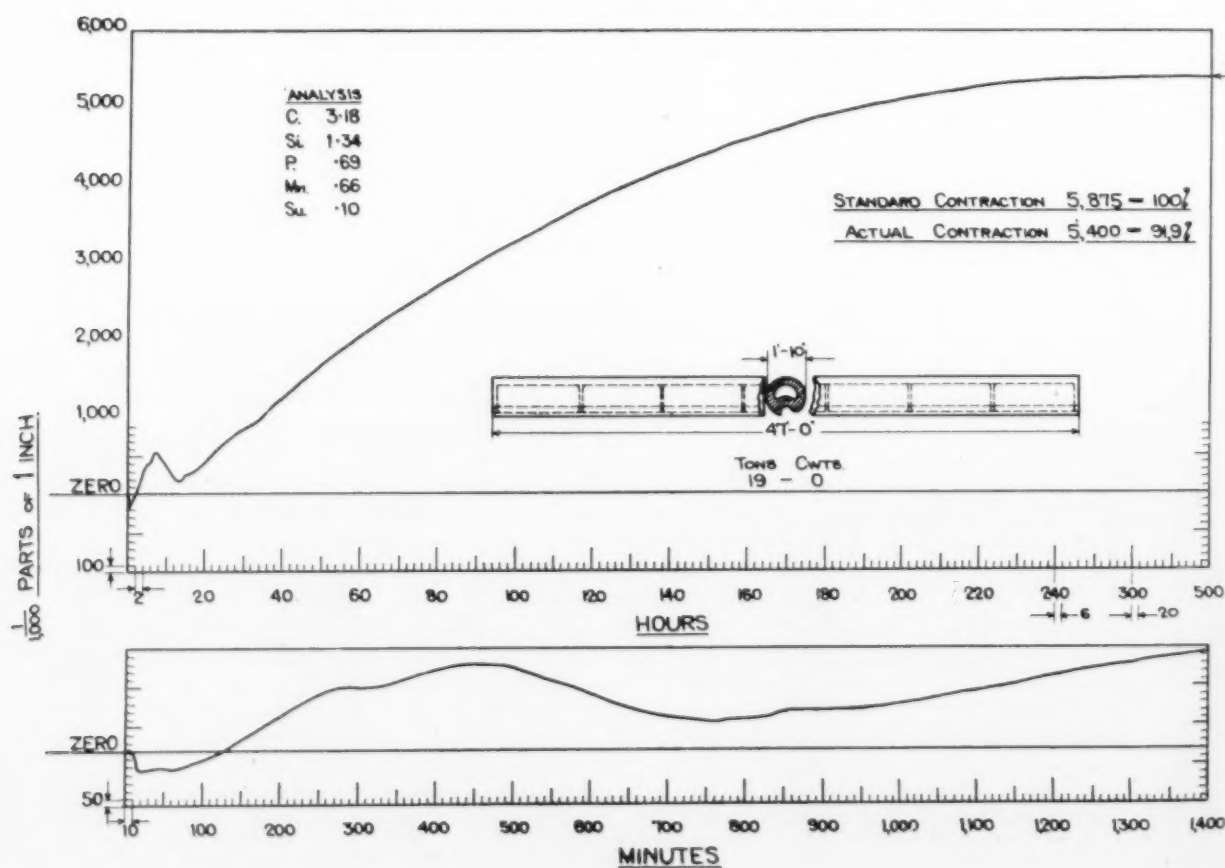
The peculiarities of invar and similar alloys are that, while steel expands and contracts normally, the special alloy shows little or no volume increase over a restricted and comparatively low range of temperature. This range of temperature appears to correspond with the magnetic transformation period, which is complete at about 328 F (220 C), depending upon the composition of the alloy. Thus the volume change is neutralized by the magnetic transformation. But at lower and higher temperatures both expansion and contraction are experienced.

It was, therefore, decided to revert to the use of mild steel rods, which could be relied upon to give the consistency of a regular expansion rate, according to the temperature gradient, which could be identified and allowed for when committing figures to the curves of the graph.

The sketch (Fig. 2) illustrates the second scheme adopted:—A, casting; B, rod tube; C, rod for recording movements of the casting; D, socket holder for the black lead pencil, E, which is welded to the rod C; E, lead pencil; F, slide, attached to the rod, C; G, weight, fastened to the rod, C; H, baseplate; I, slide-way for slide, F; J, recording paper, gummed to the baseplate, H; K, the weight for measuring the variations in the gap, L; L, gap which discloses expansion and contraction movements as described in the paper.

The movements of the casting were recorded by the pencil on the paper. When the pencil moves away from the mold an expansion is recorded, and when towards the mold, contraction. Since the pencil must traverse a section of the penciled line more than once, however, because of the expansion and contraction

Fig. 4—Volume changes of the cored boring bar shown in Fig. 5.



phenomena, the changes could be detected only by the density of the penciled line, along with the attention paid by the person recording the times of the changes on the paper on which the measuring scale was drawn. An example of a record so obtained is shown in Fig. 2 B.

The method outlined was used on several occasions to check the general method of following the volume changes, as explained in Fig. 1.

Figure 3, illustrates the automatic recording mechanism specially designed to take care of extraneous thermal influences and to provide a check on Method 1 which was almost universally applied. As described in detail in Fig. 3, the instrument consists of automatic clock recording mechanism housed in a strong fabricated steel casing welded to a substantial steel base and stress annealed. The pencil holder is shown at I (Fig. 3), attached to the indicator rod, L, which is in turn attached by welding to the water-cooled pipe, N. The indicator rod, Q, on the casting, S, side is also welded to the water cooled pipe N. The short section of the indicator rod, Y, is welded to the main indicator rod, Q, after locating it in the lug, R, of the casting, S. The path of movement is indicated by the slots W<sup>1</sup>, and, again, the position XX before the casting movement during contraction and ZZ on the final cooling down of the casting. The diameter of the rod is constricted at the juncture Q,V, to reduce the transfer of heat from casting to indicator rod. This section is also cooled by a stream of air from a compressed air line.

#### Selected Earlier Tests

Before outlining more recent research it will be helpful to refer to previous studies<sup>2</sup> on the subject of contraction and distortion. Two outstanding examples are selected, illustrated in Figs. 4 and 5, a machine boring bar, 47 ft. in length, weight 19 tons, and a roll grinding machine bed, 41 ft.-6 in. length, weight 13 tons, and shown in Figs. 6 and 7<sup>2</sup>.

#### Cored Boring Bar

A test, applied to a cored boring bar, 22 in. in diameter and 47 ft. in length, and weighing 19 tons, confirmed the tests applied to an 18-in. solid bar made

previously. Figure 4 shows the design and dimensions of the bar, and the cooling behavior of the casting by the curves plotted on the graphs, the casting is shown in Fig. 5. The composition of the metal is set out in Fig. 4. The hours' curve shows the completion of the cooling process of the casting. After cooling down to atmospheric temperature, the total contraction of the casting was 5,400 thousandths against 5,875 thousandths for the standard allowance. Contraction equaled 91.9 per cent of standard allowance. The casting had taken up 94 per cent of its contraction in 10 days.

#### Test on Long Bed Castings

Figures 6 and 7 illustrate the cooling behavior of a 41 ft.-6 in. length bed casting for a roll-grinding machine. The design and dimensions of the bed are outlined by the inset in Fig. 7. A cross-sectional elevation of the bed, across the points, AA, of the portion of the longitudinal elevation, L, is shown at E.

The casting is poured in the reverse position to that which is shown in the sketches, Fig. 7, with the slideways on the bottom face of the mold. The cooling behavior is followed and checked, as previously described. In this case, however, four curves were formed. All allowances have been made in the curves for the expansion of the rods due to heating, and also for the curvature of the casting consequent upon the camber necessary to produce a straight casting. Camber is required downwards (in the center as poured) and sideways (in the center) in the direction indicated by the arrow, K, in Fig. 7.

The early volume changes of the casting are expressed in minutes in the graph, Fig. 6, and the complete movements, in hours, are shown in Fig. 7. The letters A, B, C, and D identify the curves taken of the cooling casting, on the lines indicated by A, B, C and D in the sketch E, Fig. 7. The line of contraction, A, carries the heaviest mass of the casting—the vee slideway. The line, B, holds the flat slideway, being the second heaviest line of contraction. The line, C, influenced by the additional metal at H, thereby takes the third position in order of mass influence, and D lightest, most quickly cooled and contracting line.

The standard contraction allowance for a 41 ft.-6 in. casting is 5,188 thousandths. The actual final contraction of the bed is as follows:—

Fig. 5—Cored boring bar weighing 19 tons.



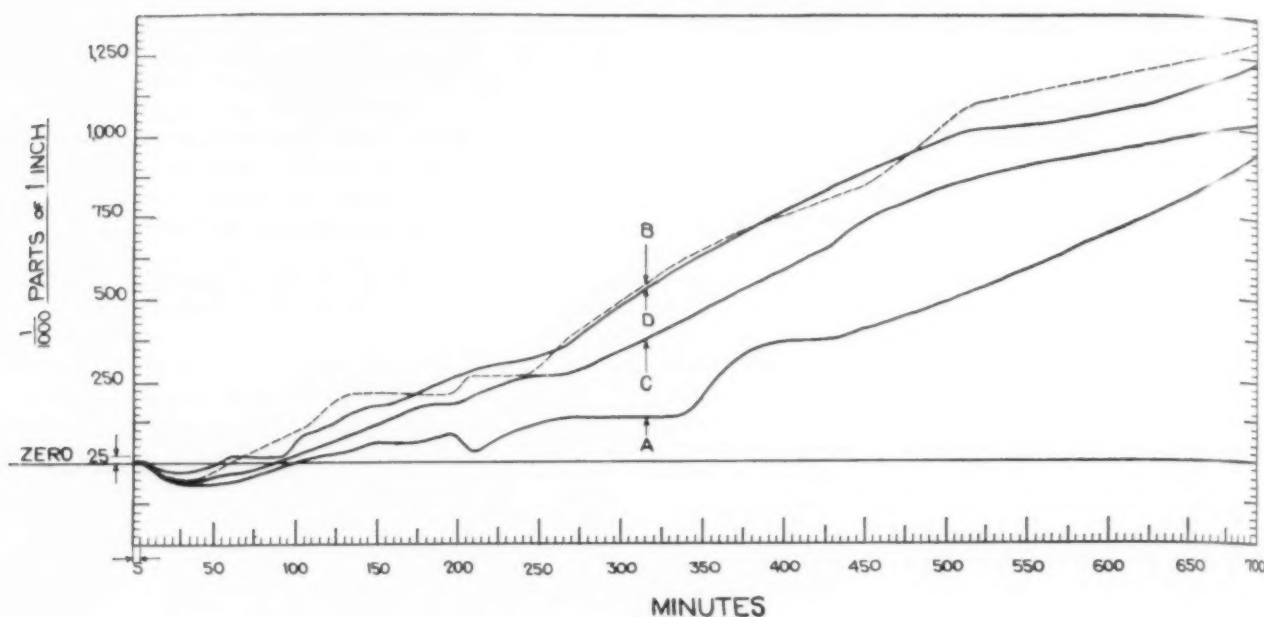
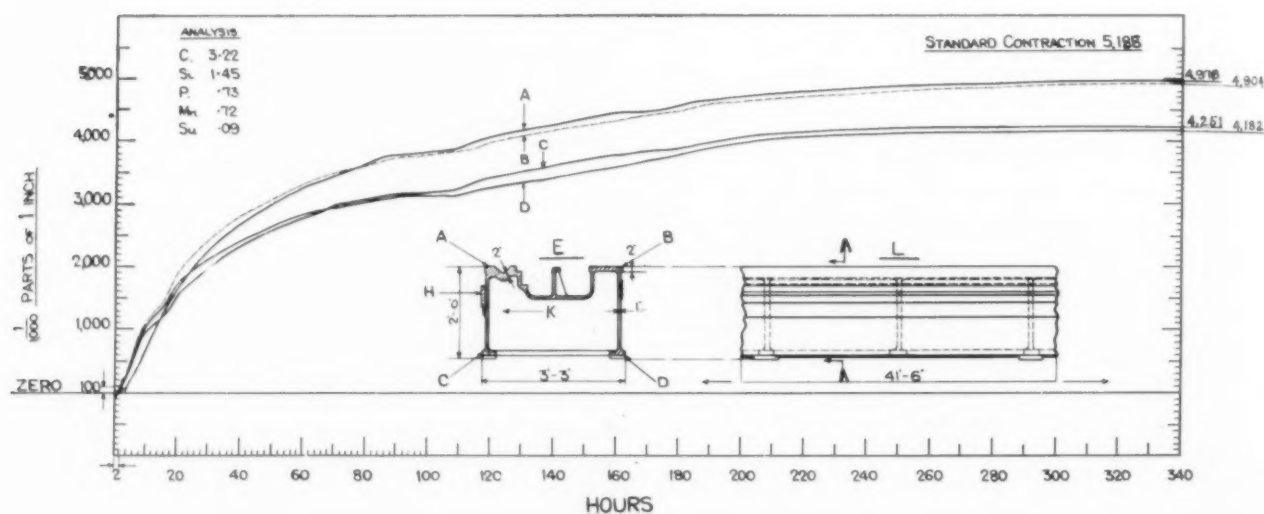


Fig. 6—Cooling behavior of a 41-ft. 6-in. roll grinding machine bed.

Fig. 7—Completed cooling behavior of the roll grinding machine bed.



A. 4,978 thousandths, which is 96.0 per cent of the standard allowance.

B. 4,904 thousandths, which is 84.6 per cent of the standard allowance.

C. 4,251 thousandths, which is 82.0 per cent of the standard allowance.

D. 4,182 thousandths, which is 81.0 per cent of the standard allowance.

Average contraction is 87.4 per cent of the standard allowance. Between the lowest contraction of D and the highest contraction of A there is a difference of 796 thousandths. A, C side has contracted an average of 72 thousandths more than B, D side. A (bottom heavy vee slideways) has contracted 74 thousandths more than B (bottom flat slideway). A, bottom, has contracted 727 thousandths more than C, top. B, bottom, has contracted 722 thousandths more than D,

top. The average contraction of A,B, (bottom face) over C,D (top face) is 725 thousandths.

#### Camber Allowances

These varied contractions, recorded on the four extremes of the castings, confirm the camber allowances needed to counteract distortion due to the differing heat gradients and hindrance, by design, to free contraction. The camber allowance, downwards, was  $2\frac{3}{4}$  in. which must be associated with the excess contraction of 725 thousandths of A,B (bottom face) over C,D (top face). The lateral camber requirements on the A,C (heavy side) was  $\frac{5}{8}$  in. This allowance must be associated with an excess contraction of 72 thousandths of the A,C side over the contraction of the opposite side, B,D.



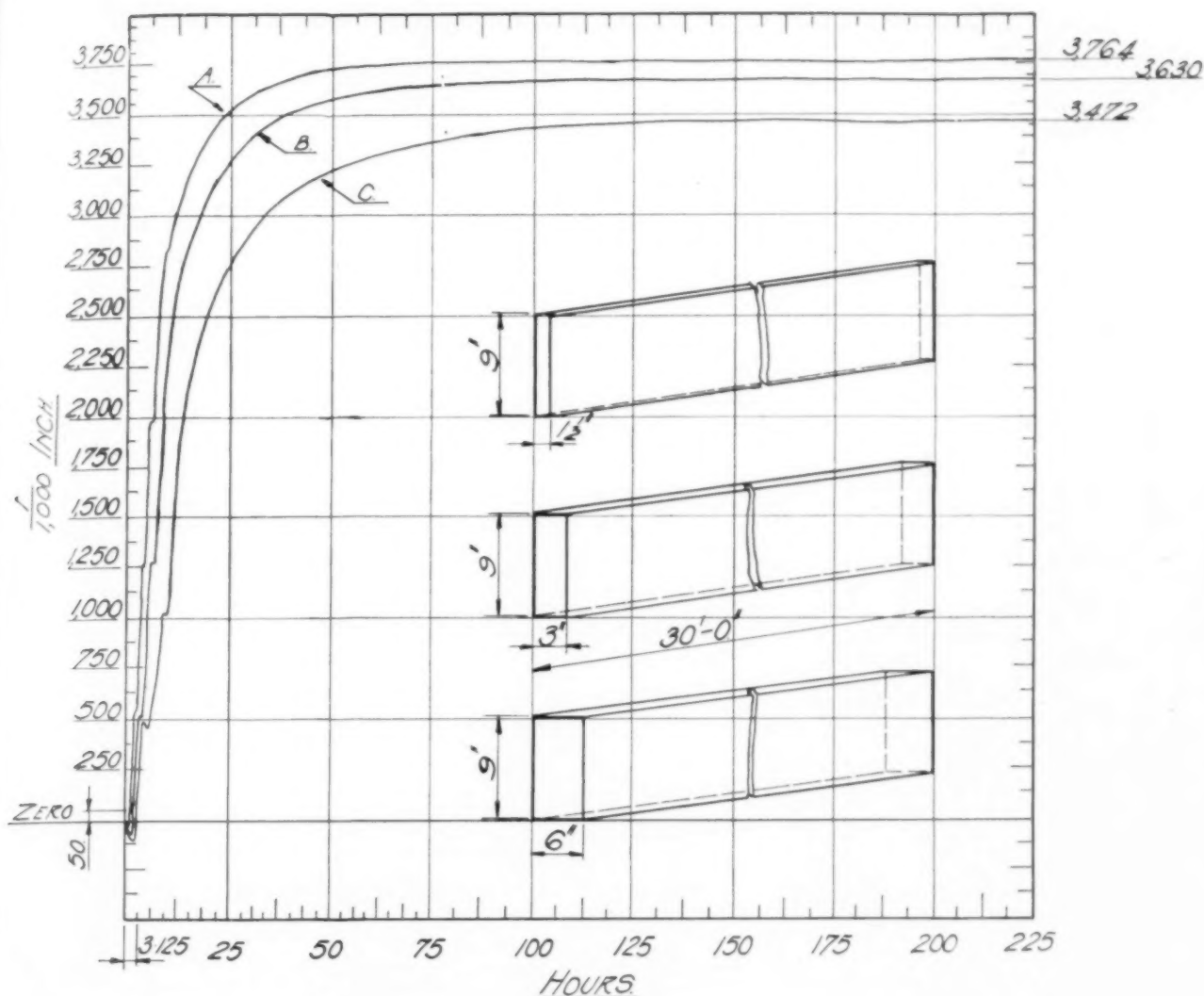


Fig. 8—Thermal history of simple gray iron bar castings, extended to embrace almost 10 days.

Without the heavy facing, H, on the A,C side it is necessary to allow approximately  $\frac{1}{4}$  in. more camber on the A (heavy vee slideway) line than on the B (flat slideway) line. The influence of the heavy side facing, H, was to reduce slightly the downward camber, but to create, along with the heavy vee bottom slideway, a side distortion if not counteracted by camber allowance.

It will be noted, on close examination of the cooling curves, that B (bottom flat slideway) contracted slightly ahead of C (top with side facing) and then later falls behind to take its place in order of mass cooling. This may be due to the early effect of the denseners, on the B slideway, and the acceleration of the cooling rate immediately after solidification of the metal. The C (top side), however, is influenced by a heavy facing, carrying nearly as much metal as B (bottom) which is also densened. Since the side facing is subjected to a sharper cooling rate by virtue of its outward position in the mold, cooling and contraction is, subsequently, more rapid than B, which is located in the bottom of the mold. Again, much of the heat

absorbed by the denseners is retained by the densener to slow up the rate of cooling at a later period.

Adverting to the arrests in the cooling curves, it is possible that the influences, when cooling has reached a stage when a considerable difference in contraction stress is upon the casting, cause a seemingly late arrest which is not actually due to an expansion, but may be attributed to distortion of the casting.

#### Cooling Behavior of Simple Bar Castings

The graphs shown in Figs. 8, 9, 10 and 10A, illustrate the cooling behavior of large uniform section bars in gray iron and white iron. Figures 8 and 9 indicate the thermal reactions of gray iron and Figs. 10 and 10A, white iron. The analysis of the gray iron bars is similar to the metal of a roll grinding machine bed, Fig. 11, to be described later.

The dimensions of the bars are as follows:—

$1\frac{1}{2}$ in. x 9 in. x 30 ft.	3 in. x 9 in. x 30 ft.
3 in. x 9 in. x 30 ft.	6 in. x 9 in. x 30 ft.
6 in. x 9 in. x 30 ft.	

#### Analysis

T.C. .... 3.23	T.C. .... 2.95
Si .... 1.46	G.C. .... trace in C
P .... .062	Si .... .021
Mn .... .078	P .... .022
S .... .009	Mn .... .051
	S .... .012

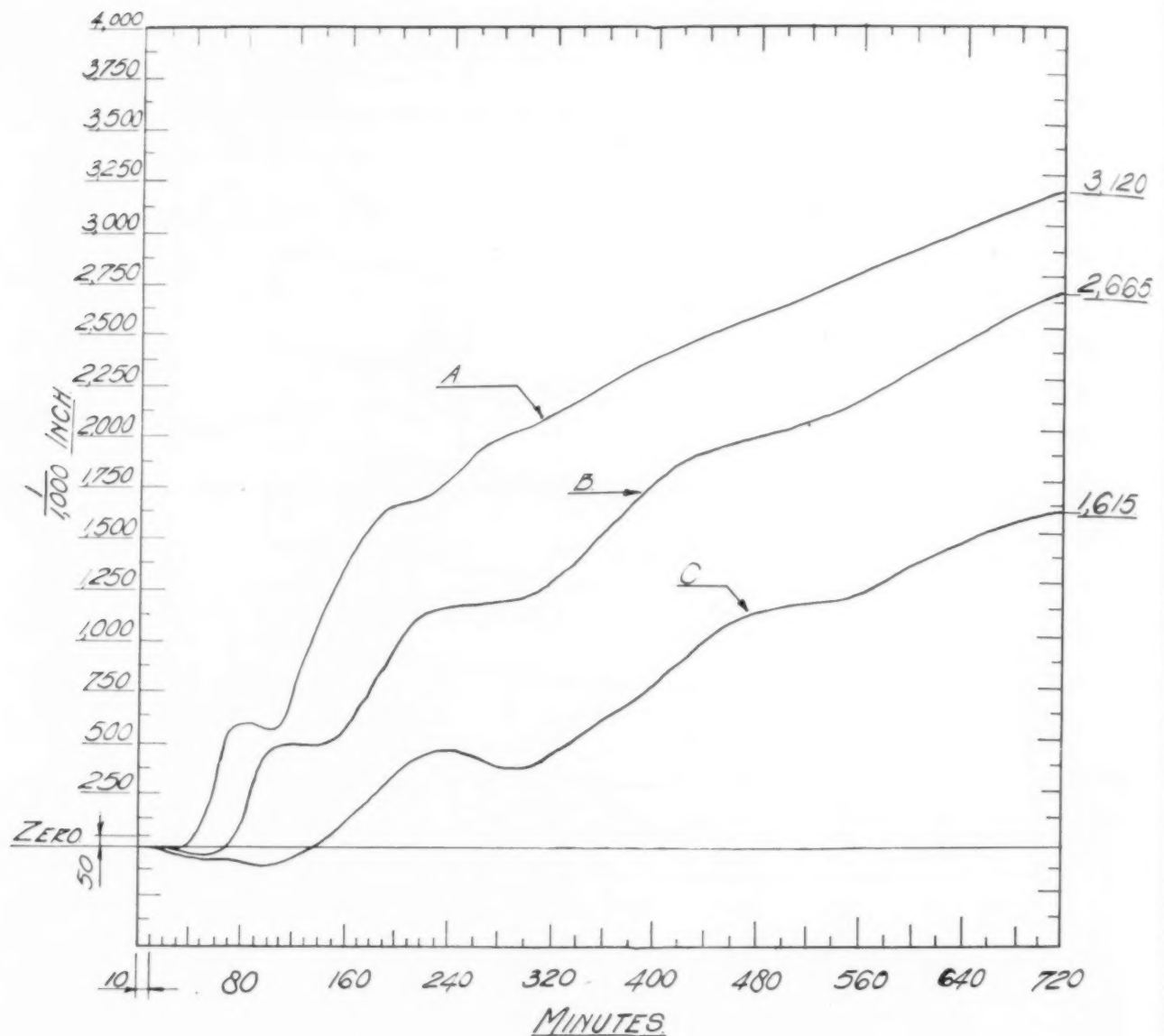


Fig. 9—Thermal history of simple gray iron bar castings of increasing section.

TABLE I—ARRESTS AND EXPANSIONS DURING COOLING  
(Minutes after Pouring)

	Gray Iron			White Iron	
	1½ in.	3 in.	6 in.	3 in.	6 in.
First arrest	8 to 25	10 to 55	14 to 100	Not detected	
Second "	65-110	100-145	180-300	240-340	330 slight
Third "	190-220	200-295	370-430	Not detected	

#### Comparative Amount and Rate of Contraction

Gray Iron			
Rate of Contraction — Total Contraction		1/1000 in.	% Standard
Hours			
A. 1½ in. bar in 70		3,764	= 100.4
B. 3 in. bar in 120		3,630	= 96.8
C. 6 in. bar in 165		3,472	= 92.6
White Iron			
B. 3 in. bar in 80		6,324	= 84.3
C. 6 in. bar in 135		6,286	= 83.8

#### Roll Grinding Machine Bed

The roll grinding machine bed, illustrated in Fig. 11 presents more than the average problem if a satisfactory casting is to be obtained. It will be noted, by reference to Figs. 12 and 13, that the four slideways, A, B, C and D vary in section compared with each other and against the vertical walls and plated areas between the slideways. Further, the bed is stepped which introduces another complication. The casting is poured with the slideway face down in the mold with metal of the following analysis:—T.C., 3.23; Si, 1.46; P, 0.62; Mn, 0.78; S, 0.09.

The principal complications of the design are:—

1. The great difficulty in selecting a metal which will produce the necessary close-grained wear-resistant structure and freedom from local porosity about the hot spots on the slideway surface. Such a metal which would suit the metallurgical requirements of the slideway would create a hard brittle metal in the light sections of the casting, increasing the danger of excessive distortion and fracture. Hence a densening technique is applied to the comparatively heavy slideway

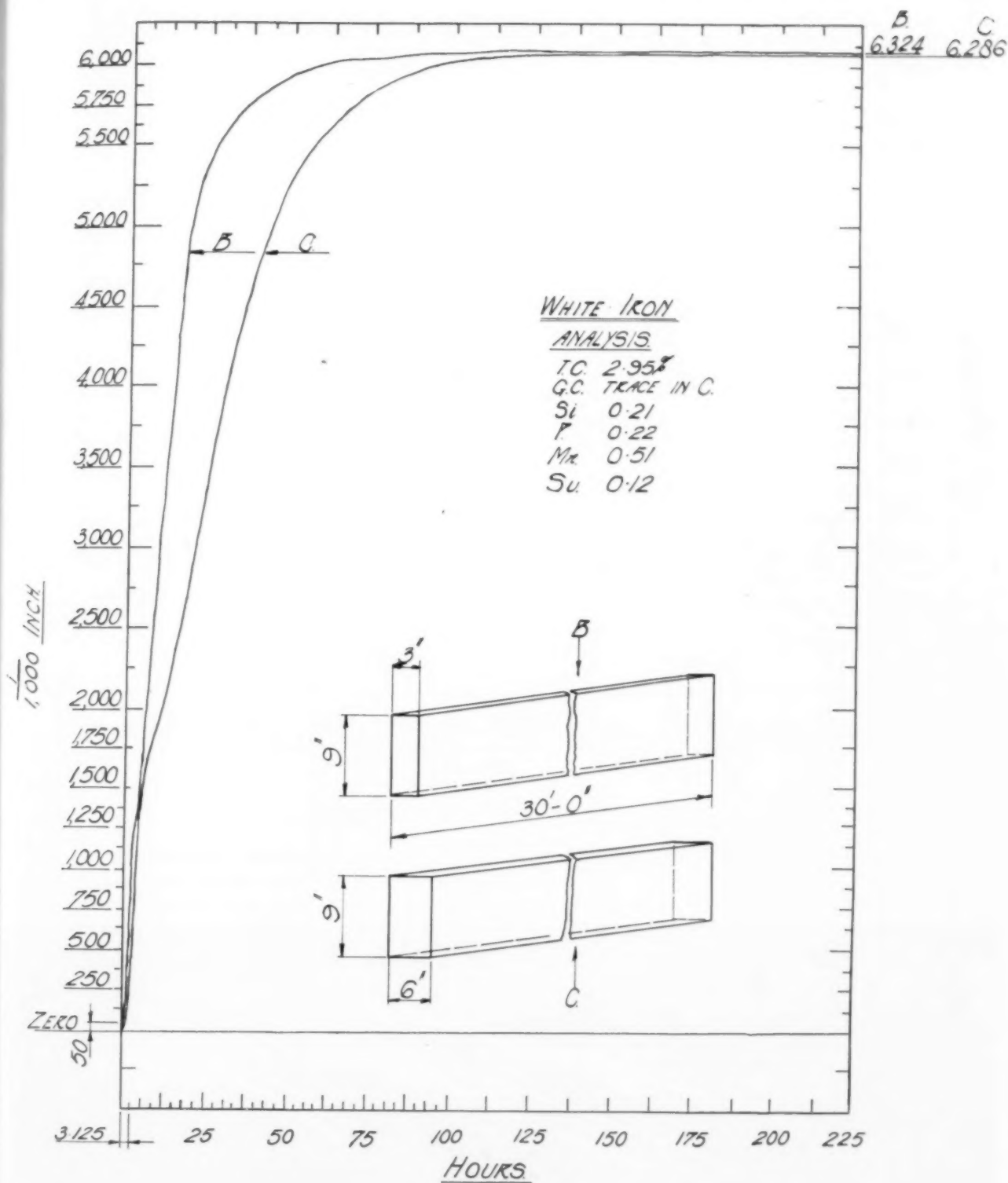


Fig. 10—Thermal history of simple white iron bar castings.

permitting the use of a grade of metal which favors a gray iron structure in the light sections of the casting. This may be considered as a compromise so far as metal is concerned.

2. The varying sections of the slideways demands varying counterbalancing cambers.

3. The differing depths of the bed introduce the need for varying camber allowances.

4. Differing rates of heat transfer from the central areas of the lower face of the casting, as poured, will, inevitably, create stressed material due to strain set up by the heat gradient and the consequent distortion and, possibly fractures. It will be noted on examining the sketch, Fig. 12, that the slideways B, and C, are of more massive section than A and D. Without much



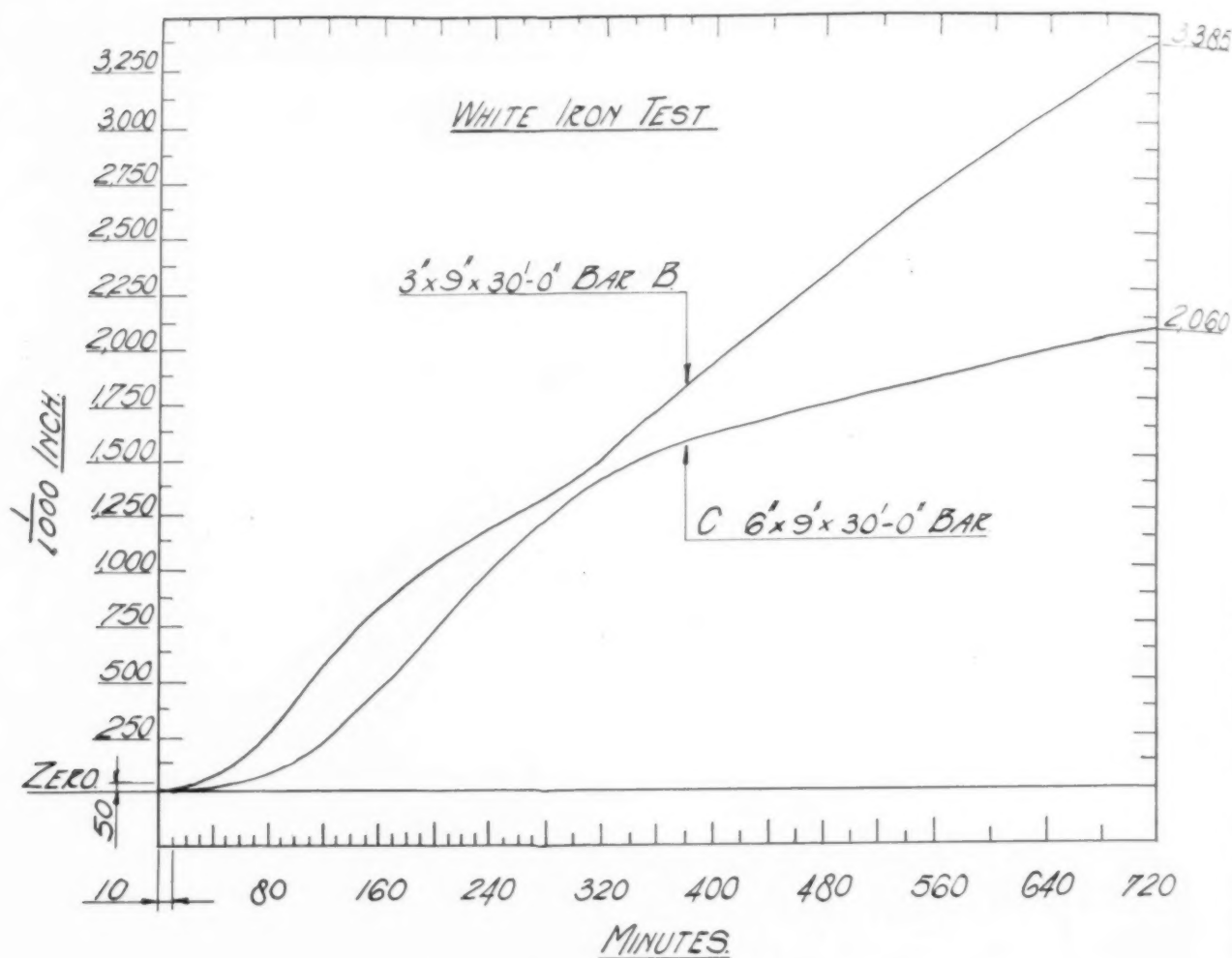
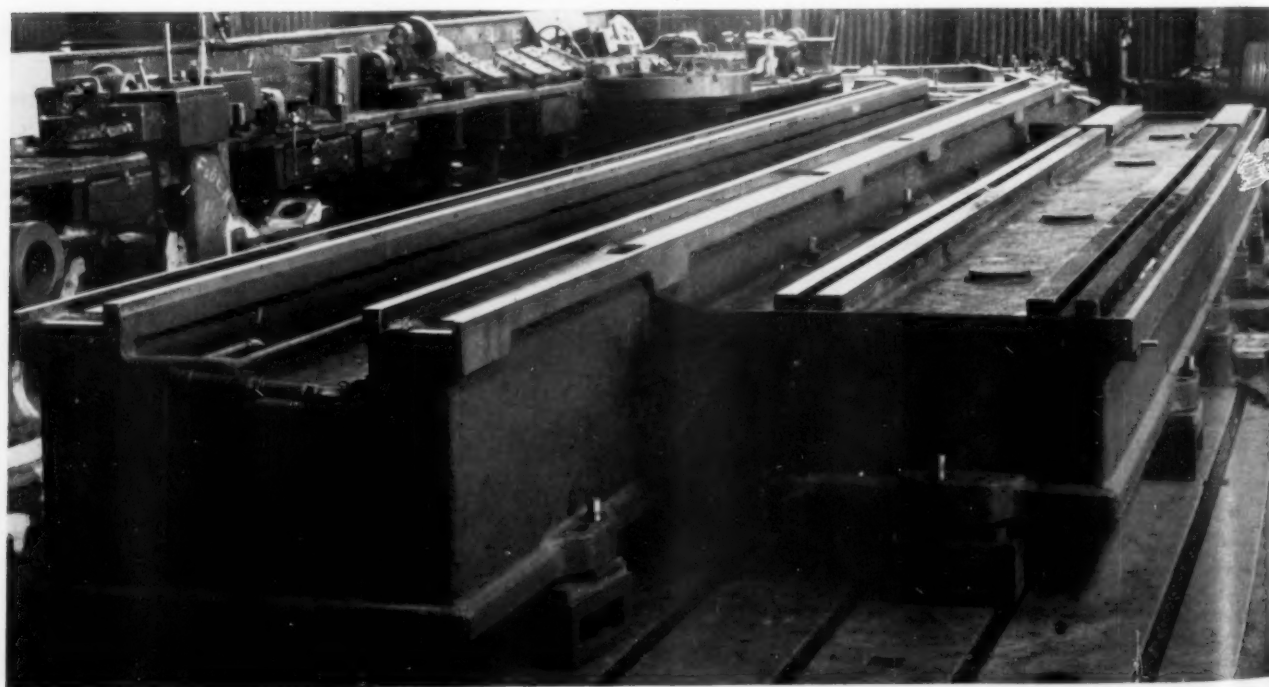


Fig. 10A—Thermal history of simple white iron bar castings.

Fig. 11—Roll grinding machine bed.  
(By courtesy of Craven Bros. Ltd., Machine Tool Engineers, England)



	Inches			
	A	B	C	D
Previously allowed	1.35	1.50	1.25	1.18
Calculated on air cooling	0.90	1.00	0.70	0.75
Actual recorded—air cooled	1.12	1.14	0.75	0.70

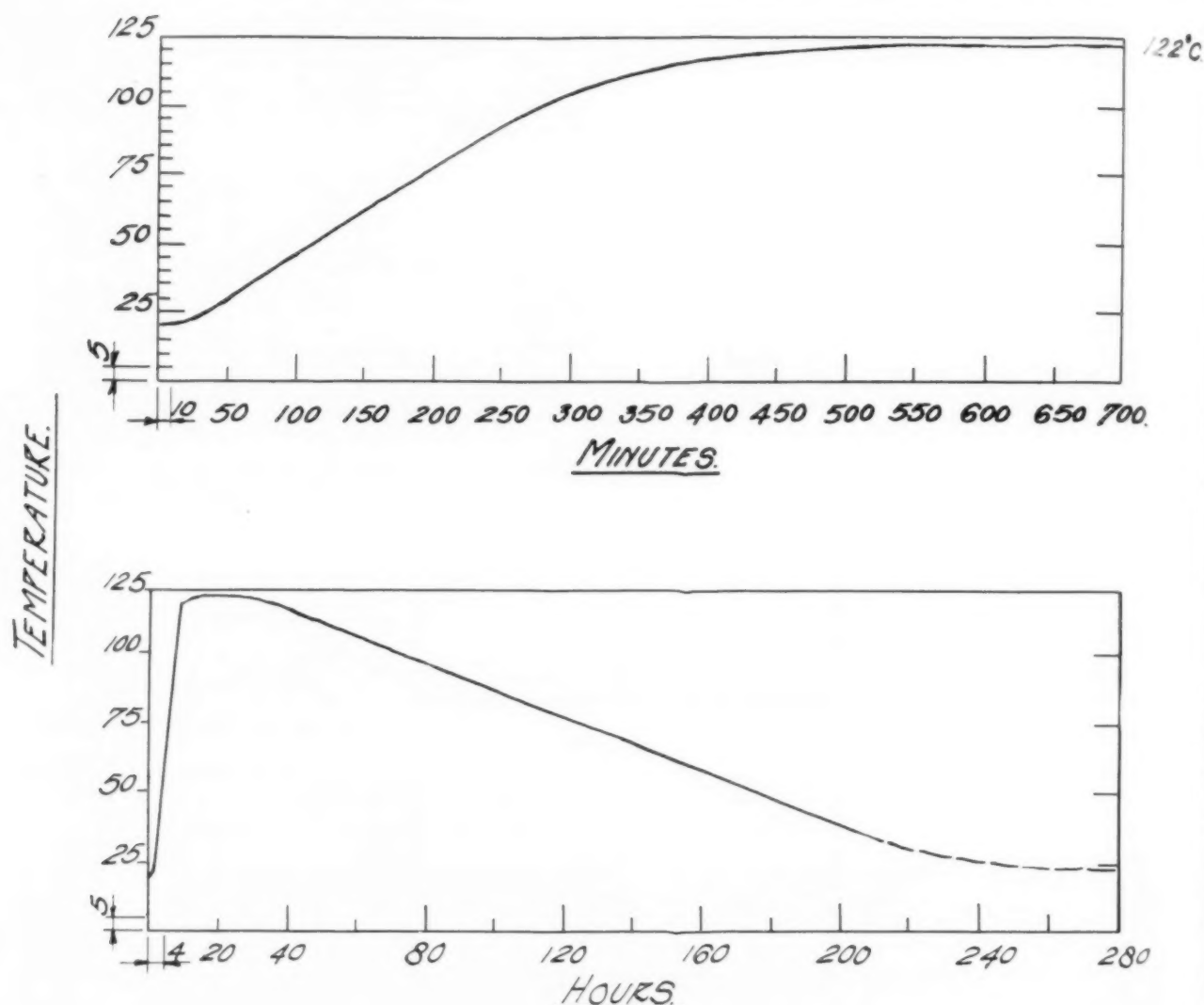


Fig. 14—Temperature of exhaust air admitted to air cooling system.

The system of air cooling, as illustrated in Figs. 12 and 13, show that air pipes are located (during the mold building) longitudinally, below the surfaces of the two middle slideways, B and C, and extend for 12 ft. of the central areas which normally constitute the most slowly cooled areas of the casting. The inlet and outlet pipes are located at right angles to the slideway coolants.

It will be noted that the air pipes are in immediate contact with the metal of the denseners (high conductivity). Within 5 min. of the cease pouring operation air is admitted to the air-cooling system. The exhaust air temperature was checked at intervals and recorded on the graph (Fig. 14). Recordings of air temperatures were suspended after 212 hr.

The thermal behavior of the roll grinding machine bed casting, with air cooled slideways, is shown in Figs. 15 to 18. It will be noted that the movement of the casting was checked at 16 points, four on the lower face of the casting and four on the upper face, at each end of the casting as sketched in the inset Fig. 15 and

16. The movements were followed by the set-up outlined in Fig. 1 except that the automatic recording mechanism, illustrated in Fig. 3, was used at one end of the casting at the point D. The automatic check confirmed the records obtained at the opposite end of the casting on the same line of slideway.

An attempt was made to correlate temperature with the transformation points during the cooling of cast iron and to register bending moments and the critical stress periods. This was only partially successful.

In view of its limited practical application, it was decided to check the temperature at only four central points on the outward and accessible lines, A and D on the lower face and E and H on the upper face of the casting. A portable pyrometer was employed the thermocouples being inserted in holes arranged in the sides of the slideways, A and D and in the sides of the base reinforced flange, sections E and H. In spite of careful sealing of the thermocouple guide tubes leading to the face of the casting a considerable lag in recording the temperatures was experienced. After a number of false readings, it was found necessary to allow a soaking time of 15 min. with each reading taken. The temperature readings, therefore, cannot be correct to within 50 C. It may be stated, however,



122°C

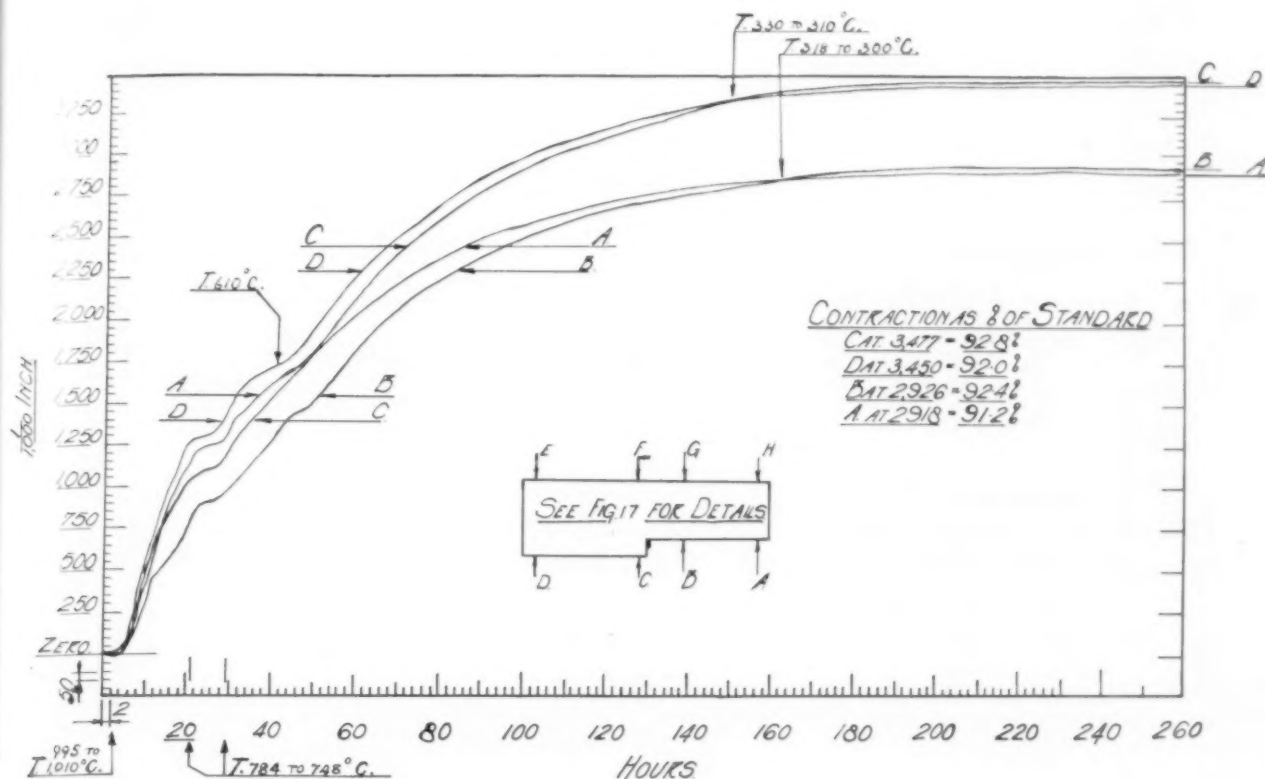
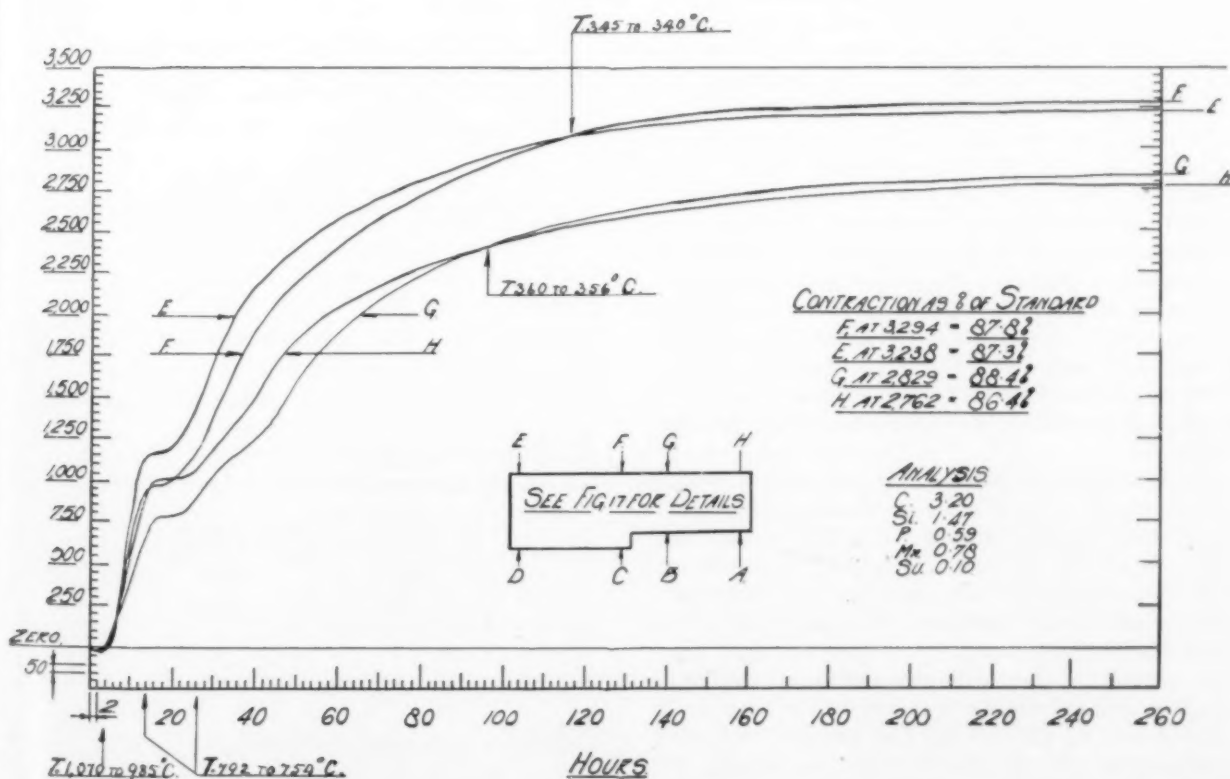


Fig. 15—Thermal behavior of roll grinding machine bed casting with air cooled slideways.

Fig. 16—Thermal behavior of roll grinding machine bed casting with air cooled slideways.



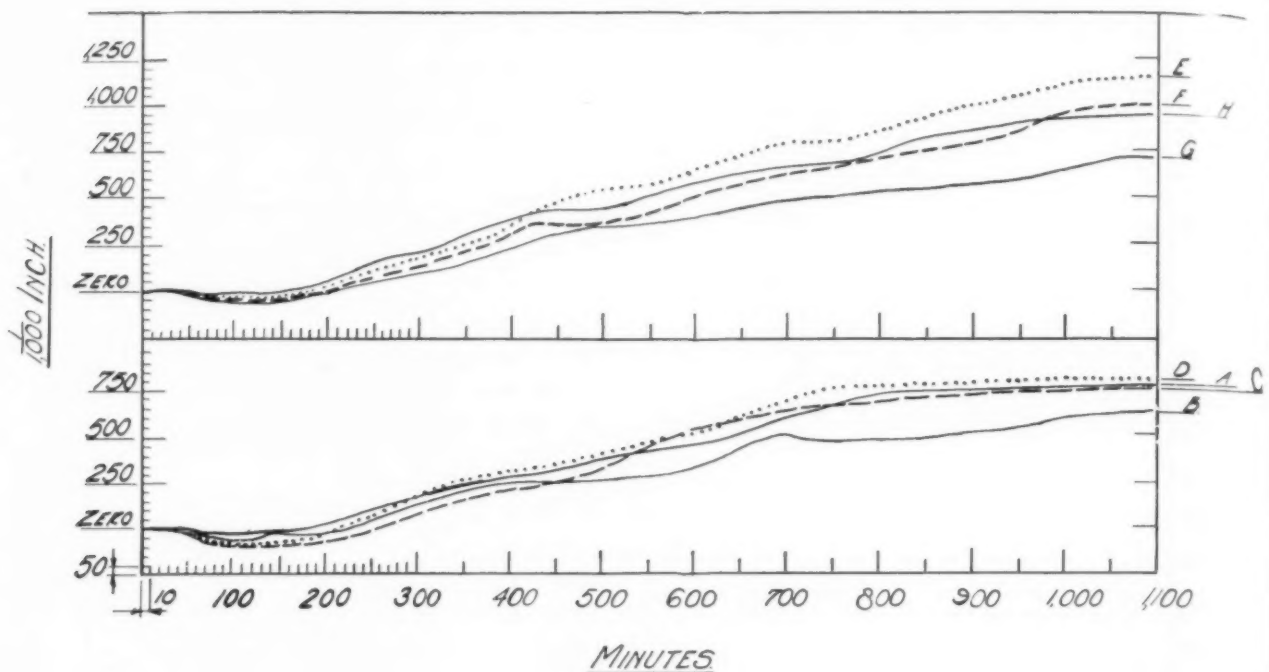


Fig. 17 (below) and Fig. 18 (above)—Thermal behavior of roll grinding machine bed casting with air cooled slideways.

that with the subsequent care exercised in repeating the conditions of checking that the records are comparative over the period of observation. An examination of the graphs Figs. 15 and 18 indicate a number of critical points during the cooling of the casting.

At periods varying from 10 to 15 min. after pouring, on the line A, an expansion is detected lasting for about 3 hr., on B for 3¾ hr., C for 4 hr. and D 3½ hr. A second outstanding transformation period will be noted ranging on the various lines from 18 to 30 hr. In 32 to 40 hr. A and D lines indicate a third critical change. In 42 to 46 hr., B shows a third arrest, but the line C did not indicate any very clear change.

Following the curves for the lines E, F, G and H arrests occur in 10 min. to 2½ hr. Second arrests show very clearly in from 14 to 26 hr. A third change can only be slightly detected on the lines G and H in between 34 to 46 hr.

Contraction on the line C (Fig. 15), overtakes and exceeds D at about 146 hr. B reaches A and exceeds it in 160 hr.

At about 88 hr. the line G (Fig. 16), crosses H and it continues to exceed it in contraction until final cooling. After 116 hr. the line F crosses E and continues to exceed it in contraction until final cooling.

It is interesting to examine the approximate temperatures recorded and indicated on the graphs, Figs. 15 and 16.

The records of the final contractions on the eight lines show that the thicker sections on the lines A, B, C and D exceed that of the less massive and more rapidly cooled areas on the lines E, F, G, and H.

Figures 19 and 20 show a comparison of curves between the bed air cooled and a previous record taken

of a casting without air cooling on the line C. The air-cooled bed is seen to cool and contract much more rapidly than the bed not air cooled.

#### Large Test Castings

Figures 21, 22 and 23 illustrate a series of test castings designed to provide further precise information in respect to the influence of contrasting depths and sections on contraction and distortion. The depths, A, B, C, varied at 4 in., 8 in., and 12 in. respectively. Figure 21 shows a plan of the test casting and cross-sectional elevations are indicated in Figs. 21, 22 and 23. The metal on the bottom face of the castings varied as at J, K, and L. The upper face varied as

TABLE 3—TEST RESULTS

	Depth 4 in.			Depth 8 in.			Depth 12 in.		
Convex	A	B	C	A	B	C	A	B	C
Distortion	0.05	0.05	0.02	0.03	0.01	0.46	Nil	0.01	0.03
Contraction 1/1,000 in.	1,258	1,260	1,255	1,254	1,247	1,246	1,247	1,240	1,236
Average Contraction	1,257			1,249			1,241		

TABLE 4—TEST RESULTS

	Depth 4 in.			Depth 8 in.			Depth 12 in.		
Concave	A	B	C	A	B	C	A	B	C
Distortion	Nil	.005	.002	Nil	.001	0.10	Nil	Nil	0.01
Contraction 1/1,000 in.	1,254	1,254	1,253	1,250	1,249	1,246	1,242	1,239	1,238
Average Contraction	1,254			1,248			1,239		

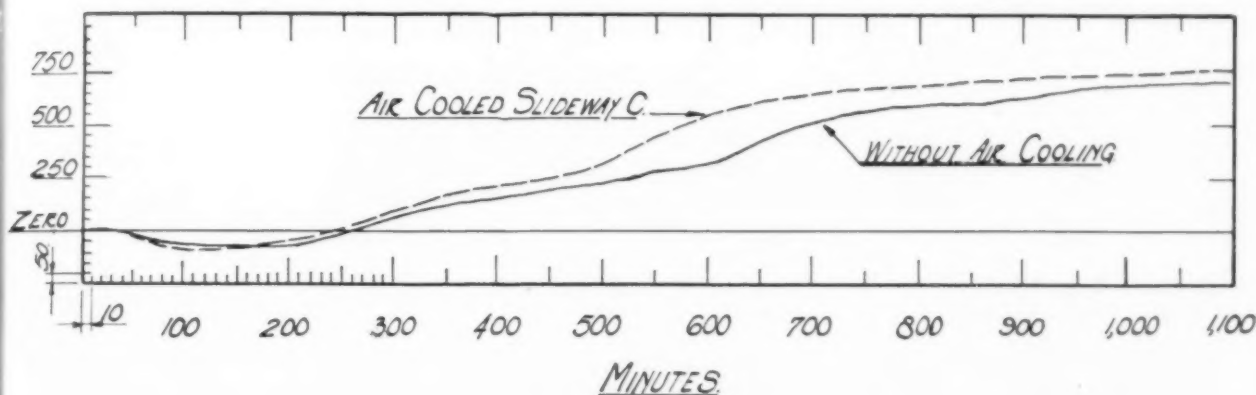
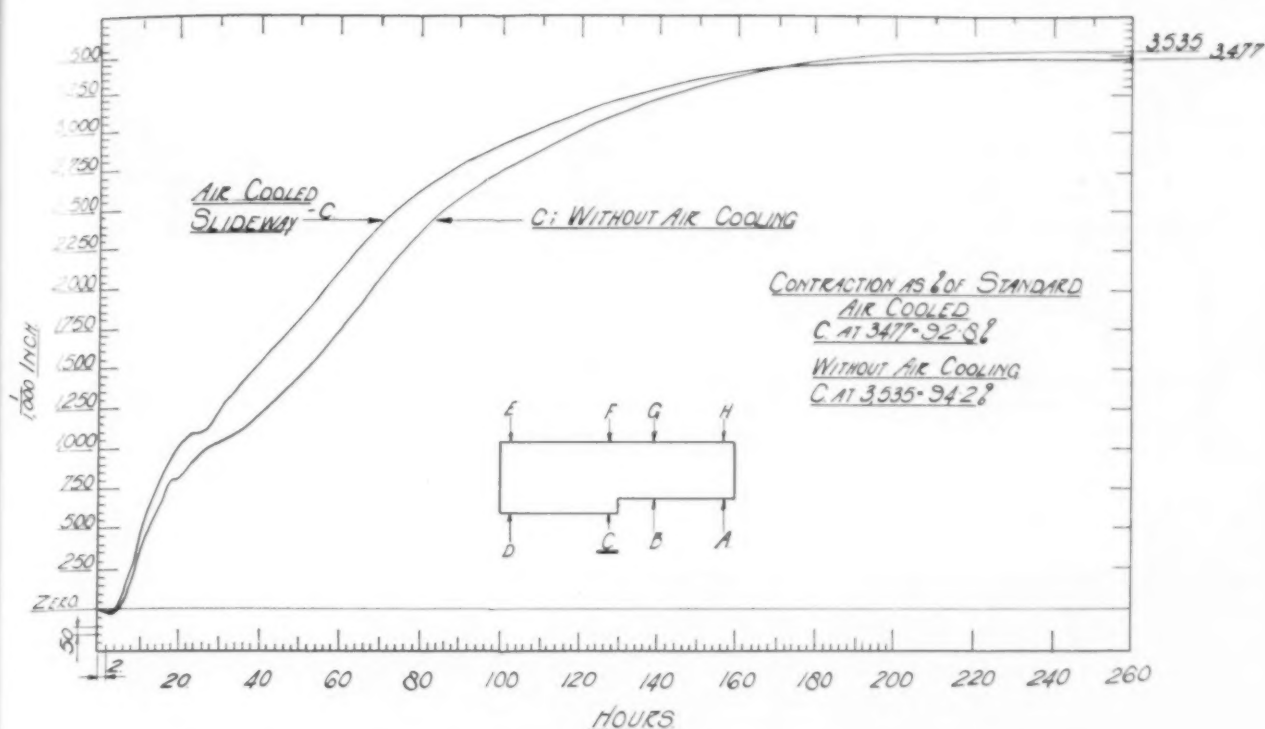


Fig. 19 (above) and Fig. 20 (below)—Comparison of curves between the bed air cooled and without air cooling.

shown in Fig. 23 at M and N and on one side wall as identified by D, E and F.

The analysis of the metal, at T.C. 3.31, Si 2.13, P 0.70, Mn 0.64 and S 0.10, showed a carbon and silicon content arranged to yield a metal structure as might be obtained in more massive castings with lower percentage of the two elements. The results of the tests are shown in Tables 3 and 4.

The influence of the added metal to one of the vertical walls of the castings, shown at D, E and F, Fig. 23, are as follows:

D—concave sideways 0.30, E—concave sideways 2.20, F—concave sideways 0.05.

It will now be clear that the heavier sections of the class of casting referred to contract last and, in taking up the shorter length, bend the weaker members of the casting, but the resistance offered by the shape of

the casting in turn restricts the shortening of the thick members which bends inward—thus the thick member becomes concave. The amount of bend we may now convert to camber required—a bending in the opposite direction.

As a check on contraction, two test bars, 10 ft. in length were poured in metal of similar analysis as used for the test castings, Fig. 21, 22 and 23, with the following recorded:—

10 ft. x 6 in. x  $\frac{1}{2}$  in. contraction 1,410 thousands parts 1 in.

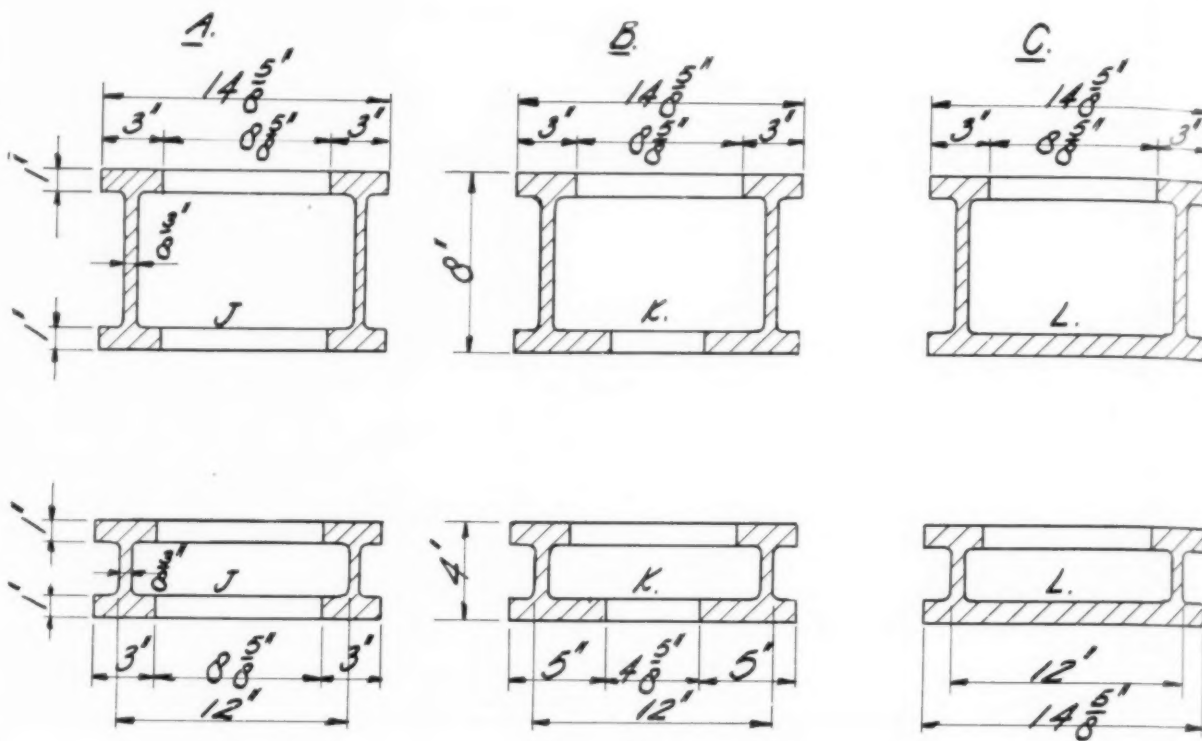
10 ft. x 6 in. x 1 in. contraction 1,385 thousands parts 1 in.

It is shown that contraction in the cored castings, outlined in Figs. 21, 22 and 23, is hindered when compared with the simple bar castings which have more freedom to contract.

#### Camber Graph

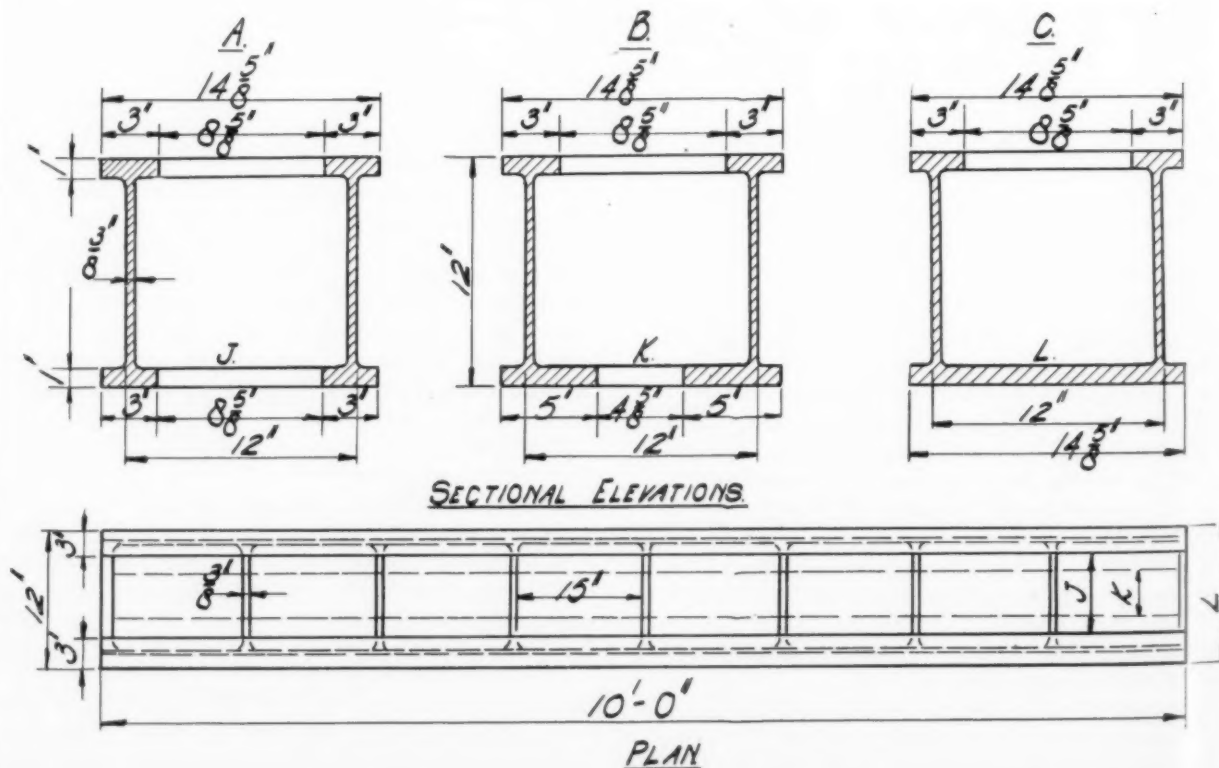
The camber graph, Fig. 24, has been designed to serve a large range of castings within the classification of beds, bedplates, tables and similar castings. The

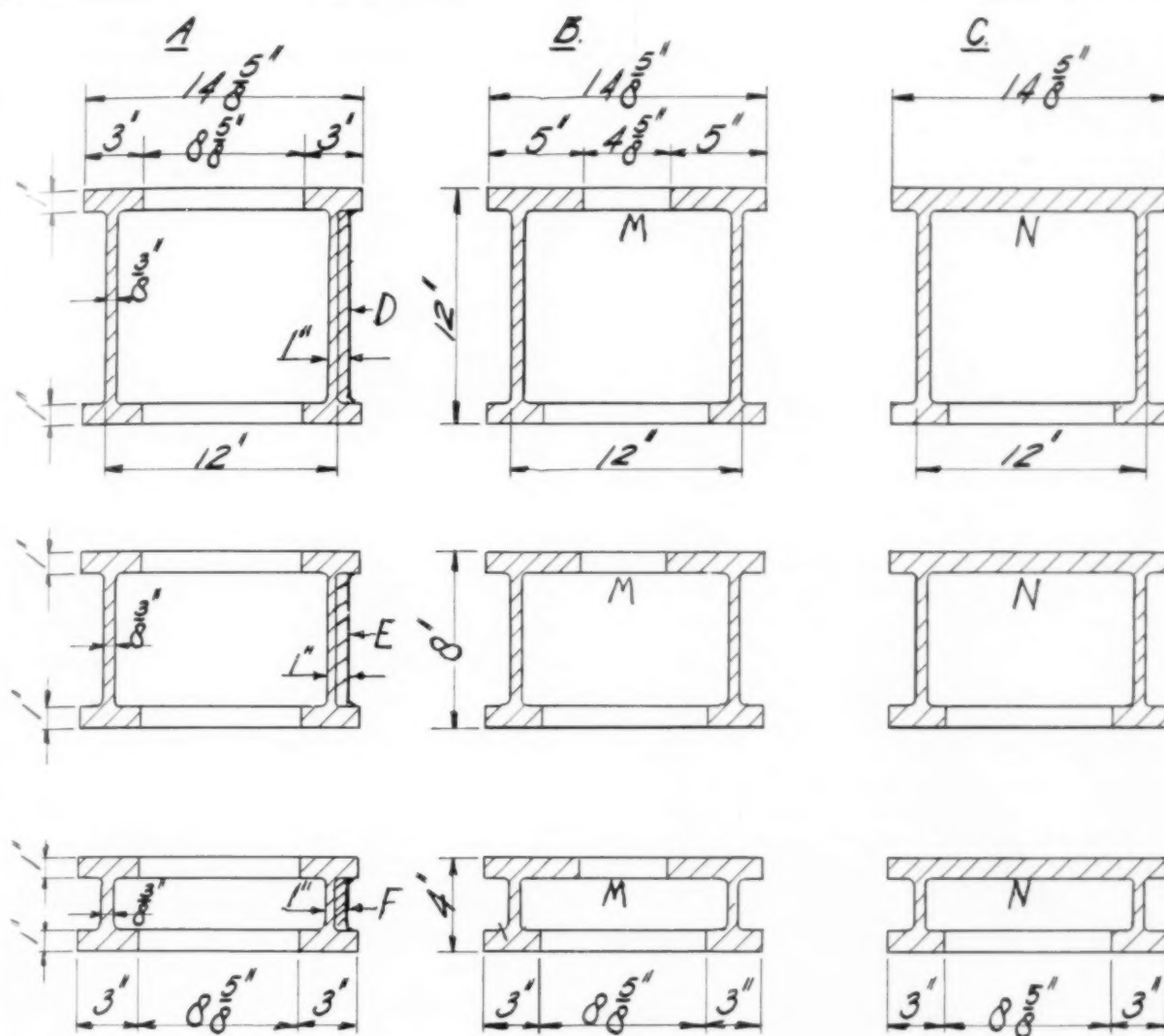




### SECTIONAL ELEVATIONS.

Fig. 21 (above) and Fig. 22 (below)—Test castings designed to provide further information on influence of contrasting depths and sections on contraction and distortion.





### SECTIONAL ELEVATIONS.

Fig. 23—Test castings designed to provide further information on influence of contrasting depths and sections on contraction and distortion.

example of the four-slideway machine tool bed, illustrated in Fig. 25, is taken as the basis of the graph. The camber requirements will vary little if the casting be a two- or three-slideway bed so long as there is a corresponding reduction in the width of the casting.

If the data obtained from the test castings, Figs. 21, 22 and 23, is studied, along with the generalizations on camber, it is possible to employ the graph, Fig. 24, to calculate the camber needed for castings carrying increments of metal on the lower or upper faces of the castings as poured.

It will be noted that the graph is arranged in two divisions. The left-hand divisions take care of the length and vertical sectional thicknesses of the casting and the right-hand division the depths and camber scale.

The graph is based on the main vertical longitudinal wall section of 1.0 in. to 1.1 in. and the lower faces (as poured) carrying metal sections equivalent to between 250 to 300 per cent of the vertical sections and covering from 30 to 50 per cent of the area of the bottom face of the casting, as poured. Any reduction in the vertical wall section calls for an increase in camber allowances and a thickening of the wall sections require a corresponding reduction in camber allowances.

The curves on the graph are plotted from information and records obtained from a large variety of castings produced over a long period of years. The scatter points were not wide and the graph is workable within very reasonable and safe limits.

At the lower end of the scale, it will be noted that

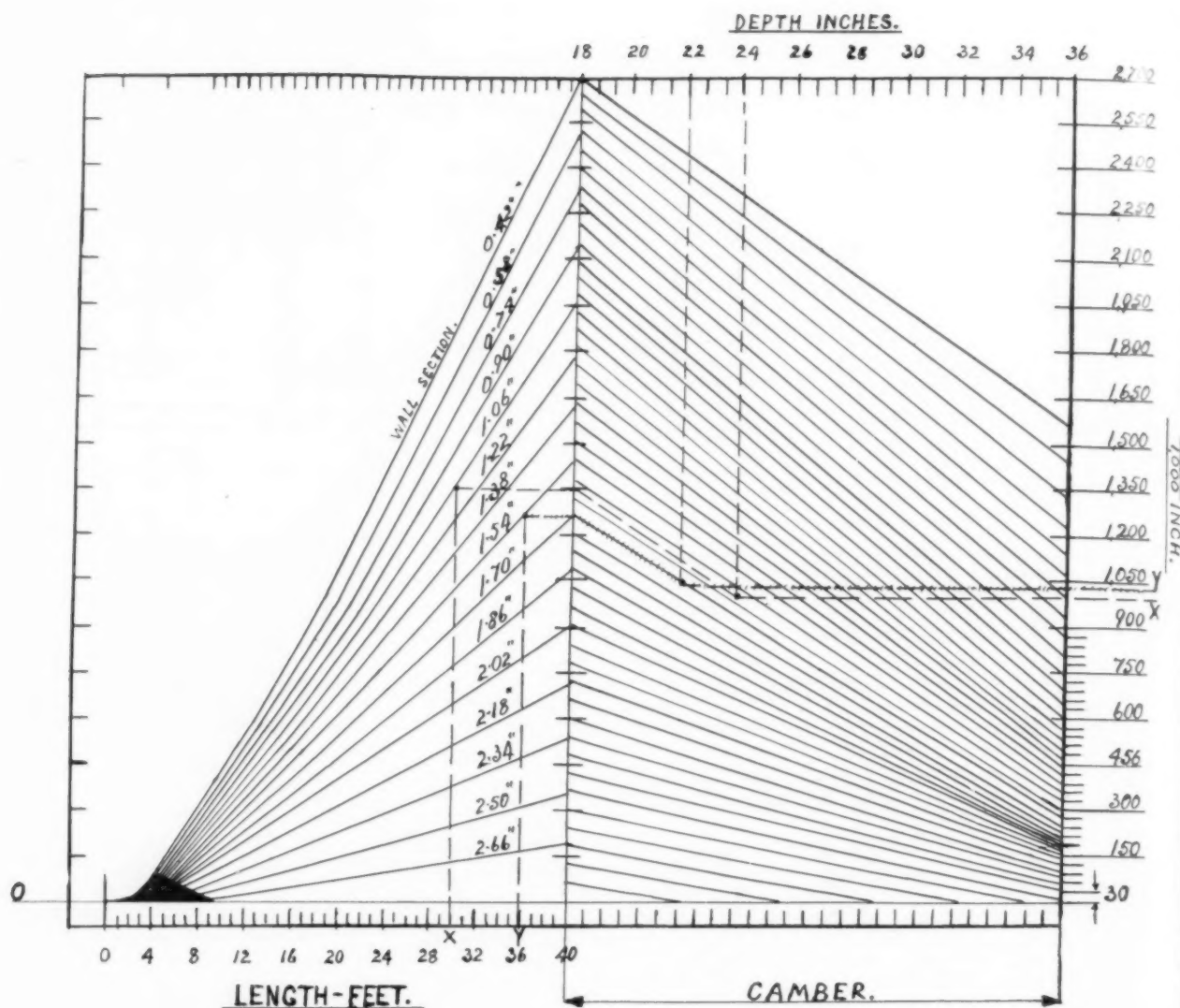


Fig. 24—Camber graph.

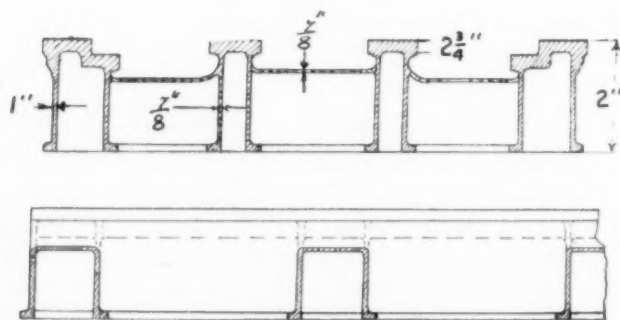


Fig. 25—Lathe bed casting.

camber requirements can be progressively ignored as length diminishes, especially about the shaded areas.

Camber requirements are not just an automatic increase with an increase in length. An examination of the graph (Fig. 24) will show that there is an acceleration in the rate of increase with increasing length.

It will be realized that from the thinnest vertical sectional walled castings to the thickest sectioned cast-

ings there will be adjustments made in the composition of the metal to suit the main sections. So that when the vertical wall sections reach a similar dimension as, say, a slideway the problem of the choice of suitable metal is of easier solution than in the case of the casting shown in Fig. 25.

#### How to Use the Graph

**Example 1:** Find the camber required for a bed 30 ft. in length, 24 in. in depth and with 1.06 in. vertical wall sections. From X, 30 ft., strike a vertical line until it joins the wall section of 1.06 in., from this point draw a line at right angles until it meets, or is guided by, the first diagonal camber line and trace it down until it joins the vertical line descending from the 24 in. depth point.

**Answer:** Camber needed is 0.99 in.

**Example 2:** Find the camber needed for a bed 36 ft. x 22 in. x 1.54 in. vertical wall sections. Starting at Y, 36 ft. the line ascends to the wall section line 1.54 in., then it crosses to meet the first diagonal camber line intersecting at the point of 22 in. depth line.

**Answer:** Camber needed is 1.02 in.



## Generalizations on Camber

The following additional rules may be used in conjunction with the camber graph, Fig. 24:—

Camber allowance, other dimensions remaining constant, increases with an increase in the length of a casting.

Camber allowance increases with a reduction in metal section of the vertical walls of the casting.

Camber allowances are reduced with an increase in depth of the casting, without any corresponding increase in the length.

Camber allowance is slightly decreased with a reduction in width.

Camber allowances increase with an increase in the plated area on the lower face of the casting as poured.

Camber allowances decrease with an increase in the plated area on the top side of the casting.

Camber allowances are modified slightly with pouring temperature, allowances increasing slightly with lower temperatures.

## Steel and Iron Compared

Figure 26 shows a sketch of a typical gear blank casting, 9 ft. in diameter, which may be made in steel or cast iron. The casting consists, essentially, of a boss,

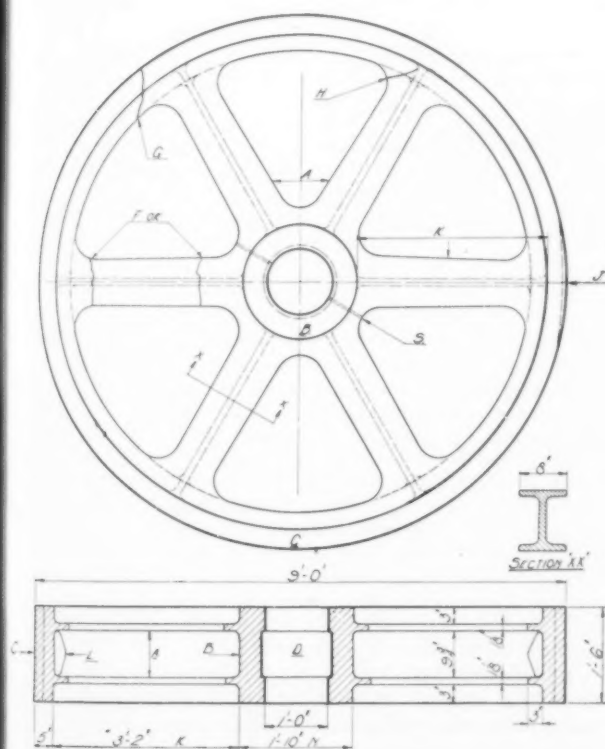


Fig. 26—Gear blank casting.

B, arms, A, and rim, C, from a study of which it will be noted that the cooling rates of the members of the casting vary considerably. The order of cooling will be, first the arms, secondly the rim and lastly the boss. Such a combination of sections introduces the hazards of fractures in the arms and possibly in the rim. The margin of safety is not always of sufficient magnitude to avoid the occasional defectives which appear when

variations of design, mold and metal may combine to produce a failure in the casting.

The stresses present in a gear casting when carrying a large heavy boss may be explained by an examination of the sketch, Fig. 26. The type of fractures indicated by F, G and H may occur in either steel or cast iron castings, but most frequently with cast iron unless certain expedients are employed. The differences in behavior of the two metals would appear to be due to the constitutional differences in the two alloys. Cast iron is comparatively brittle and can be much stressed from the dual expansion and contraction strains which are operating probably simultaneously in the same casting. The stresses in steel are of more direct or one-way effect. The contraction on the arm, K, at 0.4 in. is of a higher percentage to the total contraction of the casting when made in cast iron than it is with a contraction of 0.7 in. with a steel casting.

The force required to break the arm if cast in gray iron, with a section as shown at XX, Fig. 26, will be of the order of 450 tons tensile. To break a steel casting of similar section it will require about 1,150 tons tensile stress. Steel will respond to elongation. Cast iron has little or no property of elongation. It will be realized that the contraction stresses are of a high order in such cases referred to.

The effective strength and elasticity of steel results in a flattening of the rim opposite the arm end as indicated at J in Fig. 26 instead of a fracture, in this case the arm has had sufficient strength to pull in the rim to accommodate its contraction. In a 9-ft. diameter blank casting this flattening may exceed  $\frac{1}{4}$  in. in depth from the radius line. Subsequent stress annealing stabilizes this distortion of the shape of the casting. With cast iron the strength of the arm is insufficient to bend the rim to accommodate its contraction if the heat gradient contrasts as stated in the example.

The royal remedy to casting stresses lies in design accommodation to not only favor an equalization of casting section but also to favor a cooling rate to suit the heat gradients in the mold. Failing such accommodation, it is necessary to bare the bore, D, Fig. 26, and the faces of the boss immediately after solidification so that it may cool and contract slightly in advance of the arms and rim. The safe order of contraction would be: boss, arms and lastly the rim.

A split boss, as indicated at S, Fig. 26, may be introduced as a mechanical convenience, but it is very welcome to the foundryman since it relieves him of the contraction hazards referred to.

## Conclusion

It may be deduced from the information submitted that in gray iron castings of contrasting sections, or of varied cooling gradients, the thicker members of the castings referred to contract by a greater percentage than the thinner members of it. If, however, the same contrasting sections are cast as simple uniform and separate items, contraction is in line with standard allowance—that a thin section of gray iron contracts more than a thicker section for the same analysis, because of the more rapid freezing of the lighter section and the effect on the grain size and graphite formation.

In the case of a casting carrying thick and thin sections, the thin, or comparatively light sections, are subjected to an extensional stress during cooling of the metal, created by the resistance of the thicker sections which are not yet contracting, indeed, they may be actually expanding. The frictional resistance of the mold and cores, and the expansion of the cores on being heated by the molten metal, will also tend to subject the cooling metal to extensional stress. A study of the very clear arrest and expansion periods noted on the cooling curves, especially those of the heavy and large boring bar, indicates that a thick section of a casting may be undergoing an expansion at a time when a thinner section has passed its expansion phase and is in a state of contracting. Under these conditions the thin sections will suffer extensional stress.

Conversely, the thicker sections will be subjected to a compressional stress by the effort of the earlier cooled members to contract. Finally, the heavy sections, on cooling, must take up a shorter length by bending certain parts of the casting, or by fracture in the weakened or most highly stressed sections.

The introduction of air cooling to accelerate the cooling rate of the zones of the casting with a high heat content is of obvious benefit in reducing stresses and the camber allowances. This innovation is extensible to a variety of castings which present unusual and contrasting heat gradients.

While it is difficult to provide universal camber standards, the camber graph illustrated in Fig. 24, with the accompanying qualifying notes, provide a basis for the solution of many camber problems.

It is noted that while steel may have a greater contraction than cast iron its greater tensile strength and plasticity will result in reducing the hazards of destructive fractures in the type of castings referred to in the foregoing notes.

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#### DISCUSSION

*Chairman:* A. E. SCHUH, U. S. Pipe & Foundry Corp., Burlington, N. J.

*Co-Chairman:* F. G. SEFING, International Nickel Co., New York.

*CO-CHAIRMAN SEFING:* When one reads a paper like this regarding the stresses that can occur in castings, I wonder how we can ever get a casting true to pattern or one that is not cracked. This study by Mr. Longden is a very useful one because we all have problems of getting castings that are not distorted and that have not torn themselves apart.

*J. T. MACKENZIE:*<sup>1</sup> What is meant by "plated surface"?

*MR. LONGDEN:* The term refers to the working or sliding surface of the casting. It usually carries a heavier section than the vertical walls of the casting, therefore, I refer to the plated area as that surface.

*CO-CHAIRMAN SEFING:* I should like to ask the author whether he thinks that the camber chart, Fig. 24, could be used for a lathe casting, say, 20 ft long, weighing 10 tons. Could you, from the design of the casting, predetermine the camber merely from this chart?

<sup>1</sup> American Cast Iron Pipe Co., Birmingham, Ala.

*MR. LONGDEN:* Yes, it can be used for a very wide range and weights of lathe beds and of other castings.

*CO-CHAIRMAN SEFING:* That does not mean, however, that you are better off with regard to reducing residual stresses.

*MR. LONGDEN:* No, it is a matter of counteracting poor design.

*CO-CHAIRMAN SEFING:* This is, therefore, another means for the foundrymen to get around bad designs that constantly confront them.

*MR. LONGDEN:* Yes.

*CO-CHAIRMAN SEFING:* Would you say that increasing the pouring temperatures could be used as a means to decrease the stresses in a casting or, to follow your paper, that the camber factors could be reduced by increasing the pouring temperature?

*MR. LONGDEN:* Yes, definitely so. In large castings, where the rate of cooling is comparatively slow the pouring temperature of the metal is not so influential as with a thin casting, you will realize, but it has certainly a very great effect on the amount of distortion. Castings poured at a very low temperature will bend more than those poured at high temperatures because of the accommodating time of the high temperatures on the surrounding mold and the slower rate of cooling. In thin castings, undoubtedly, you can get very much distortion from metal poured at a low temperature and also destructive fractures.

*MEMBER:* Have any measurements been taken of the changes in the actual mold dimensions starting with the original before pouring, down through the solidification stage? Presumably, on a long casting such as shown in the paper there must be quite a change in the length of the actual mold itself before contraction or solidification starts.

*MR. LONGDEN:* Solidification expansion at the point of freezing actually does press the mold face back behind its original line and for a short time the mold is actually slightly larger than before pouring. The indicator or recording rods show that the casting is larger than the original mold. That is, all we can do is to test the movement of the casting itself and that is what we have concerned ourselves with in a large casting mold. The construction of the sand surface would be a very good guide, indeed, to test the change in the size or shape of the mold.

*MEMBER:* You do not feel, then, that it would be worth while to correlate, let us say, the ram density of the mold or the rigidity of the mold with the casting section and the heating effect upon the mold to make a really accurate test.

*MR. LONGDEN:* If you refer to the thermal conductivity of the mold, which is indicated by the density of the material, that is a further study. I think that might be a further study in connection with the behavior of gases also. So far as accuracy is concerned the tests outlined in the paper have been very extended over many years. The information obtained is quite accurate in its implications. Air cooling is put forward as very helpful and a chart is presented indicating the camber requirements for an extremely large range of castings. Such a chart has never before been presented for use by the foundrymen. It is the result of much experience in making the castings referred to and from results of tests named in the paper.

*CO-CHAIRMAN SEFING:* Recently some lathe bed castings running up to 12 ft in length were flame-hardened. The flame-hardening stresses, of course, were added to the casting stresses. We found that the amount of distortion on the ways could be kept within the necessary limits by merely using mold and core sand that relieved itself a great deal more. In other words, by changing to much softer cores and being careful that every part of the casting was relieved in the sand, we minimized the distortion enough so that only a small amount of the flame-hardened surface had to be ground to get it flat. Is that a sensible approach to the problem?

*MR. LONGDEN:* I should suggest that by relieving the casting of stresses during the cooling down, it would ultimately help slightly in relieving the stresses during flame hardening although a hardened surface will inevitably take up a shorter length and in so doing bend the casting somewhat depending upon the degree or depth of hardening.

# THE TECHNOLOGY OF COPPER-LEAD ALLOYS

By

R. W. K. Honeycombe \*

IT IS PROPOSED to limit this paper to a discussion of those alloys of copper containing between 10 and 40 per cent lead, with or without small percentages of other elements such as tin, silver, nickel and phosphorus. The majority of the copper-lead alloys used in heavy-duty bearings fall into this category; however, such alloys find other applications such as slide valves, metallic packing, pump liners and seals for refrigeration compressors. As the term "leaded-bronze" should be applied strictly to those copper-lead alloys containing tin, the more general name "copper-lead" will be used in this paper.

Copper-lead alloys are reputed to be difficult to cast; however, many of the difficulties encountered would be eliminated if the facts revealed by a study of the copper-lead equilibrium diagram were clearly appreciated.

## Binary Alloys of Copper and Lead

The copper-lead equilibrium diagram (Fig. 1) shows that copper and lead are mutually insoluble in the solid state, and that the solubility of copper in molten lead is not appreciable until a temperature of 900 C is reached. Alloys containing less than 36 per cent lead can become homogeneous liquids by heating above the line CD into Area 1. However, alloys containing more than 36 per cent lead lie within a "miscibility gap," that is, an area in which an emulsion of molten copper and lead is formed. This emulsion contains two liquid phases—copper-rich and lead-rich.

When a copper-lead alloy is slowly cooled from the completely molten state, the two metals will separate, forming a coarse mixture of the pure metals. Such an "alloy" would be of little use. However, when alloys containing less than 36 per cent lead are rapidly cooled from Area I, globules of lead are trapped in the branches of the copper dendrites and at the grain boundaries of the copper-rich phase, thus producing a fine distribution of lead in copper. It is clear that the faster the rate of cooling, the finer will be the resultant structure, and the greater will be the improvement in mechanical properties.

Figure 2 (A and B) shows the structure in a 75 per cent copper, 25 per cent lead alloy after air cooling and after water quenching. The water-quenched alloy has a much finer lead distribution than the air-cooled

alloy. The addition of 2 per cent tin alters the nature of the lead distribution but has no marked effect on the fineness of the structure of the alloy (Fig. 2 C and D).

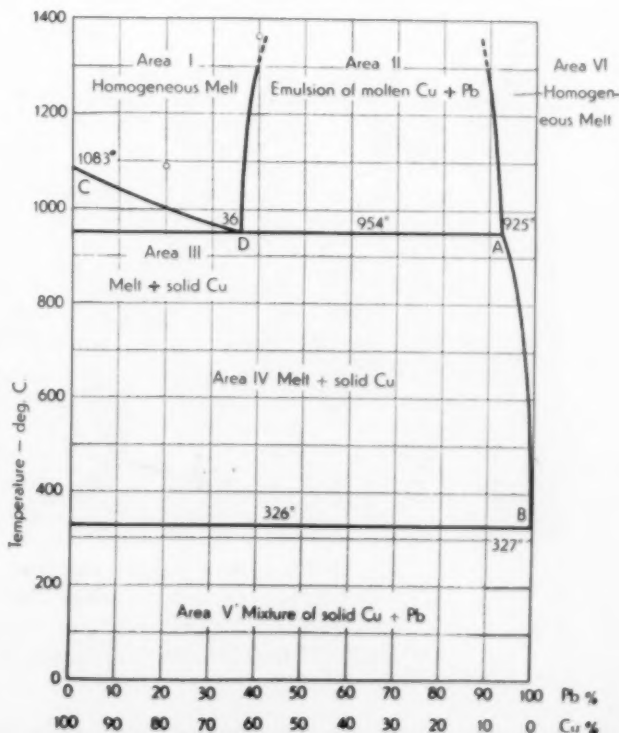
Alloys lying within the miscibility gap are still more difficult to cast satisfactorily, as the liquid alloy is not homogeneous but consists of an emulsion of molten lead in molten copper. The distribution of the lead can be improved by vigorous stirring, also by raising the temperature; however, the miscibility gap greatly increases the likelihood of lead segregation.

*A. Effect on Equilibrium Diagram*—Small percentages of many elements have a marked effect on the extent of the miscibility gap.<sup>1</sup> These elements fall into two classes:

1. Those which cause the miscibility gap to contract, e.g., silver, nickel and sulphur.
2. Those which cause the miscibility gap to widen, e.g., antimony, iron, tin and zinc.

Relative effects of the various elements can be seen in Fig. 3. Thus metals such as antimony and tin, by

Fig. 1—The copper-lead equilibrium diagram (Hansen).



\* Melbourne, Australia. Official exchange paper of the Institute of Australian Foundrymen presented by Wm. Ball, Jr., R. Lavin & Sons, Inc., Chicago, May 4, 1948 at the 52nd Annual Meeting, A.F.S., Philadelphia.



widening the miscibility gap, will increase the difficulty of dispersing a high percentage of lead in the copper-rich phase, whereas metals such as nickel and silver will facilitate the dispersion of lead by narrowing the miscibility gap.

*B. Effect on Microstructure*—The foregoing two classes of elements also have dissimilar effects on the microstructures of copper-lead alloys. The metallographic method used by the author and details of the metallography of copper-lead alloys have been described elsewhere.<sup>3</sup>

Both nickel and silver, when added to a copper-lead alloy even in small amounts (1 per cent), accentuate the semicontinuous network of the lead surrounding the copper dendrites. This type of structure is shown in Fig. 4 (*A* and *B*) which are sections of a copper-lead bearing, the first being parallel to the direction of growth of the copper dendrites on solidification, whereas the latter is at right angles.

Figure 4 thus emphasizes a fact often overlooked; namely, that the structure as seen in a polished speci-

men varies with the direction of the section as a result of the preferred direction in which solidification occurs. In the microexamination of bearings it is particularly important to remember this fact.

Addition of tin to copper-lead alloys results in a breaking up of the semicontinuous lead network to form a more or less globular distribution of the lead. As the tin content of the alloys is raised, the lead becomes increasingly globular in habit. When the tin content exceeds 8 per cent, the familiar alpha-delta constituent of copper-tin alloys appears and causes an increase in hardness. In general, the addition of tin to copper-lead alloys eliminates stringers of lead segregates which often occur when the lead is present as a network outlining the copper dendrites.

*C. Effect on Dispersion of Lead*—The effect of many metals on the segregation of lead in copper-lead alloys has been extensively studied. A summary of these investigations has been made by Bassett.<sup>2</sup> These results can be more or less systematized if it is remembered that the effect of an element on the miscibility

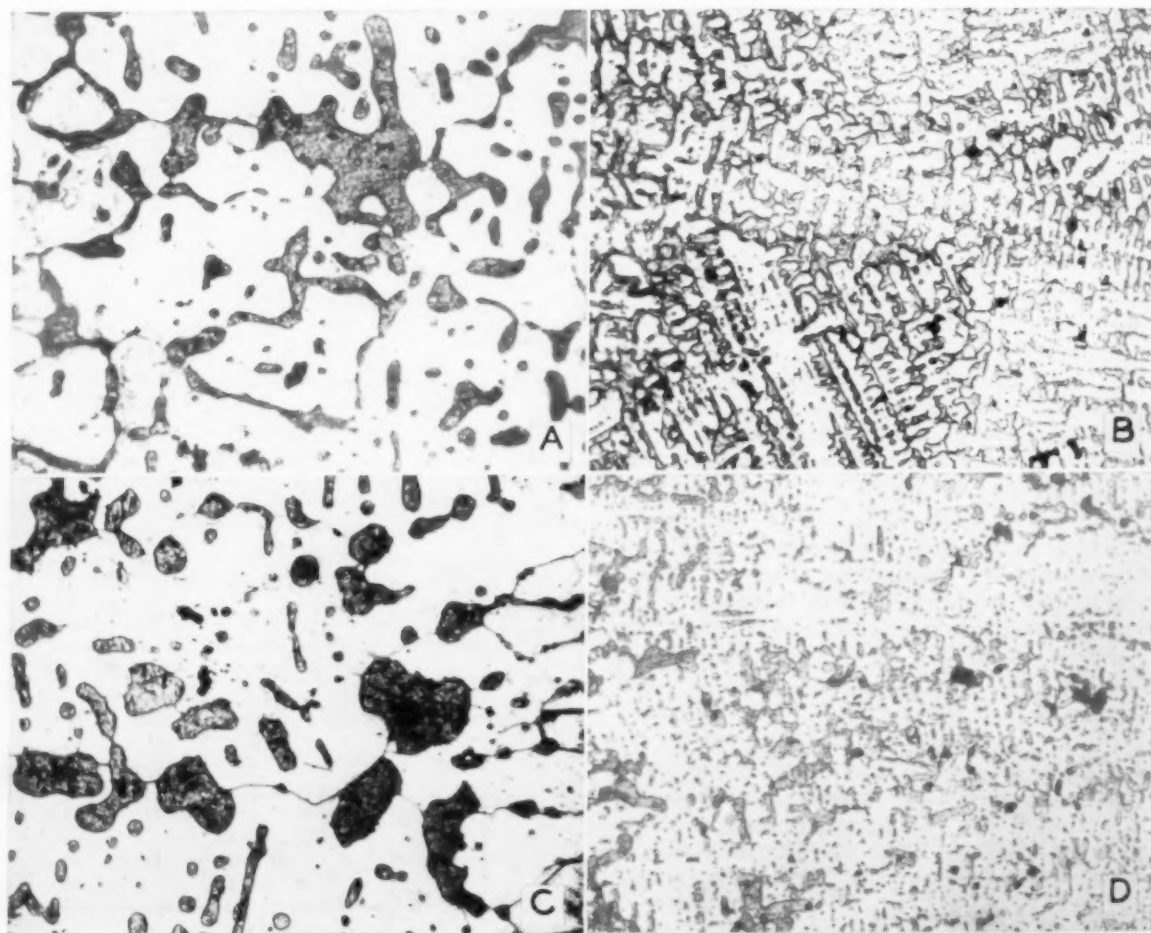


Fig. 2—Photomicrographs showing structures; *A*—copper-lead alloy (25 per cent lead) air cooled; *B*—copper-lead alloy (25 per cent lead) water quenched; *C*—copper-lead-tin alloy (25 per cent lead, 2 per cent tin) air cooled; *D*—copper-lead-tin alloy (25 per cent lead, 2 per cent tin) water quenched.  $\times 150$ .

gap will determine to a large degree its influence on the solution of lead in copper. Tin is the most familiar example of these elements.

Although tin has a beneficial effect on the microstructure of the alloys, it limits the percentage of lead which can be taken into solution at high temperatures. Tin is typical of the metals which widen the miscibility gap. The higher the percentage of tin, the more limited will be the region over which miscibility is possible at high temperatures.

#### Tin Content—Casting Temperature

Some experiments were carried out<sup>4</sup> to determine the effect of casting temperature and tin content on the distribution of lead in a copper-lead alloy (25 per cent lead). Charges containing 0, 1, 4, 8 and 12 per cent tin, the remainder being 25 per cent lead plus copper, were put into small steel cylinders (6x2 in.) and placed in an electric furnace at constant temperature for 20 min. They were then withdrawn carefully without agitation and water quenched. The cylinders were sectioned and one half was polished in each case to determine the extent of solution of the lead. The results are summarized in Table 1.

Even at a temperature of 1300 C only partial solution of lead occurred when the tin content exceeded one per cent. In general, the higher the tin content the more difficult it was to bring the lead into solution. It should be emphasized that the conditions of the experiment compared unfavorably with practice as no attempt was made to stir or disturb the molten alloy in any way. However, similar results were obtained from chemical analyses of castings. Table 2 shows both the actual composition of the copper-rich portions of a series of castings and the nominal composition of the alloy charge. The divergence of the lead percentages at high tin contents is quite evident.

The following general conclusions can be drawn:

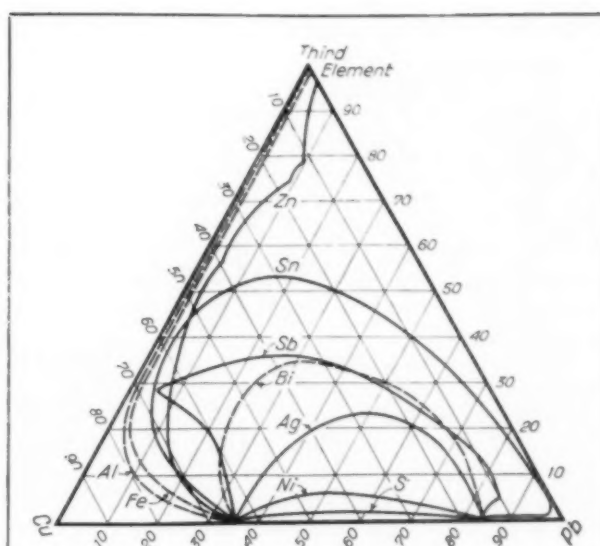


Fig. 3—Isothermal sections of ternary system of copper, lead, and a third element showing effect of various elements on the miscibility gap (Glaus).

1. As the tin content of an alloy is increased, the temperature *prior to pouring* (not necessarily the pouring temperature) should also be increased.
2. Alloys containing high percentages of tin cannot be expected to retain a high percentage of lead as a *fine distribution* at room temperature.
3. Accurate temperature control is essential.
4. Agitation of the melt is desirable.

On the other hand, elements which tend to close the miscibility gap will promote the solution of lead in copper at high temperatures. High-lead alloys containing up to 30 per cent silver can be readily cast.

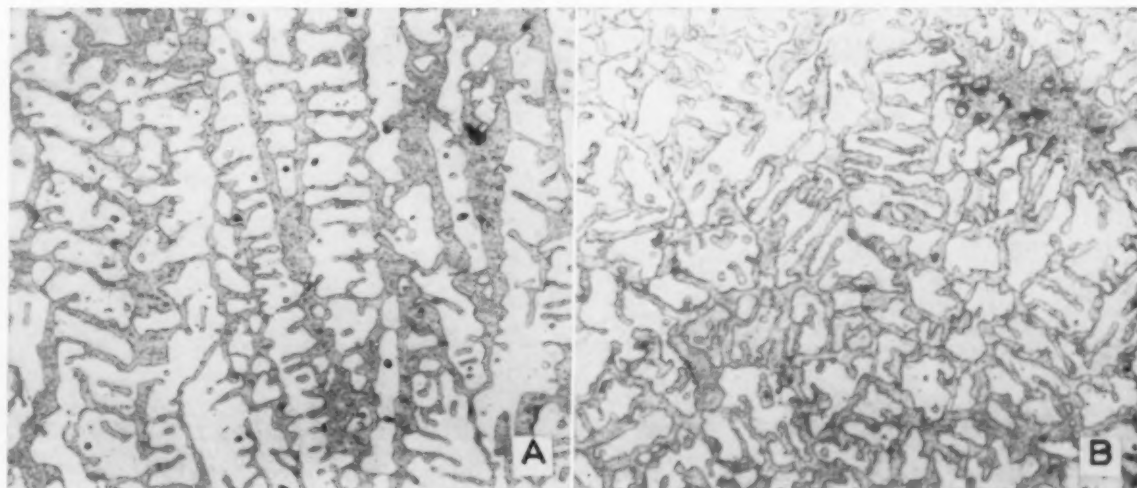


Fig. 4—Microstructures of an alloy containing approximately 25 per cent lead and 1 per cent silver: A—section parallel to direction of growth of copper dendrites; B—section at right angles to direction of growth of copper dendrites. X 250.

With such alloys up to 50 per cent lead can be uniformly dispersed. Thus, if a very high lead content is required, a metal which narrows the miscibility gap, e.g., nickel or silver, should be added.

However, this means of producing a fine distribution of lead in copper has some disadvantages. First, the fine semicontinuous lead network produced has an adverse effect on the physical properties of the alloy, and second, microsegregation readily occurs. Lead tends to segregate between copper dendrites in the form of long stringers parallel to the direction of solidification.

It is thus obvious that there are two causes of lead segregation in copper-lead alloys:

1. "Macrosegregation" due to lead which is not dissolved in the molten copper at high temperature.
2. "Microsegregation" due to the coarseness of the lead precipitated from the copper on cooling. This may occur as isolated globules or as dendritic stringers.

It might be more accurate to define the segregation

described in (1) as primary segregation of lead, because it occurs as a result of the lead not going completely into solution. Segregation of the second type could then be described as secondary segregation, i.e., segregation of lead precipitated from solution in copper on cooling.

It is thus not strictly accurate to say that an alloying element reduces segregation, for in reducing the amount of microsegregation it may quite well increase the amount of macrosegregation. Furthermore, it can only be said that an alloying element tends to reduce macrosegregation as variations in casting technique and cooling rate after casting may completely eliminate the defect.

*D. Use of Deoxidants*—It is general practice to use small percentages of certain elements such as phosphorus as deoxidants. Actually, addition of as little as 0.05 per cent phosphorus is ample to efficiently deoxidize copper lead alloys, although much higher percentages of phosphorus often are recommended on the

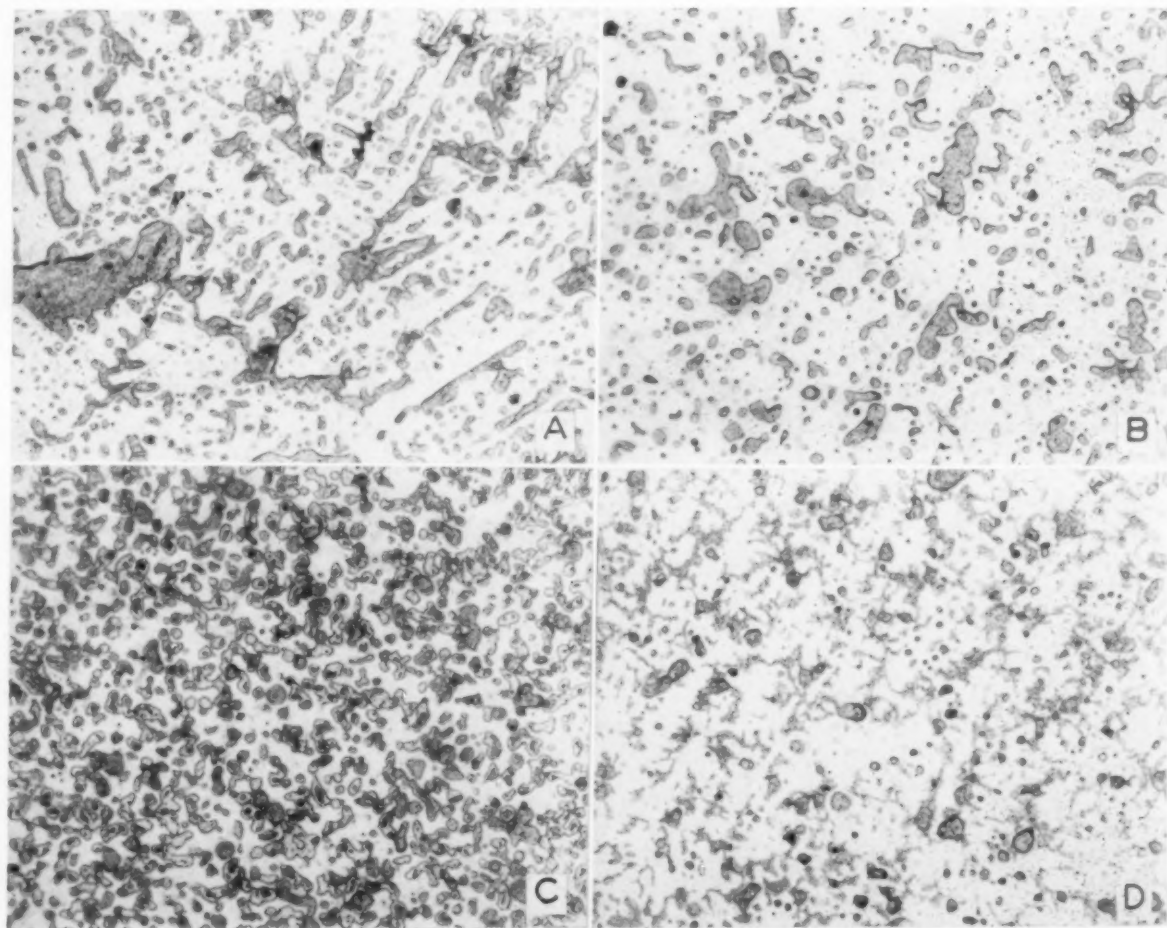


Fig. 5—Microstructures of copper-lead alloys containing increasing percentages of tin. A—0.8 per cent tin. The inter-dendritic lead network is partly broken up. B—4.0 per cent tin. Lead is present as discrete globules. C—9.0 per cent tin. Lead occurs as small globules. The copper-rich solution is cored and a small amount of the alpha-delta eutectoid is present. D—15 per cent tin. There is only a small percentage of lead present as small globules. Alpha-delta eutectoid has greatly increased.  $\times 250$ .



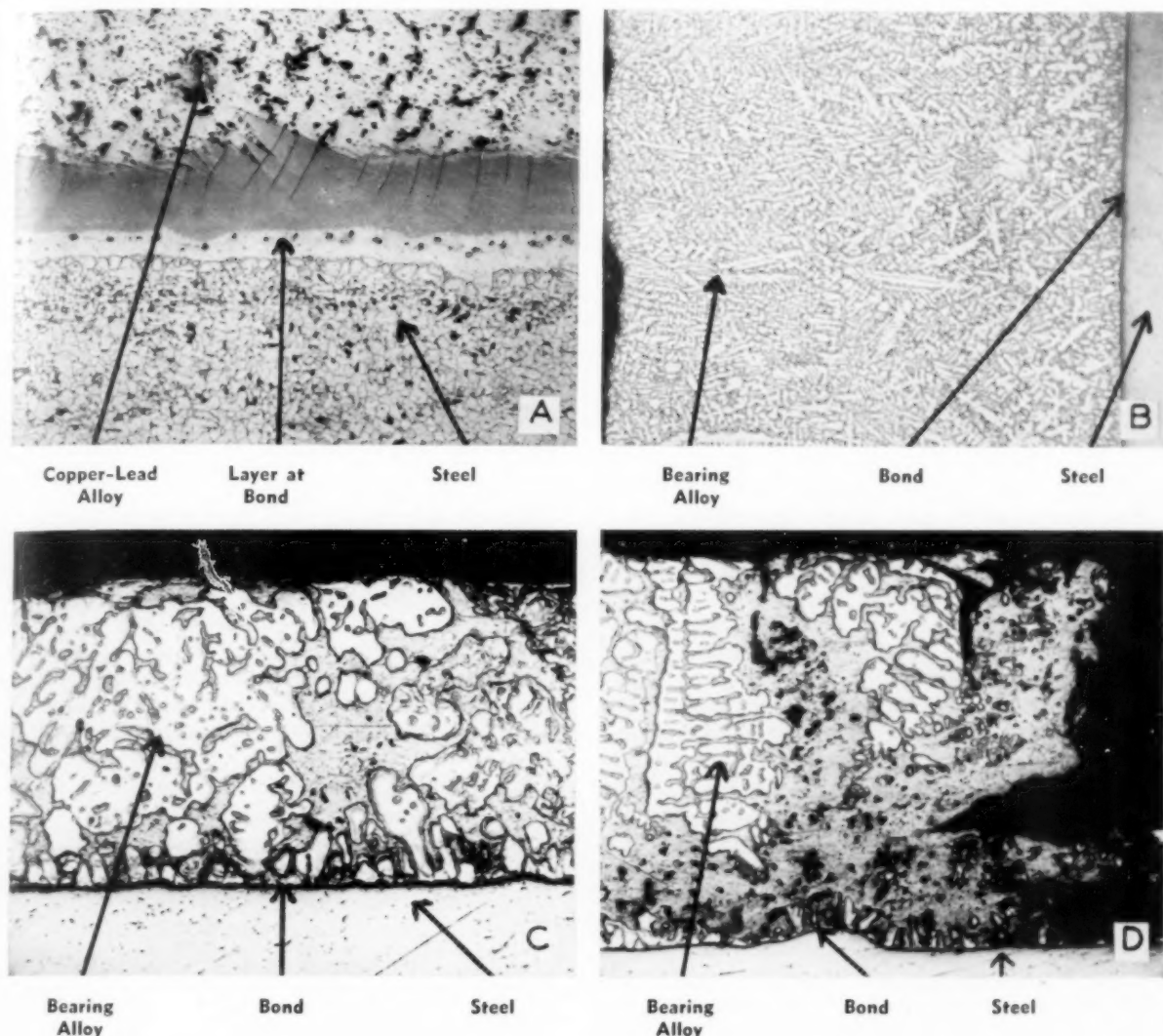


Fig. 6—A—Bond between steel and a copper-lead alloy containing 0.55 per cent phosphorus showing the formation of a thick brittle layer (tapered section).  $\times 50$ . B—Cross section of a typical copper-lead aircraft engine bearing. The lead is present as a fine semicontinuous network.  $\times 75$ . C—Cross section of a copper-lead bearing showing marked lead segregation.  $\times 75$ . D—Failure of a copper-lead bearing showed marked lead segregation.  $\times 75$ .

grounds that lead segregation is minimized and a finer structure obtained. It is the firm opinion of the author that high percentages (i.e., 0.1 to 1 per cent) of phosphorus should be avoided. Lead segregation can be eliminated and a suitable structure obtained without the need to resort to large additions of phosphorus.

Even small quantities have a very detrimental effect on the thermal and electrical conductivities of copper-lead alloys, a fact which should severely limit the use of phosphorus in alloys to be used for bearings where high thermal conductivity is a desirable property. The presence of more than 0.2 per cent of phosphorus is particularly undesirable if the alloy is to be bonded to a steel backing, as in car and airplane main bearings.

It has been shown<sup>5</sup> that the bonding of the copper-lead alloy may be unsatisfactory if the phosphorus content of the alloy approaches 0.5 per cent. The bond

is rendered weak and brittle because of the formation of a phosphorus-rich layer on the surface of the steel. This layer can be clearly seen under the microscope (Fig. 6 A) and adversely affects bonding test results.

In the light of the foregoing it is possible to formulate some general rules to apply to the casting of copper-lead alloys. During melting it is essential to ensure the complete solution of the lead in the copper. To accomplish complete solution the alloy must be heated to a high temperature, 1200-1300 C, although not necessarily poured from this temperature range. The maximum temperature (to which the alloy is raised) during the melting should depend on: (a) the alloying elements present; and (b) the lead content.

In melting alloys containing high percentages of tin and lead, it is advisable to heat to 1250-1300 C prior to pouring, whereas in alloys containing silver, it need

TABLE 1—EFFECT OF TEMPERATURE ON SOLUTION OF LEAD IN COPPER-TIN ALLOYS

Alloy Composition, per cent Tin	Temperature of Alloy		
	1100 C	1200 C	1300 C
0	Lead partially in solution	Practically all lead in solution	Complete solution
1	Lead partially in solution	Partial solution	Practically complete solution
4	Much lead undissolved	Much lead undissolved	Partial solution of lead
8	Most lead undissolved	Much lead undissolved	Partial solution of lead
12	Most lead undissolved	Most lead undissolved	Partial solution of lead

not be necessary to heat above 1150 C. If the lead content of the alloy exceeds 30 per cent, then special precautions should be taken to ensure maximum dispersion of the lead, e.g., the temperature of the alloy should be raised and the melt stirred vigorously.

Temperature of pouring is not necessarily the maximum temperature reached; in fact, it is usual to allow the alloy to cool somewhat after complete solution of the lead has been achieved. The pouring temperature chosen should be above 1100 C, preferably in the range 1150-1250 C, for alloys with a high lead content. Unless the alloy is to be water quenched, as in the case of some bearing castings, the lowest pouring temperature compatible with homogeneity of the alloy should be chosen as it is very desirable that the alloy solidify quickly. The slower the rate of solidification, the coarser will be the lead distribution.

The molten alloy should be oxidized vigorously by removing the crucible lid and allowing a blast of air to pass over the molten metal for a minute or so. Finally, just prior to pouring, deoxidation can be effected with a small amount of phosphor-copper (0.05 per cent phosphorus in final alloy). This deoxidation with phosphor is not absolutely necessary, but it usually is safer to use it.

#### Alloying Elements—Effect on Physical Properties

Microstructures of copper-lead alloys have an important bearing on their physical properties. Those alloys in which the lead is present as a semicontinuous network will obviously have much poorer mechanical properties than the alloys in which the lead is present as discrete globules. Thus a copper-lead alloy containing 25 per cent lead and one per cent silver has much lower tensile strength and hardness than an alloy con-

taining 25 per cent lead and 2 per cent tin. Table 3 shows the hardness and tensile strengths of three different copper-lead alloys which were cast and water quenched in the same way as are many bearing castings, thus producing a fine structure.

The effect of increasing tin content on a series of copper-lead alloys containing nominally 25 per cent lead is to steadily increase the hardness. On the other hand, the addition of small percentages (2-3 per cent) of silver to a copper-lead alloy results in little change in hardness.

Thermal properties of the alloys are important in bearing applications where the rapid conduction of heat from the bearing surface is important. In copper-lead alloys the copper-rich phase is a good conductor; however, its thermal conductivity is markedly reduced by small quantities of certain elements,<sup>1</sup> of which phosphorus has the most pronounced effect. Other elements which markedly change the thermal conductivity are iron, nickel and tin. On the other hand, silver has a much smaller influence on the thermal properties.

#### Copper-Lead Alloys—Friction and Wear

Frictional and wear properties of copper-lead alloys are very important as the majority of these alloys are used in applications where it is essential that these properties be of a high order, e.g., bearings, slides, liners, etc. The presence of lead in the elemental state is all-important in these applications. Bowden and

TABLE 3—TYPICAL MECHANICAL PROPERTIES OF SOME COPPER-LEAD ALLOYS

Alloy	Lead Distribution	Hardness, BHN	Ultimate Tensile Strength, ton/sq. in.
25 per cent lead, balance copper	Semi-continuous network	31	4.2
25 per cent lead, 1 per cent silver, balance copper	Semi-continuous network	31	4.3
25 per cent lead, 2 per cent tin, balance copper	Globules	46	7.1

Tabor<sup>6</sup> have suggested that when copper-lead alloys form rubbing surfaces, in the absence of a lubricant, the lead tends to smear over the surface and form a thin film. This thin film of lead acts as an extreme pressure lubricant and lowers the friction between the contacting surfaces.

It has also been suggested that the good machining characteristics of leaded brasses and bronzes are due to the formation of this lead film on the surface during cutting. Observations of the machining of copper-lead alloys of high lead content indicate that this is indeed the case. Frequently, the chips produced are leaden gray in appearance, whereas the machined surface possesses the characteristic reddish appearance of copper.

That the gray color of the chips is due to a thin lead film formed as a result of high local temperatures can be shown by testing with a lead etchant, e.g., hydrogen peroxide-acetic acid mixture. The thin lead film is immediately dissolved and the chips assume a reddish appearance. The advantage of such a lead film in bearings is that it tends to minimize bearing seizures.

TABLE 2—ANALYSIS OF A TYPICAL SERIES OF COPPER-LEAD-TIN ALLOYS

	Alloy No.						
	1	2	3	4	5	6	7
	Actual Composition, per cent						
Copper.....	74.7	74.1	74.2	74.1	71.6	74.3	73.2
Lead.....	24.6	25.0	24.6	23.4	24.1	16.4	11.5
Tin.....	0.31	0.48	0.83	2.1	4.1	9.3	15.0
	Nominal Composition, per cent						
Copper.....	74.75	74.5	74.0	73.0	71.0	67.0	63.0
Lead.....	25.0	25.0	25.0	25.0	25.0	25.0	25.0
Tin.....	0.25	0.50	1.0	2.0	4.0	8.0	12.0

TABLE 4—TYPICAL SPECIFICATIONS FOR LEAD BRONZES

Specification	Composition, per cent					Applications
	Copper	Tin	Lead	Zinc	Others	
SAE 900	81-85	6.5-7.5	6-8	2-4	—	Bronze bearing castings
ASTM B144 (Alloy 3C)	83-86	4.5-6	8-10	0.5-2	Nickel 0.50	—
ASTM B144	77-81	9-11	8-11	0.75 max.	Nickel 0.50	Sand castings
ASTM B30 (Alloy 3D)	76-79	6.5-7.5	14-16	0.75 max.	Antimony 0.75	Sand castings
ASTM B66 (Med. Bronze)	69-79	6-8	14-22	1.25 max.	0.75 max.	Locomotive wearing parts (sand or metal molds)
ASTM B66 (Soft Bronze)	64-70	6-8	23-27	0.75 max.	0.75 max.	Locomotive wearing parts (usually metal molds)
ASTM B144 (Alloy 3E)	67-71	4.25-5.5	23-27	0.50 max.	Nickel 1.0	Sand castings

Bowden and Tabor<sup>6</sup> have shown that the distribution of the lead plays an important role in the frictional behavior of copper-lead alloys. Those alloys possessing a more or less continuous lead network possess better frictional properties than the alloys in which the lead is globular in habit. In the former type, the channels of lead provide a continuous supply of lead to the rubbing surface, particularly if the bearing becomes very hot, whereas in the latter case only a limited supply of lead is available at the surface.

In silver-lead aircraft engine bearings a thin film (0.001 in.) of lead is electroplated on a silver backing which is very resistant to fatigue and possesses good thermal conductivity but lacks good frictional properties. The frictional properties of the surface are considerably enhanced by the presence of the thin, soft lead film.

#### Copper-Lead Alloy Applications

Copper-lead alloys of high lead content are used principally in applications where good anti-frictional properties are required, e.g., in bushings and bearings. The alloys are sand cast or chill cast, depending on the composition of the alloy and the actual application. For example, the alloys containing tin (leaded bronzes) are frequently sand cast. The globular habit of the lead in this type of alloy minimizes segregation, which occurs more readily in the alloys in which the lead is present as a semicontinuous network around the copper crystals. Typical compositions used for bearings and bushings are shown in Table 4.

The alloys used for heavy-duty bearings in car, diesel and airplane engines usually possess a high lead content (20-30 per cent) and only small quantities of other alloying elements. Typical compositions are shown in Table 5.

These alloys have excellent bearing properties, but as their mechanical properties are inadequate they usually are bonded to a steel shell. Some overseas methods for producing these bearings have been described.<sup>7</sup> During the war, methods were developed locally for the use of manufacturers.<sup>8</sup> The problem was to bond firmly to steel backing a thin layer of copper-lead bearing alloy with a fine structure completely free from defects such as lead segregation and shrinkage porosity.

Figure 6 B shows a typical cross section of an aircraft engine main bearing. The alloy is similar to No. 3 referred to in Table 5; the lead is visible as a gray network. Somewhat coarser structures are quite satisfactory for automobile and diesel engine bearings. Figure 6 C shows an example of bearing lead segrega-

TABLE 5—TYPICAL COPPER-LEAD BEARING ALLOYS USED IN AIRCRAFT AND DIESEL ENGINES

Alloy No.	Composition, per cent				
	Lead	Tin	Silver	Nickel	Copper
1	25-30	1	—	—	balance
2	25-30	—	0.75	—	balance
3	25-30	—	—	1	balance
4	15-20	8	—	—	balance

tion which should be avoided in these bearings, as it frequently leads to a failure such as is shown in Fig 6D.

Overseas, the technique of handling such copper-lead alloys has reached a stage of refinement such that the alloy can be bonded to a continuous steel strip, either by casting or by sintering. The resultant structure is quite uniform and the bearings have been extensively used in cars and trucks throughout the war.

In conclusion, it must be emphasized that many of the difficulties encountered in the treatment of copper-lead alloys arise from a neglect of the principles outlined in the foregoing. If these are appreciated, and if such important matters as temperature control and deoxidation are not overlooked, most copper-lead alloys should prove quite amenable to treatment.

#### Acknowledgments

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## DISCUSSION

Chairman: R. M. BRICK, University of Pennsylvania, Philadelphia

Co-Chairman: C. A. ROBECK, Gibson & Kirk Co., Baltimore.

JOHN HAGENAUER:<sup>1</sup> Mention was made that the lead content of the alloy was as high as 40 per cent but no mention was made of the nickel content of the alloy.

WM. BALL, JR.:<sup>2</sup> The nickel content is 1 per cent.

MR. HAGENAUER: What melting controls does the author use?

MR. BALL: It is important to minimize hydrogen absorption by the melt. Hydrogen absorption by the melt will tend to bring about oxidation of the melt. When that happens phosphorus must be added for deoxidation but the amount added must be carefully watched, because it will affect the thermal characteristics of the alloy. This was important in airplane bearings which were used to a considerable extent during the last war. The alloy was used in aircraft and in diesel engine bearings where thermal conduction of the heat is important.

The author says that phosphorus will tend to reduce that characteristic. Whenever you get up around 2380 F or 1300 C that is near boiling for this type alloy with 40 per cent lead. Hence it is important that the alloy not become burnt.

E. M. SMITH:<sup>3</sup> It would be well to compare the practices as described by the author and the American practices, as to their similarities or their differences.

MR. BALL: The practices as set forth in the paper by the author are similar to the practices in our own country.

H. J. ROAST:<sup>4</sup> The practice described in the paper follows the Canadian practice very closely. I have made several million pounds of this copper-lead alloy without tin. The author mentions use of not only tin but sulphur as closing the gap between solution of the lead and the copper at the high temperature. His point is that the closing of that gap is a desirable thing from one point of view. He also mentions that tin tends to cause the lead to be globular or spheroidal in form, whereas the use of the items closing the gap has a tendency to distribute it around the different areas of the copper.

The temperatures the author uses are similar to the temperatures used in Canada. The secret is, in my opinion, as the author says, in using a high melting temperature. The pouring temperature is something entirely different. The author should have elaborated on the use of sulphur. The reason for using sulphur was never clear to me, although that is the only addition we use to get a thoroughly good mixture or emulsion of the lead and the copper.

As to the question of tensile strength, it is of little moment except in regard to the deformation. I prefer the deformation limit. Then it becomes a question of designing your bearing so that your deformation limit is not reached. Undoubtedly as they were successful, they took care to have sufficient area of bearing to take care of the pressure per square inch. Some engineers like to have a small bearing. Many of them seem to always want too small a bearing surface.

I would say that the practices set forth in this paper compare with the American and with our Canadian practice.

MR. BALL: Mr. Roast told me that when he makes this alloy he drops a little brown paper bag into the molten metal.

MR. ROAST: I put the sulphur in a brown paper bag, drop it in a red hot crucible, and pour the leaded copper on top, but I am not doing it as a secret. The brown paper bag gives you a fractional second to keep the sulphur there so that the metal goes on it. I never found it satisfactory to try to put the sulphur in in any other way than dropping it into another pot and pouring the copper and lead onto it in that manner.

R. A. COLTON:<sup>5</sup> In using the high lead-copper alloys, it has been traditional to recommend the use of nickel to prevent segregation. This suggestion is common in the technical literature and you can find it in many recommendations on analyses and handling of this type of alloy. I would like to know whether there is any basis in fact for adding nickel to a copper-lead, or a copper-lead-tin alloy to reduce segregation. When you look at the constitution diagram of the copper-nickel system and find complete miscibility across the diagram I wonder whether nickel

has any effect at all. It may, but I believe that 1 per cent nickel in the alloy would more than likely be dissolved in the copper. A complex compound could form, of course, and influence rates of nucleation. I wonder whether there is any basis for believing that nickel does prevent segregation or inhibit it.

MR. BALL: I agree that nickel goes into solution with the copper. We have never relied on nickel to give us lead distribution in this alloy.

G. P. HALLIWELL:<sup>6</sup> Our experience with these alloys has been similar to that of the authors. Best results were obtained by melting in an oxidizing atmosphere at a high temperature to effect a liquid solution of the copper and lead, and then pouring at a low temperature to prevent lead segregation and retaining it in a finely divided state in the solid alloy.

As far as nickel is concerned, I have failed to obtain any noticeable benefit from it. The results have been too uncertain. The action of any reducing gas being evolved during solidification might conceivably modify the effect of nickel on the lead particle size, shape or distribution.

MR. BALL: Do you believe that sulphur when added to this molten alloy goes into solution with the lead or is it a washing action, mechanical action, or a chemical action?

MR. HALLIWELL: I believe it forms a sulphide.

MR. BALL: I was wondering whether we would get a sulphide.

MR. HALLIWELL: I have no definite information on it, but knowing the chemical characteristics of the two elements, I would expect it to form a copper sulphide.

MR. ROAST: Here is a point. When you dissolve the alloy in nitric acid, the sulphur comes up as a complete yellow globular pure sulphur whatever that indicates.

L. M. LONG:<sup>7</sup> I believe the sulphur remains as elemental sulphur. There is no reaction to convert it to a sulphide.

It is recognized in bearing applications that addition of nickel to this alloy increases wear resistance. It does refine the grain and therefore gives a finer lead distribution. Nickel does not help make the lead more soluble. We have always accepted nickel as being helpful for lead distribution, but it is not, except the fact that it refines the grain, and it is detrimental to wear.

CLAIR UPTEGROVE:<sup>8</sup> Does not the factor of adsorption of the added element enter into this problem? We are adding various materials to the melt. Anything that can interfere with the growth of the lead particle, once it forms, will help us in solving this problem of segregation. I think that is why many of these things which are added to the melts do work out.

CHAIRMAN BRICK: There are perhaps some other things beyond adsorption that could be mentioned. The author and the paper quite properly distinguish between macro and micro segregation. Macro segregation would be the separation in a gross sense; one part of the casting will have 30 per cent lead and elsewhere will have only 20 per cent in a really high lead alloy bearing.

Micro segregation would refer particularly to the degree to which the lead is separated in the microstructure. The appearance of the author's micrographs is markedly affected by whether the specimen was polished parallel to a dendritic axis or perpendicular. If you polish perpendicular or take a photograph at a place where you have a perpendicular view of the lead dendrites or of the interdendritic lead filling, they will have one appearance. If you examine it parallel to the dendritic growth, it will have a different appearance.

Presumably the author, as any microscopist, is completely honest and selected just the proper view, which really represented the entire casting. Apparently the judgment of the discussors bears out the author on the effect of tin in altering the distribution of lead. There may be some other explanation than the effect of tin on the miscibility gap or the monotectic dome in the phase diagram of copper-lead system. Many of you were at the Institute of Metals Lecture in February in New York and heard Dr. Cyril S. Smith of the Institute of Metals of the University of Chicago lecture on the effect of surface tension. I am sure the concepts he invoked there are very important in this question of the form of the distribution of lead, and the effect of nickel or tin or sulphur in relation to the distribution of the lead. When tin is added to a ternary alloy, there is the effect of tin on the miscibility dome. However

<sup>1</sup> Bridesburg Foundry Co., Fullerton, Pa.

<sup>2</sup> R. Lavin & Sons, Inc., Chicago

<sup>3</sup> Baicelle Memorial Institute, Columbus, Ohio

<sup>4</sup> Canadian Bronze Co., Ltd., Montreal, Canada

<sup>5</sup> American Smelting & Refining Co., Perth Amboy, N. J.

<sup>6</sup> H. Kramer & Co., Chicago

<sup>7</sup> L. M. Long & Associates, Toledo, Ohio

<sup>8</sup> University of Michigan, Ann Arbor, Mich.

some of the tin must be dissolved in the lead globules. The effect of tin on the surface tension of liquid lead or the interface between the lead and copper dendrites surely must be very significant in determining the actual micrographic appearance of the lead globules and therefore the strength of the bearing and its bearing qualities.

Mr. COLTON: We have had some experience with the effect of cooling rate on the lead distribution and the lead particle size in the conventional bearing alloys. I do not believe that the 5 per cent of metal to freeze could have too great an effect because it is the copper matrix which has already frozen that would affect lead distribution. There will be liquid lead present well after the copper has solidified, so that the lead is interdendritic. The lead then has no choice but to freeze where it is.

We have made some experiments which may be of interest. The American Smelting & Refining Co. has a successful process for continuously casting bronzes including the high-lead bronzes. We have run some experiments comparing sand castings, chill castings, and our continuously cast bushing bars, all 1-in. round bars. We tried to make them in a comparable fashion, using good practice, and then sectioning the various bars and comparing their microstructure and also comparing their respective properties. We have found that the lead distribution from the sand cast to the continuous cast (a high chill process), progresses in a manner that you might expect. With the slow cooling rate that you get in sand, and this is green sand, you find a very coarse lead distribution. In chill castings, made in cast iron molds, you get a somewhat better distribution of lead and smaller-sized lead particles. In the continuous cast bar you get an excellent distribution with fine lead particles.

The alloys we studied were the 75-5-20, one of the most difficult to cast in the foundry, 85-5-5-5, 83-7-7-3 and 80-10-10, all common bearing alloys. It looks as if the cooling rate in itself, the amount of chill in the casting, in the bushing, may have as big an influence on the size and distribution of the lead particles as any other factor.

Mr. BALL: At the beginning of cooling and the forming of the dendrites we have the lead filling in. We have had excellent experience with these alloys in high pressure castings. I would say these leaded alloys, especially the 80-10-10 and the 85-15 are excellent for pressure castings as long as you do not have heat to contend with. If the castings are to be subjected to pressure, such as steam valves on a locomotive where the pressures are up to 260-270 lb steam pressure, you then have a problem in creep. These alloys are poor in creep resistance. But for room temperatures they are fine, showing that I do not believe that the lead does fill in on these segregation points.

J. T. EASH:<sup>9</sup> During the war, we conducted an investigation of the effect of different addition elements on lead distribution in copper-base alloys in order to find a substitute for tin in the 80-10-10 bronze. The alloys were melted in an induction furnace and sand cast. In the 80-10-10 bronze, the lead was agglomerated and well distributed throughout the matrix; but, in the absence of tin, lead was distributed in an intergranular formation. The addition of nickel tended to shorten the intergranular films. Nickel in combination with antimony was particularly beneficial in agglomerating the lead and produced an alloy having good mechanical properties and low wear rate. Based upon the Amsler wear test, it was found that a bronze containing 10 per cent lead, 8 per cent nickel and 1 per cent antimony was equally as good as 80-10-10 in resisting wear. It was necessary to avoid superheating the melt to obtain the best distribution of the lead.

B. M. LORING:<sup>10</sup> If the chill is important in influencing the structure in these alloys, then the method by which the alloy is

placed in contact with the steel base must be rather important. What method was used in the present investigation in placing the copper-lead on the steel?

W. B. GEORGE:<sup>2</sup> Fusing copper-tin-lead alloys on a steel strip is a complicated process practiced by only a few plants. The process requires heating the steel strip in a bath of borax at 1950 F and then pouring the molten copper-lead-tin alloy on the steel strip surface through the liquid borax and then allowing it to solidify.

#### AUTHOR'S CLOSURE

R. W. K. HONEYCOMBE (*Author's Closure*): I would like to take this opportunity of thanking Mr. Ball for presenting my paper to the meeting and for so ably dealing with the discussion. I was very interested to read of the various American methods of handling copper-lead alloys, and to compare them with those used in my own country. It is clear that the difficulties encountered have been solved in very similar ways.

The addition of small percentages (1 per cent) of nickel to copper-lead alloys results in a microstructure in which the lead is present as a semi-continuous network, very like the structure obtained by addition of the same percentage of silver. However, the nickel appears to harden the copper matrix to a much greater extent than silver, and thus limits the use of the alloy in bearings. This hardening of the matrix no doubt accounts for the increase in wear resistance mentioned by Mr. Long, but it would also result in increased crankshaft wear. It is rather for the user to choose which part he wants to wear more! I think that nickel does tend to reduce lead segregation in a similar manner to silver.

I entirely agree with Prof. Brick that the effect of various metals on the lead distribution can be explained in terms of surface tensions as has recently been shown very beautifully by the work of Dr. C. S. Smith at Chicago. Silver and nickel encourage the formation of a continuous network of lead, because the tension between the lead-rich and copper-rich phases is much less than the tension between copper-rich grains. At the other extreme, a near globular distribution of lead is obtained when tin or antimony is added, the inference being that these elements have an opposite effect on the surface tensions of the two phases.

Mr. Colton holds the view that once the copper has solidified then the distribution of the lead is fixed. I agree, and would mention that in Australia we have employed methods for producing copper-lead aircraft engine bearings in which the alloy was poured into a red-hot (1000 C) steel shell which was then quenched not in water, but in molten aluminum or lead. This resulted in rapid solidification of the copper without detriment to the lead distribution. Once the casting had attained the temperature of the bath, it was then finally cooled in a water spray. This two-stage treatment reduced the thermal shock to the casting and minimized the formation of cracks in the copper-lead alloy.

Mr. Loring mentions the chilling produced by pouring copper-lead alloys on to steel. In our work we were mainly concerned in producing steel-backed bearings, and in the process used, the steel shell was heated in molten borax at 1000 C for a short time. The molten copper-lead alloy was then poured in and displaced the borax, thus coming in contact with a clean, hot steel surface. No chilling occurred and on quenching the casting in metal or water, an excellent bond was obtained between the copper-lead and the steel.

I agree with Mr. Roast that tensile test results are of little use when one is dealing with bearings. I would go so far as to say that most if not all mechanical test results are of little direct use in evaluating bearing alloys. Frictional and wear properties are perhaps the most useful, but these properties have no direct relation with mechanical test results. Many engineers and metallurgists would probably agree that in the case of bearing alloys the proof of the pudding is in the eating.

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<sup>10</sup> Naval Research Laboratory, Washington, D. C.

# INFLUENCE OF CHROMIUM ON GRAPHITIZATION OF WHITE CAST IRON

By

Gabriel Joly \*

PIG IRON and steel scrap that malleable foundry charge into the melting furnace often contain impurities which affect graphitization of the iron during annealing. The most frequently occurring impurity is chromium. When present in the iron even in small percentages, it precludes formation of a normal microstructure with standard annealing practice.

In the present study the author proposes to determine the annealing cycles necessary to obtain, with each chromium content accidentally occurring, an iron having ferritic-nodular graphitic structure.

## Chromium Content of Iron Studied

The base pig iron that the author used had been refined in a reverberatory furnace heated by pulverized coal. Its chemical composition was as follows: C—2.16, Si—1.50, Mn—0.43, S—0.09, P—0.13 and Cr—0.03 per cent.

This iron was remelted in a coke-fired crucible furnace. Ten consecutive heats were made, and they are divided into two groups:

1. Five heats with chromium additions, and
2. Five heats with chromium additions to which 0.002 per cent boron was added to the melt in the form of 17-18 per cent ferro-boron.

Table 1 gives the chromium content of the ten heats made. Chromium was not intentionally added to Heats 1 and 6b. Heats 6b, 7b, 8b, 9b and 10b contain boron.

\* Head of Cast Iron Metallurgy at the Foundry Industry Technical Center, Paris, France.

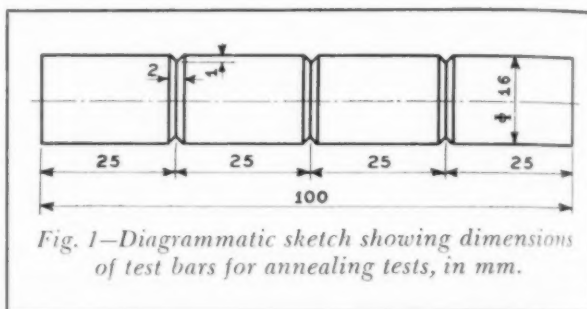
Official Exchange Paper of The French Foundry Technical Association for presentation at the 1948 A.F.A. Annual Meeting May 3 in Philadelphia, Pa. Manuscript translated from the French. Presented by C. O. Burgess, Technical Director, Gray Iron Founders' Society, Cleveland.

TABLE 1—CHROMIUM CONTENT OF TEN HEATS POURED

Heat No.	Cr Content, per cent
1	0.032
2	0.082
3	0.099
4	0.125
5	0.130
6b	0.035
7b	0.056
8b	0.086
9b	0.124
10b	0.137

The exact boron content has not been determined. It is about 0.002 per cent.

These ten heats permit not only determination of annealing cycles when the metal contains small percentages of chromium, but also determination of the



influence of 0.002 per cent boron in the presence of chromium. The 0.002 per cent boron addition has been used because it corresponds to the amount usually added in foundry operations.<sup>1</sup>

## Dimensions of Test Bars

Cylindrical test bars, 16 mm in diameter and 100 mm in length (Fig. 1) were poured. Each bar had three notches which permit breaking into four small bars 25 mm in length. Each mold contained 20 cylindrical test bars. Two molds were poured from each of the ten heats of metal.

Chemical analysis and heat treatment tests were made on test bars from the same heat and from the same mold. Seven adjacent test bars from one mold yield 28 small bars of 25 mm in length for test purposes. Twenty-three bars have been used to develop the annealing cycles and five bars for chemical analysis. This arrangement assures homogeneity of chemical composition of different bars in each heat. This homogeneity has been verified by spectrographic analysis.

## Annealing Cycles Adopted

Most annealing furnaces used in France are heated with pulverized coal or coke. The test pieces were placed in annealing pots stacked four or five high in the furnace. The interspaces between the castings are filled



with red firebrick, sand and other refractory material which does not adhere to the castings at the annealing temperature and which protects the castings from oxidation by combustion gas and prevents their deformation or warpage.

This procedure gives the furnace a great thermal inertia. The period of rising temperature for annealing is, consequently, quite extended, as is the duration of cooling between the two stages of graphitization.

As the author hoped that results of the present study could be put to practical use by foundries, he has tried to simulate as much as possible industrial operating conditions. He has therefore adopted the time of temperature rise (Fig. 2) to the annealing temperature and the descent between the two stages of graphitization equal respectively to 24 hr and 12 hr.

The annealing was carried out in an electric furnace. The temperature of the first stage of graphitization was  $920\text{ C} \pm 10$  (1690 F). The temperature of the second stage of graphitization was  $740\text{ C} \pm 7$  (1360 F).

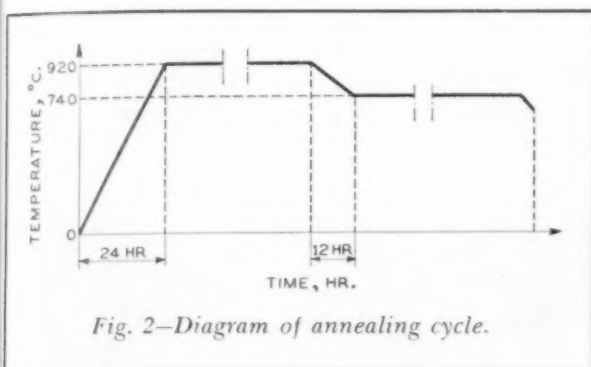


Fig. 2—Diagram of annealing cycle.

TABLE 2—HEATING TIME OF TEST BARS AT 920 C

Order of Removal from Furnace Test Bar Identity	Time at 920 C, Hr
A	2½
B	3½
C	4
D	5
E	6½
F	8
G	10
H	12½
I	16
J	20
K	25
L	31½

TABLE 3—HEATING TIME OF TEST BARS AT 740 C

Order of Removal from Furnace Test Bar Identity	Time at 740 C, Hr
M	10
N	12½
O	16
P	20
Q	25
R	31½
S	40
T	50
U	63
V	80
W	100

TABLE 4—TIME NECESSARY FOR DECOMPOSITION OF CARBIDES

Heat No.	Cr Content, per cent	First Test Bars Free From Cementite	Time for Decomposition of Carbides, Hr
1	0.032	1E	5 to 6½
2	0.082	2H	10 to 12½
3	0.099	3J	16 to 20
4	0.125	4J	16 to 20
5	0.130	5K	20 to 25
6b	0.035	6bF	6½ to 8
7b	0.056	7bG	8 to 10
8b	0.086	8bH	10 to 12½
9b	0.124	9bJ	16 to 20
10b	0.137	10bK	20 to 25

TABLE 5—TIME NECESSARY FOR DECOMPOSITION OF PEARLITE

Heat No.	Cr Content, per cent	First Test Bars Free From Pearlite	Time for Decomposition of Pearlite, Hr
1	0.032	1M	Less than 10
2	0.082	2U	50 to 63
3	0.099	3V	63 to 80
4	0.125	4W	80 to 100
5	0.130	5W	More than 100
6b	0.035	6bM	Less than 10
7b	0.056	7bN	10 to 12½
8b	0.086	8bQ	20 to 25
9b	0.124	9bT	40 to 50
10b	0.137	10bW	80 to 100

The test bars (16 mm in diameter and 25 mm in length) were packed in extra siliceous white sand in iron tubes 40 mm in diameter and 400 mm in length. The tubes were carefully luted at each end with refractory clay to avoid air seepage. Each tube contained a test bar from each of the ten heats studied.

A baffle constructed of refractory bricks had been placed at each side of the annealing furnace and the packed tubes were placed on the hearth in the center of the furnace. Thus temperature differences which would be present in the immediate vicinity of heating elements were avoided.

#### Time Cycle of Two Stages of Graphitization

Time intervals for each of the two temperatures of graphitization have been arranged according to a sensitive exponential law. Therefore, after the temperature of 920 C had been reached, the first 12 tubes were removed successively from the furnace at the end of the time intervals shown in Table 2.

After holding the furnace temperature for 31½ hr at 920 C (1690 F) it was dropped to 740 C (1360 F) in 12 hr. After this temperature had been reached the last 11 tubes were removed from the furnace at time intervals shown in Table 3.

As the test bars were removed from the furnace their microstructure was studied to determine the decomposition time of cementite and of pearlite. The influence of chromium and boron on the dimensions of graphite nodules was also studied.

Table 4 indicates time necessary for decomposition of the carbides. Table 5 indicates time necessary for decomposition of the pearlite. Table 6 indicates diameter of graphite nodules and the mean number of

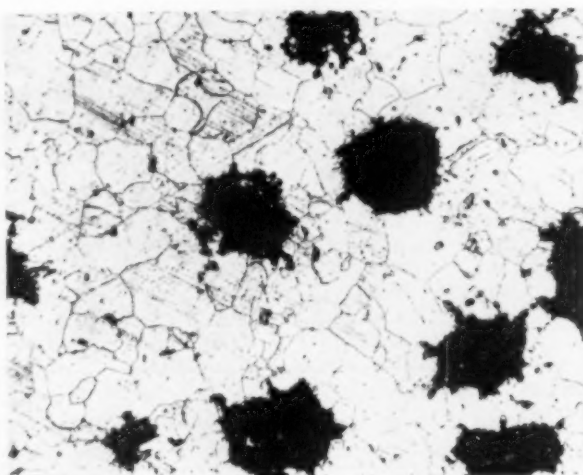


Fig. 3—Test bar 1M (Cr-0.03%).

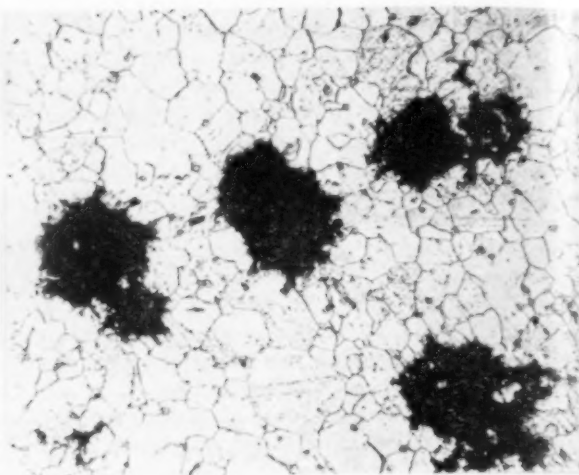


Fig. 4—Test bar 2U (Cr-0.08%).

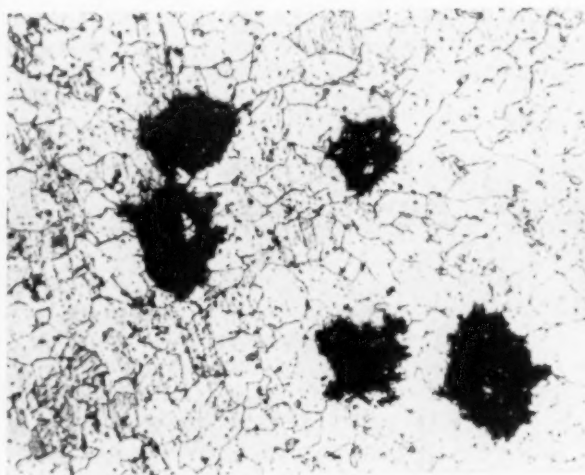


Fig. 5—Test bar 3V (Cr-0.09%).

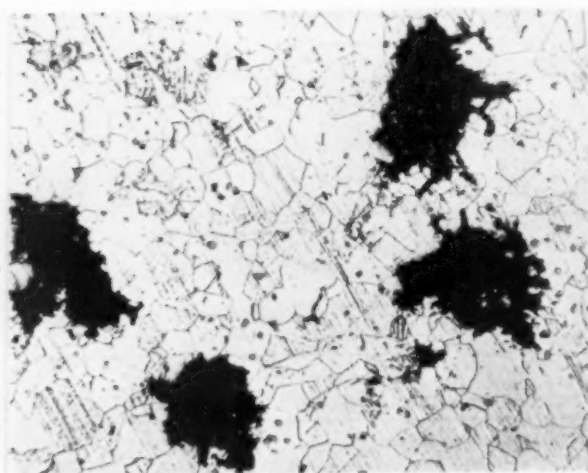


Fig. 6—Test bar 4W (Cr-0.12%).

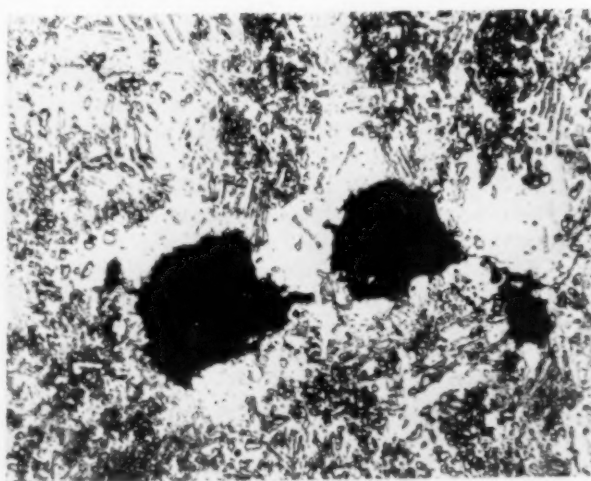


Fig. 7—Test bar 5W (Cr-0.13%).

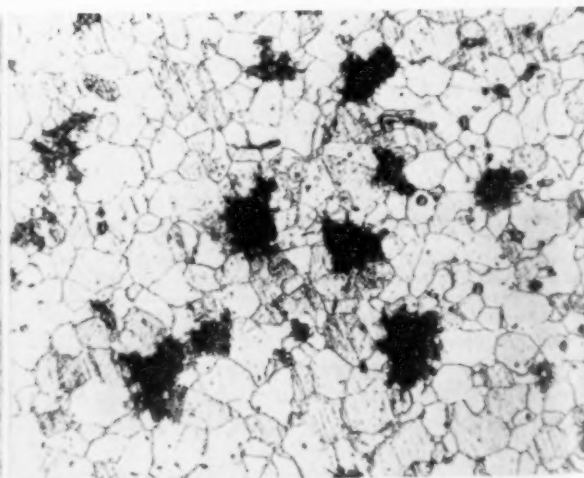


Fig. 8—Test bar 6bM (Cr-0.03%).

Microstructure of test bars. Mag. 150 $\times$ .

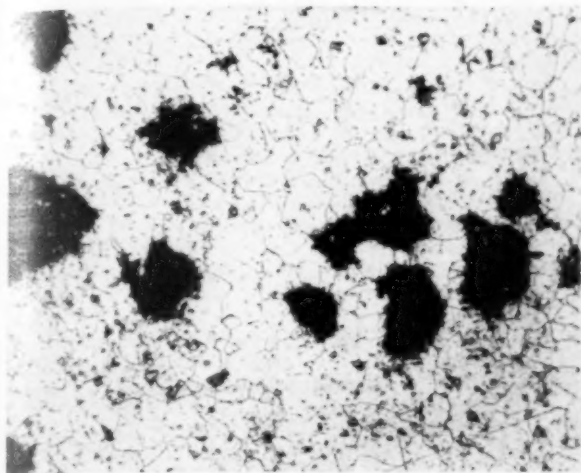


Fig. 9—Test bar 7bN (Cr-0.05%).

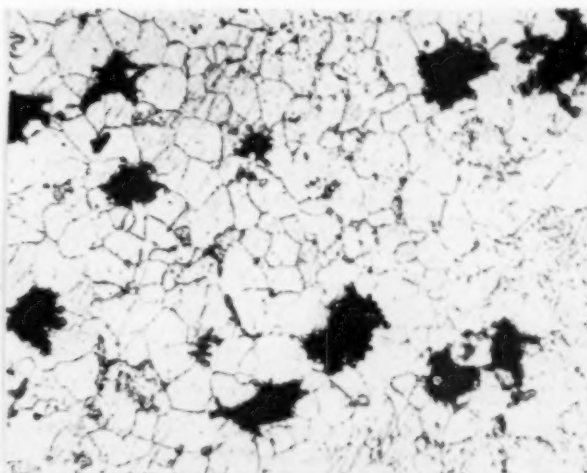


Fig. 10—Test bar 8bQ (Cr-0.08%).

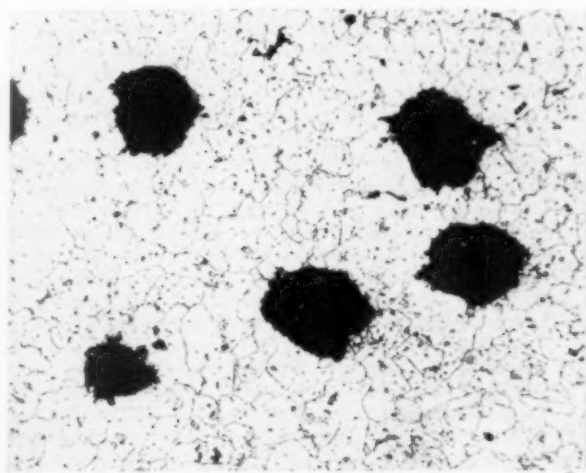


Fig. 11—Test bar 9bT (Cr-0.12%).

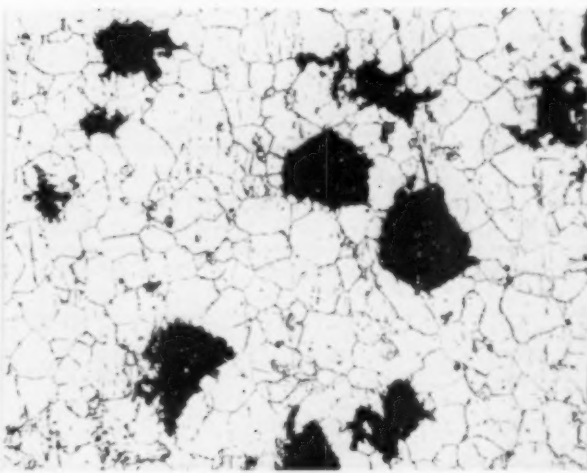


Fig. 12—Test bar 10bW (Cr-0.13%).

Microstructure of test bars. Mag. 150X.

nodules per sq mm. The photomicrographs at 150 X (Figs. 3 to 12) show the microstructure of the test bars.

### Conclusions

From examination of the results herein presented the author draws the following conclusions:

1. *Influence of chromium on the two stages of graphitization of malleable black-heart iron.* Chromium retards the rate of decomposition of cementite during the first stage of graphitization as well as the rate of decomposition of pearlite. Its retarding action is not as effective in the first stage of graphitization as in the second stage.

Many French foundries experience significant temperature differences between the extreme points in their furnaces, which lead them to adopt long periods at temperature for the first stage of graphitization. The time is at least 40 hr. The presence of chromium is not therefore a hindrance inasmuch as decomposition of cementite is concerned. On the contrary, the duration at temperature in the second stage of graphitization is, in general, short. In most cases it is less than 60

hr. When the chromium content reaches 0.06 to 0.07 per cent the test pieces contain pearlite. To overcome this inconvenience it would be necessary to prolong annealing to 100-hr or even 120-hr duration in the second stage of graphitization to obtain a normal structure even when the chromium content reaches 0.12 per cent.

2. *Action of boron in presence of chromium on duration of both stages of graphitization.* Boron in percentages less than 0.002 per cent added to the molten metal before pouring does not influence the rate of decomposition of carbides in the first stage of graphitization. On the other hand it reduces considerably the time necessary to effect decomposition of pearlite.

The diameter of the nodules of graphite is reduced at the same time their number is increased.

3. *Annealing cycles to adopt when metal contains chromium.* When boron has not been added to the melt prior to pouring it is necessary to anneal as shown in Table 7 to overcome the undesirable influence of chromium.

When 0.002 per cent boron has been added to the melt at the time of pouring, the figures of Table 7



TABLE 6—DIAMETER OF NODULES AND NUMBER OF NODULES PER SQ MM

Heat No.	Cr Content, per cent	Test Bar Identity	Mean Dia. of Graphite Nodules in mm	No. of Nodules per sq mm
1	0.032	1M	0.08	26
2	0.082	2U	0.13	16
3	0.099	3V	0.10	20
4	0.125	4W	0.10	16
5	0.130	5W	0.15	11
6b	0.035	6bM	0.05	29
7b	0.056	7bN	0.06	29
8b	0.086	8bQ	0.07	29
9b	0.124	9bT	0.06	28
10b	0.137	10bW	0.06	37

TABLE 7—TIME OF GRAPHITIZATION DURING BOTH STAGES OF ANNEALING FOR IRON WITHOUT BORON

Cr Content, per cent	Time at 920 C, Hr	Time at 740 C, Hr
0.03	7	10
0.06	12½	45
0.08	16	65
0.10	20	80
0.12	25	100
0.14	30	More than 100

TABLE 8—TIME OF GRAPHITIZATION DURING BOTH STAGES OF ANNEALING FOR IRON WITH 0.002% BORON

Cr Content, per cent	Time at 920 C, Hr	Time at 740 C, Hr
0.03	7	10
0.06	12½	15
0.08	16	30
0.10	20	40
0.12	25	50
0.14	30	100

can be modified to conform to those shown in Table 8.

These annealing cycles hold only for the foundry that the author studied and for the temperatures that he has chosen for the two stages of graphitization. They are not applicable if one modifies the factors that affect rate of malleabilization, some of which are noted in the following:

- Silicon content of white cast iron,
- Annealing temperature,
- Manganese content in relation to sulphur,
- Carbon content,
- Furnace atmosphere,
- Degree of superheating of metal before pouring,
- Rate of cooling after pouring,
- Source of the pig iron.

The preceding results can be useful to foundries melting and refining a different metal. They should, when their first composition contains chromium, guard against its retarding action by the addition of boron at the time of pouring and in prolonging sufficiently the second stage of graphitization.

### Reference

<sup>1</sup> N. F. Tisdale, "Boron in Malleable Iron," *The Foundry*, pp. 107, 222, April (1945).

### DISCUSSION

Chairman: W. D. McMILLAN, McCormick Works, International Harvester Co., Chicago

Co-Chairman: W. B. McFERRIN, Electro Metallurgical Co., Detroit

N. F. TISDALE: <sup>1</sup> I have had considerable correspondence over a period of three years on this subject. I do not entirely agree with the author when he says that boron does not affect the first stage of graphitization because we do get a slight effect. Our main difference in working with the author was that he remelted in the crucible. He took a synthetic iron, whereas we started out from iron originally poured from the furnaces.

In regard to annealing time, he has given a factor of safety in comparing those irons with and without boron.

We did not say originally that 0.002 per cent boron was the figure. We tried to explain that we work it according to chromium content, and today the only time a recommendation of 0.002 per cent boron is given is when the chromium content exceeds the amount which the author has shown as 0.012 per cent. Most of our additions are listed as 0.001.

In addition to this, in his earlier work the author had some trouble in melting, with his type of melting, so his first conclusions have been varied to put in those exceptions which are contained in his paper and we do not have the same conclusions he has. For instance, the annealing temperature which he used is not exactly the same as common practice. We cannot find any relation between annealing time and his furnace atmosphere, except to reduce the carbon content. While in the main we are in agreement with the author we have to make some exceptions particularly in his statement that there is no difference in the first stage of graphitization because we find that there is.

F. COGHLIN, JR.: <sup>2</sup> Mr. Tisdale brought out the point that these facts and figures were based on findings of varying chromium contents on a synthetic iron. We have actually done some work on the problem with our regular run malleable iron. Being faced with chromium difficulties from time to time, as many other malleable foundries are, the opportunity for running tests in production has presented itself and has revealed at least a little information. We found in actual practice, with 0.086 chromium iron and 1.50 silicon iron, that the time used for breaking down the carbides was considerably more than described in the paper, amounting to approximately 20 hr. On the other hand, decomposition of the pearlite, we found required considerably less time than described, our average being approximately 10 hr. However, inasmuch as this is a synthetic iron and ours is not, I cannot account for the effect of residual inhibitors on the comparative findings.

MILTON TILLEY: <sup>3</sup> Our experience checks the author's in a general way, but I cannot say what effect we would find on the first stage of graphitization since we did not bother with the first stage. On the second stage, we find that the effect is favorable up to about the point arrived at by the author and then it does not seem to do us any good with the time that we have in the critical, which is about 25 hr. I think it might clear this up a little bit if we knew how much was in the metal, not what was added. You may add 0.002 per cent boron and get only 0.0005 per cent. We should know how much we get in before we talk too much in detail about what the effect is.

Co-CHAIRMAN McFERRIN: That is right.

J. H. LANSING: <sup>4</sup> I think this paper is very interesting. Discussing the effect of chromium and of boron is always interesting. At the same time I believe we should point out in any discussion of this sort the possible effect of boron on the physical properties, especially on the impact value. A substantial amount and even some of the amount of boron that have been discussed here, at times have an adverse effect on the physical properties, especially on the impact value of malleable iron. If we do not put in that qualifying statement, some are likely to become too enthusiastic in the use of too substantial amounts of boron for any troubles they may encounter and seriously affect their impact and other values.

MR. TISDALE: It depends again on the composition of your malleable iron to start with. There are cases where as much as 0.003 per cent boron has been added without adversely affect-

<sup>1</sup> Molybdenum Corp. of America, Pittsburgh

<sup>2</sup> Albion Malleable Iron Co., Albion, Mich.

<sup>3</sup> National Malleable & Steel Castings Co., Cleveland

<sup>4</sup> Malleable Founders' Society, Cleveland

ing the physical properties, but again on another composition of iron, the impact value would drop down to a point where the malleable would be of poor quality.

When you go through the standard procedure for controlling it, the exact analysis can be obtained and can be had regularly. We do it ourselves by spectrograph when the boron content is taken into consideration with how much is in the metal. When you put returns back into your circuit, you have a slight build-up of boron. That must be taken into consideration in the amount that you add. If you just talk of added boron, you definitely can get into difficulties, but on the other hand, if, by constant practice and analysis, you know how much is in your base metal, it is not too difficult then to keep it within bounds. So long as you keep it less than 0.0015 per cent, I have never seen any bad results in connection with the addition of boron on physical properties.

MR. LANSING: Generally speaking, Mr. Tisdale's comment is a fair and comprehensive one. I would especially stress again the point he made that, in any additions of this sort, you not only watch the results from the standpoint of what you are obtaining metallographically, but also from the standpoint of actual tests, especially impact tests upon the metal.

W. R. BEAN:<sup>6</sup> In some of our past work on the effects of elements, I reached the tentative conclusion that any element which aids graphitization is to a certain extent detrimental to the physical properties of malleable iron. That is definitely true of silicon and carbon both essential elements in malleable iron. It is definitely true of aluminum. I am not prepared to express a conviction on boron because for the past several years I have not been closely enough in contact with what is being done in malleable metallurgy having spent the years in Washington D. C. rather than in the malleable foundry. But I do think that the effect of elements which are used in malleable to counteract adverse tendencies of other elements should be given serious consideration in routine practice.

MR. TILLEY: We have found in our shop, where we run 2.5 per cent carbon and 1 per cent silicon, that if boron content of the metal by spectroscopic analysis gets over 0.002 per cent, the carbon spot becomes decidedly smaller and therefore there is a little deterioration. We also notice carbide entering into the metal. Presumably, it is some boron carbide of some kind.

<sup>6</sup> Whiting Corp., Harvey, Ill.

# PRODUCTION OF NODULAR GRAPHITE STRUCTURES IN GRAY CAST IRONS

By

H. Morrogh \*

## ABSTRACT

A method has been developed for the production of nodular graphite structures in cast irons without the necessity for applying heat-treatment as in the malleable cast iron process. Irons with these nodular structures are termed nodular cast irons. The process involves the treatment of a low sulphur hypereutectic cast iron with cerium shortly before casting. The cerium is conveniently added as misch metal, and functions first as a desulphurizer and second as a carbide stabilizer. With cerium treatment alone, only the hypereutectic graphite is nodular and the remainder of the graphite is in a new pattern referred to as "quasi-flake graphite." The hypereutectic nodules are of the spherulitic variety. Nodular cast irons with all of the graphite in the spherulitic form are obtained by treatment of the molten metal with cerium, followed by treatment with a graphitizing inoculant, such as ferro-silicon or silicon-manganese zirconium.

The process can be applied to irons of a wide range of composition, and examples have been given to illustrate the effects of some of the most important variables. Nodular cast irons may be produced from a variety of melting units, including the cupola. Extremely good mechanical properties may be produced at low hardness levels. Nodular cast irons have properties midway between high-duty gray cast irons and malleable cast irons.

RECENT RESEARCH WORK conducted in the laboratories of the British Cast Iron Research Association has been devoted, in part, to the investigation of the mechanism of graphite formation in cast iron. Arising from this work has been the development of a method (patents applied for in United States, Great Britain, and elsewhere) by which gray cast irons, having nodular graphite structures in the as-cast state, can be obtained without the necessity for applying any heat-treatment process subsequent to the solidification of the casting in the mold. Such irons, which may have all or part of their graphite carbon contents in the nodular form, have been termed "nodular cast irons." It is the purpose of this paper to give a short account of this process and to illustrate the properties and the typical structures of the new material.

Preliminary work leading to this discovery has been previously described by Morrogh and Williams<sup>1, 2</sup> and recently a detailed account has been given of the development of the process and the theories behind it.<sup>3</sup>

Ordinary gray cast irons have their graphitic carbon distributed through the metallic matrix in the form of flakes, the size and shape of which vary considerably

according to composition, cooling rate, method of melting, ladle treatment, etc. These graphite flakes interrupt the continuity of the metallic matrix and so render the material relatively brittle and non-ductile. In hypoeutectic cast irons two distinctly different forms of graphite flake distribution are possible.

A considerable measure of controversy exists as to the nomenclature of these graphite patterns, but the present author will refer to them as "normal flake graphite" for the one variety and as "undercooled graphite" for the other variety. There is substantial experimental evidence<sup>1, 2, 4</sup> to indicate that normal flake graphite is deposited directly from the melt along with austenite, and that undercooled graphite forms after solidification as the result of the decomposition of cementite.

In malleable cast irons, which are produced by the annealing of a white cast iron, the graphite carbon, commonly termed temper carbon, exists in roughly spheroidal aggregates or nodules. These temper carbon nodules, by virtue of their approximately spheroidal shape, do not interrupt the continuity of the metallic matrix to the same extent as flake graphite and so, if the total carbon content is sufficiently low, a material of relatively good shock-resistance and ductility may be obtained. The author has shown<sup>5</sup> that two different forms of temper carbon nodule can occur in malleable cast iron according to the composition of the white iron being annealed.

## Graphite Dispersion Varies

When the white iron has all its sulphur predominantly in the form of manganese sulphide, each nodule in the resulting malleable iron consists of an apparently random aggregate of small graphite flakes, the degree of dispersion of which may vary between wide limits. Lacking a better term, nodules of this type were referred to as "graphite-flake-aggregate" nodules, and a typical example of this structure is shown in Fig. 1.

Nodules of this type are commonly found in blackheart malleable cast iron, where the sulphur content usually is more or less balanced by the manganese content. However, in European whiteheart malleable cast iron, the manganese is frequently inadequate to balance the sulphur and so iron sulphide occurs in the material. Under these conditions the temper carbon

\* British Cast Iron Research Association, Birmingham, England.



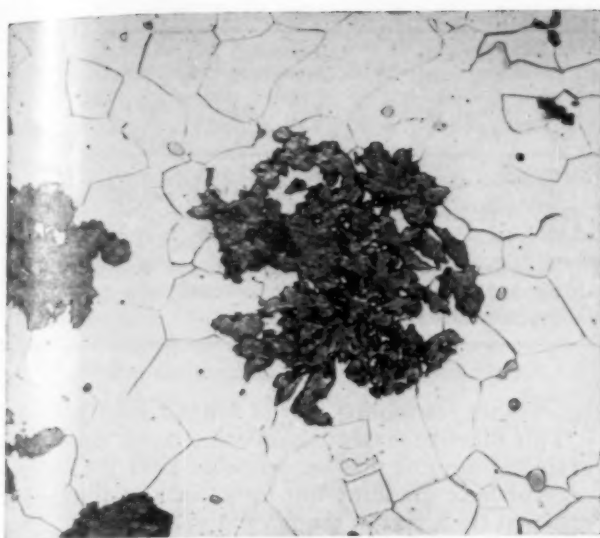


Fig. 1—Typical example of "graphite-flake aggregate" nodule in malleable cast iron having sulphur present as manganese sulphide. Etched in 5 per cent nital.  $\times 300$ .

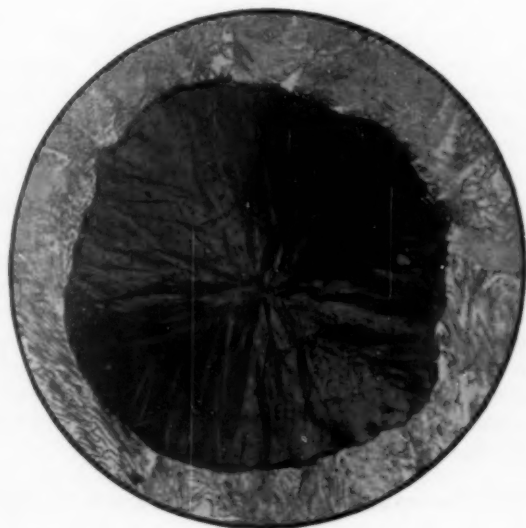


Fig. 2—Spherulitic nodule in malleable iron having sulphur present as FeS. Etched in picric acid.  $\times 350$ .

has a spherulitic structure, a typical example of which is illustrated in Fig. 2. Spherulitic nodules are obtained in nodular cast irons and it is therefore of some interest and importance to understand and appreciate this graphite structure.

At this point it should be observed that these structures can be observed only in microspecimens in which the graphite itself is polished. A method for accomplishing this has been described, and the special properties of graphite which permit the internal structure to be observed have been discussed in some detail.<sup>6</sup>

Spherulitic structures are well known to the mineralogist, and graphite spherulites conform to the general structural characteristics common to all other spherulites. Nodules of graphite having this structure each consist of an aggregate of graphite crystallites radiat-

ing from a common center or nucleus. The closely packed basal planes of the graphite crystallites are, in general, oriented at right angles to the radii of the spheroid. Graphite spherulites tend to be more perfectly spheroidal in shape than the flake-aggregate nodules. It is important to reserve the term "spherulite" for aggregates having the radial fibre structure as illustrated in Fig. 2. Subsequently in this paper the term will have this connotation only.

#### Problem and Solution

It has long been the view that the mechanical properties of gray cast iron could be considerably improved if the graphite could be obtained in the nodular instead of the flake form. To do this would at first seem to require an iron which would solidify white and subsequently graphitize at some relatively low temperature after solidification. This condition is partly fulfilled in irons having undercooled graphite structures—they solidify white and graphitize very shortly after solidification.

Studies of the analogous nickel-carbon and cobalt-carbon systems<sup>1, 2</sup> enabled the author and his co-worker, Williams, to predict that it should be possible to obtain nodular structures, in cast irons which undercool, by the addition of a carbide stabilizing element so that carbide decomposition is retarded.

Furthermore, work on the analogous systems indicated that the element should be a desulphurizer and that low sulphur contents were likely to assist in achieving the desired structure. After a somewhat unusual method of investigation (which has been described elsewhere in detail<sup>3</sup>) it was found that additions of cerium to cast iron of appropriate composition would produce spherulitic nodular graphite structures.

#### Process Developed

The method which has been developed for the production of nodular cast irons involves in its simplest form the addition and solution of an appropriate amount of cerium to a molten cast iron of appropriate composition shortly before casting. The principal composition requirements for the production of nodular irons by this process are:

1. The iron must solidify gray even without the cerium addition.
2. The iron must be of hypereutectic carbon content, that is, the carbon content should exceed the value  $4.3 - \frac{1}{3} (\text{per cent Si} + \text{per cent P})$ . When the nickel content of an iron exceeds 10 per cent, it need not be hypereutectic according to this formula.
3. Silicon content can have any value, but is preferably within the range 2.3–7 per cent.
4. Sulphur content of the metal to be treated should be as low as possible, and after treatment should be below about 0.015 per cent.
5. Phosphorus content should not exceed about 0.6 per cent, and should preferably be below 0.1 per cent.
6. Manganese, copper, nickel, chromium and molybdenum may be present in any amounts, singly or in any combination, provided condition No. 1 is observed.
7. After treatment with cerium the solidified castings must contain more than a certain minimum amount (0.02 per cent) of the element dissolved in the metallic matrix.

Of the foregoing requirements the most important from the point of view of the successful use of the process are those covering the carbon and sulphur contents.

TABLE 1—MECHANICAL PROPERTIES OF UNTREATED<sup>1</sup> AND CERIUM-TREATED<sup>2</sup> CAST IRONS

Test Bar Diameter, in.	Transverse Rupture Stress, psi		Deflection, in.		Tensile Strength, psi		BHN		Impact Strength, ft-lb		Compression Strength, psi		Shear Strength, psi	
	W/O Ce	With Ce	W/O Ce	With Ce	W/O Ce	With Ce	W/O Ce	With Ce	W/O Ce	With Ce	W/O Ce	With Ce	W/O Ce	With Ce
1.6	51,500	99,000	0.20	0.32	25,100	55,050	154	186			71,600	127,100		
1.2	64,500	101,500	0.28	0.38	32,050	55,500	160	198					36,500	49,700
0.875	68,870	105,400	0.18	0.23	37,200	59,550	162	199	12	43				
0.6	68,900	128,500	0.11	0.22	41,650	69,020	198	239			115,750	153,000		

<sup>1</sup> Analysis of remelted pig iron: total carbon 3.77; silicon, 3.05; manganese, 0.73; sulphur, 0.023; phosphorus, 0.039 per cent.

<sup>2</sup> Remelted pig iron with cerium additions: total carbon, 3.72; silicon, 3.13; manganese, 0.74; sulphur, 0.007; phosphorus, 0.038; cerium, 0.040 per cent.

The influence of cerium in cast iron has been investigated and discussed in the literature on several previous occasions.<sup>7-18</sup> Much of this evidence is conflicting, but two fairly well-established facts emerge from its study—cerium may be a potent carbide stabilizer under certain conditions, and when added to sulphur-containing alloys it is an effective desulphurizer. No previous investigators have claimed that nodular structures could be produced in cast irons by the addition of cerium, and this is understandable because in no case have all the composition requirements set out in the foregoing been met.

#### Cerium in Cast Iron

As a result of the present author's work, it has become apparent that the first major effect of cerium when added to molten cast iron is to combine with the sulphur to form a cerium-sulphur compound which floats to the surface of the metal. This accounts for the desulphurizing effect of the element. As long as the sulphur content of the metal is above about 0.015 per cent this desulphurizing action will take place. Cerium is not free to alloy with the metal until the sulphur content has reached this low value. The higher the sulphur content of the metal, the larger is the amount of cerium required to achieve desulphurization, and hence to obtain a given amount of cerium not combined with sulphur dissolved in the metal.

It is possible that deoxidation also accompanies desulphurization. At temperatures within the range of 1200-1600 C desulphurization with cerium proceeds quite rapidly, and relatively large quantities of metal may be desulphurized by this element in a few seconds.

After the sulphur content has been reduced to a value of about 0.015 per cent, metallic cerium enters into solution in the molten cast iron, and this cerium, when present in amounts greater than about 0.02 per cent, functions as a powerful carbide stabilizer. It is this dissolved cerium in excess of 0.02 per cent which is the operative factor in the process for the production of nodular graphite structures in the as-cast state.

#### Cerium Additions to Hypereutectic Cast Irons

When cerium is added to hypereutectic cast irons, and when the final sulphur content is below about 0.015 per cent and the cerium content of the metal is above about 0.02 per cent, all the hypereutectic graphite is obtained as well-formed spherulites. When the silicon content of such an iron is sufficiently high to prevent formation of the white iron eutectic, the graphite originating from the eutectic complex has an

appearance resembling that of normal flake graphite.

This flakelike graphite has been termed "quasi-flake graphite" since it may on occasions have the appearance of flake graphite, but forms in a manner quite different from that of the normal variety. This difference will become more apparent in later sections of this paper.

To illustrate this effect of cerium on hypereutectic cast irons, the following example may be quoted. A 50-lb charge of hematite pig iron of the percentage analysis, total carbon, 3.98, silicon, 3.19, manganese, 0.78, sulphur, 0.028, and phosphorus, 0.040, was melted in a crucible furnace and cast, at a temperature of 1380 C, into test bars of four different sizes.\* A further similar melt was carried out, but in this case a 28-gram addition of pure cerium was added to the molten metal before casting. Mechanical properties and chemical analyses of the two sets of bars are shown in Table 1.

Microstructures of all the untreated bars show mixtures of undercooled and normal flake graphite in matrices of ferrite with some pearlite—the amount of flake graphite increasing and the amount of pearlite decreasing with increasing section size. The microstructure of the untreated 1.2-in. bar is shown in Fig. 3. The treated bars all showed hypereutectic spherulites, together with quasi-flake graphite in a matrix of ferrite with some pearlite. The structure of the treated 1.2-in. bar is shown in Fig. 4.

#### Solidification of Cerium-Treated Hypereutectic Irons

The structure shown in Fig. 4 is, judged by previous standards, completely novel and, bearing in mind the distribution of the graphite, it is easily understandable that the mechanical properties of the treated material are at a distinctly higher level than those of the untreated material. It is to be observed in Table 1 that, even with this simple cerium addition to a high carbon and relatively high silicon iron, a material is obtained which immediately falls into the category of a high-duty cast iron.

Mechanism of solidification of the type of structure illustrated in Fig. 4 has been studied and a fairly clear picture of the process obtained.<sup>3</sup> The hypereutectic spherulitic nodules form in the melt before solidification of the metallic phases begins. It is not known for

\* The transverse, deflection and tensile figures given in this paper for 2.1, 1.6, 1.2, 0.875 and 0.6-in. diameter bars were obtained on bars cast and tested according to British Standard Specification 786/1938; the impact figures were obtained on bars tested according to British Standard Specification 1349/1947; the 3-in. diameter bars were tested in transverse on 18-in. centers.

cern whether these graphite spherulites are deposited directly from the melt or whether they form by the decomposition of a carbide phase. While the possibility of the latter has to be considered, it is not easy to imagine and it involves a case of extremely rapid carbide decomposition.

The remainder of the melt solidifies as a "eutectic" or more correctly, a binary complex of austenite and cementite which decomposes shortly after solidification to give the quasi-flake graphite. The presence of the hypereutectic spherulites appears to initiate the rapid decomposition of the "eutectic" cementite after solidification of the iron.

It will be observed in Fig. 4 that the hypereutectic nodules are surrounded in each case by a volume of metal entirely free from the quasi-flake graphite. The reason for this becomes apparent when these nodules are examined at a high magnification, when each is found to have, in addition to the characteristic radial spherulitic structure, a duplex structure consisting of a central spherulite of nodular graphite surrounded by a peripheral layer of graphite.

Such a duplex structure is seen in Fig. 5, which shows a hypereutectic spherulite in a cerium-treated iron at a fairly high magnification. This micrograph has been taken under plane polarized light. The central core of this duplex nodule represents the actual hypereutectic graphite, and the peripheral layer the graphite arising from the decomposition of the surrounding "eutectic" cementite which has crystallized onto the hypereutectic spherulite. This explains the absence of quasi-flake graphite in the vicinity of each nodule in Fig. 4.

Each hypereutectic nodule appears to have a "sphere of influence" within which all further graphite is made to crystallize on the existing hypereutectic nucleus. This thought provides the clue to the complete development of the cerium process for the production of nodular cast irons by which the whole of the graphite can be obtained in the nodular form.

This final stage of the process will be described later in this paper but, following what has been said so far in this section, it will be seen that if the number of hypereutectic spherulites in a given volume of metal

could be so increased that all their "spheres of influence" overlapped, then all of the graphite should occur in the form of duplex spherulitic nodules with a further improvement in mechanical properties. A preliminary indication of the validity of this latter idea is found by studying the structures of centrifugally cast cerium-treated cast irons.

When cerium-treated hypereutectic cast irons are centrifugally cast, the hypereutectic spherulites forming in the liquid are forced to points most distant from the mold face, by virtue of their low relative density. In the centrifugal casting of cylinder liners, for instance, a segregation of these nodules is obtained along the inner surface of the casting. By this segregation the number of hypereutectic nodules is thus artificially increased so that their "spheres of influence" overlap.

Figure 6 shows a typical segregation of these hypereutectic spherulites near the inner surface of a centrifugally cast cylinder liner made in a cerium-treated hypereutectic cast iron. It will be seen that no quasi-flake graphite occurs in the vicinity of this segregation.

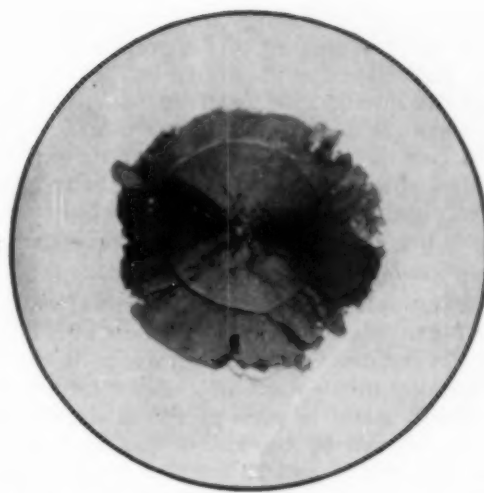
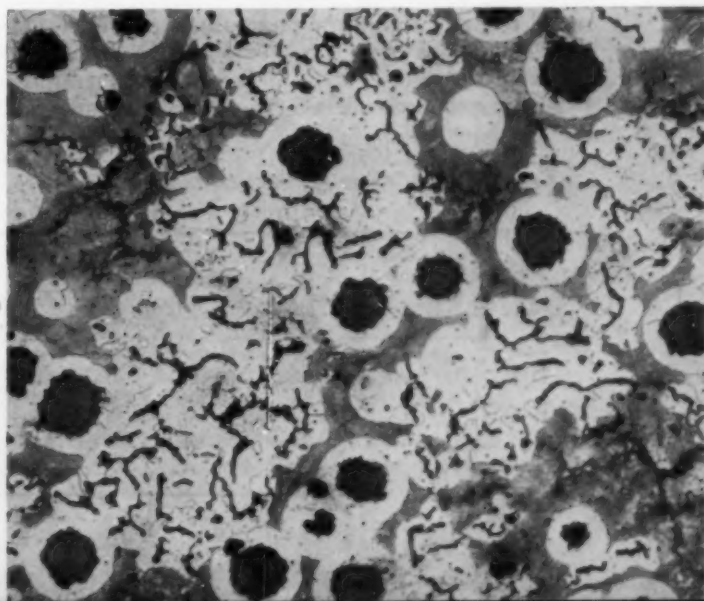
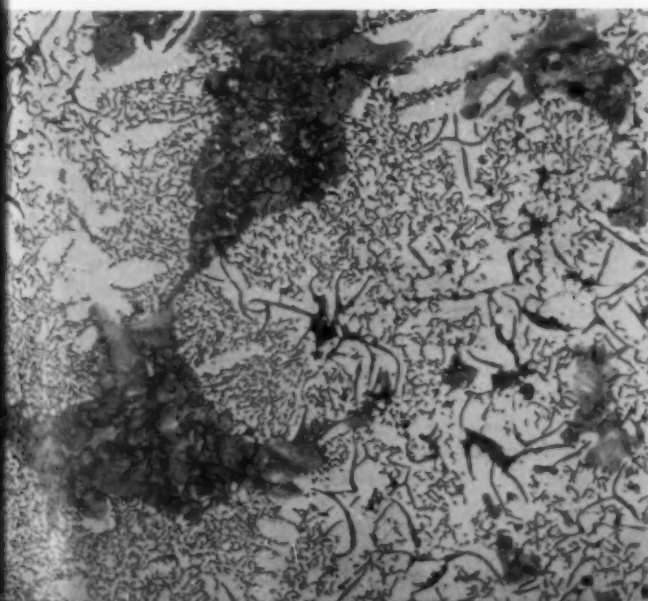


Fig. 5—Duplex structure of hypereutectic spherulite present in cerium-treated iron. Unetched.  $\times 1500$ .

Fig. 3—General structure of untreated remelted pig iron, 1.2-in. test bar. Etched in picric acid.  $\times 100$ .

Fig. 4—Microstructure of cerium-treated remelted pig iron, 1.2-in. test bar. Etched in picric acid.  $\times 100$ .





At points nearer the mold face containing very few hypereutectic nodules, all of the graphite was in the form of quasi-flakes, as illustrated in Fig. 7.

These two micrographs clearly indicate that entirely spherulitic nodular graphite structures can be achieved by increasing the number of hypereutectic spherulites. This ideal structure of spherulitic nodular graphite has been obtained by the use of what will be referred to later as the "double-treatment process."

#### Adding Cerium to Cast Iron

In the example already given the cerium was added to the molten cast iron in the form of the pure element. In this form cerium is a very expensive addition, but it can be added in a variety of cheaper forms, among which misch metal is perhaps the most convenient. Misch metal used by the author has been found to contain between 43 and 50 per cent cerium, together with the other rare earths and a little iron and manganese. The presence of these other elements does not influence the efficacy of cerium in producing nodular structures in cast irons, and for this purpose they can be ignored.

Misch metal, in pieces of appropriate size, dissolves readily in cast irons at all temperatures above 1200 C. Its solution in cast iron is not explosive or violent in any way. Misch metal has been used as a cerium addition in all the remaining examples quoted.

#### Cerium Additions—Influence of Amount

Cerium not combined with sulphur in cast iron is a powerful carbide stabilizer, and so, for any given section size, there is an upper limit of cerium which must not be exceeded if white iron structures are to be avoided. This remark applies only to the cerium content as found by analysis of the solidified casting when the sulphur content is not in excess of about 0.015 per cent. Because of the carbide stabilizing influence of cerium its effect can best be studied in relatively large sections.

To illustrate the effect of increasing additions of cerium, 270 lb of a pig iron of the following composition was melted in an oil-fired crucible furnace: total carbon, 3.79; silicon, 2.80; manganese, 0.53; sulphur, 0.028, and phosphorus, 0.015 per cent.

Five test bars, each of 21-in. length and 3-in. diameter, were cast in green sand molds. For this, five lots of metal each weighing 50 lb were taken from the furnace. No addition was made to the metal in the first ladle, but increasing amounts of misch metal were added to the metal in the remaining four ladles. The actual additions were: Tap 1—no addition; Tap 2—50 grams misch metal; Tap 3—65 grams misch metal; Tap 4—90 grams misch metal; Tap 5—100 grams misch

TABLE 2—MECHANICAL PROPERTIES AND CERIUM ANALYSES

Bar No.	Transverse Rupture Stress,* psi.	Deflection,* in.	Tensile Strength, psi	BHN	Ce, percent
1	30,470	0.21	15,000	98	Nil
2	85,080	0.21	40,100	167	0.040
3	104,100	0.35	50,200	176	0.053
4	106,600	0.34	53,800	181	0.072
5	115,600	0.42	58,850	179	0.101

\* Tested on 18-in. centers.

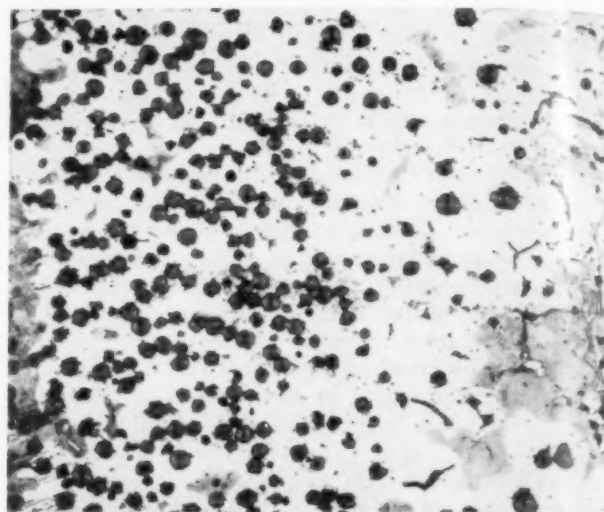


Fig. 6—Typical segregation of hypereutectic spherulites at inner surface of centrifugally cast cylinder liner (cerium-treated iron). Etched in picric acid.  $\times 100$ .

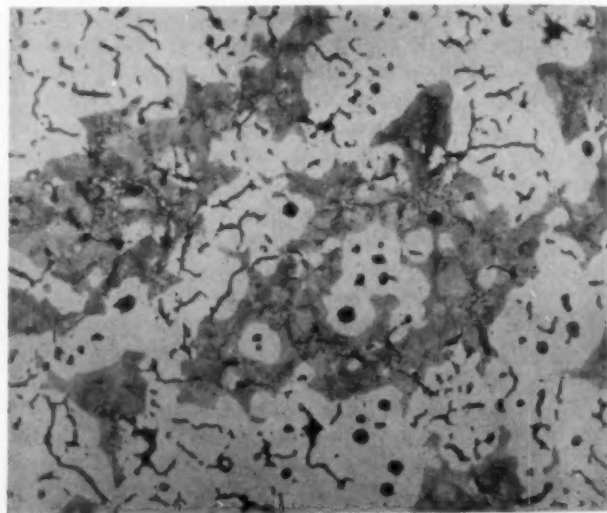
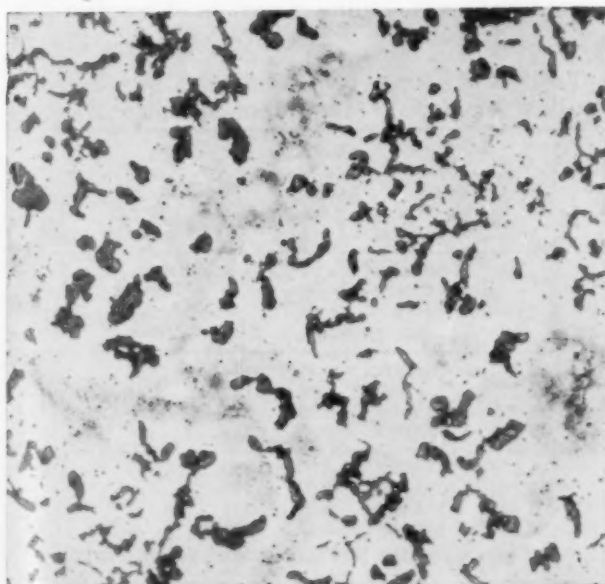
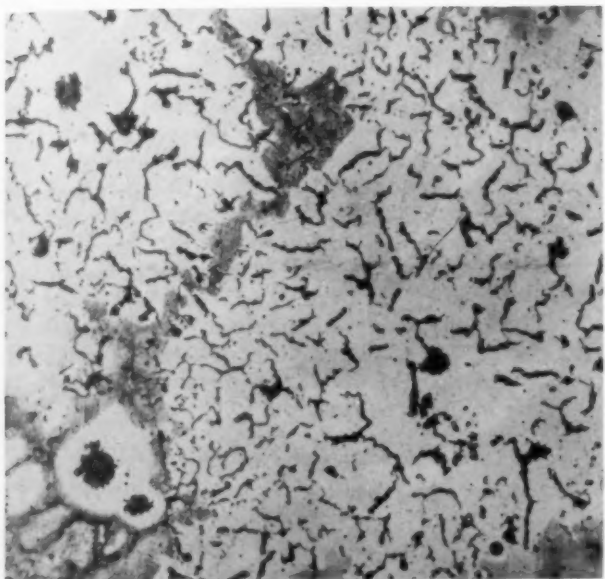
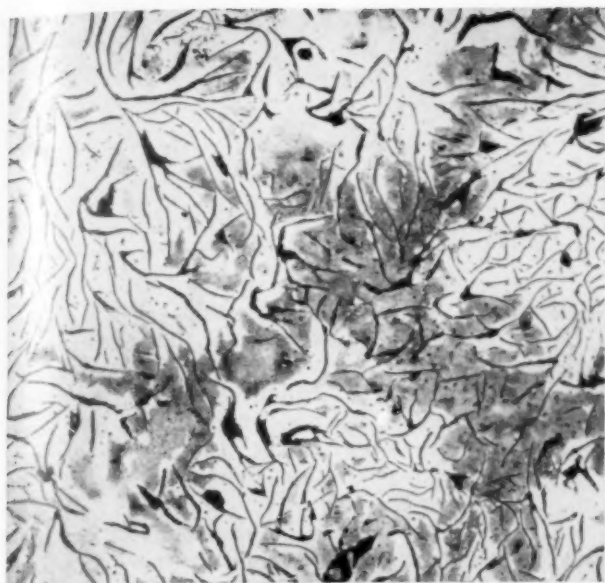


Fig. 7—Centrifugally cast liner. Structure at a point distant from inner surface. All graphite is in form of quasi-flakes. Etched in picric acid.  $\times 100$ .

metal. The mechanical properties and cerium analyses of these bars are given in Table 2.

Transverse rupture stress, deflection and tensile strength show a progressive improvement with increasing cerium content. The microstructures of these bars also showed a progressive change with increasing cerium content. The untreated bar No. 1 had coarse flake graphite in a matrix of ferrite with some pearlite, as is shown in Fig. 8; bar No. 2 had a few hypereutectic spherulites with quasi-flake graphite, as illustrated in Fig. 9; with increasing amounts of cerium in bars Nos. 3, 4 and 5 the quasi-flake graphite became more nodular in appearance until bar No. 5 had the structure shown in Fig. 10.

That quasi-flake graphite differs from normal flake graphite is made clear when Figs. 8 and 9 are com-



pared, and when the enormous difference in mechanical properties between bars Nos. 1 and 2 are considered. The relative amounts of ferrite and pearlite did not change appreciably with increasing cerium content, but in spite of this there was a sharp increase in hardness when the structure changed from the normal to the quasi-flake graphite.

#### Double-Treatment Technique Developed

The influence of relatively large amounts of cerium in causing the aggregation of the quasi-flake graphite can be utilized only in relatively large sections. For thin sections, alternative methods must be applied to obtain an entirely nodular structure. It was shown earlier in this paper that, if the number of hyper-eutectic nuclei could be increased, all of the graphite would occur in spherulitic nodular form and the quasi-flake graphite would be avoided.

A process for the production of this ideal structure has now been developed, entailing the addition of a graphitizing inoculant simultaneously with, or preferably after, the cerium addition. This treatment of the molten metal will be referred to as the "double treatment" process. Suitable graphitizing inoculants have been found to be 80 per cent ferro-silicon, silicon-manganese-zirconium, and calcium silicide, the best response having been obtained with the first and second. It is important that the inoculant is not added before the cerium addition as the success of the operation depends upon obtaining the solution of the cerium before the inoculant.

Owing to the extremely rapid rate of solution of misch metal in cast iron, this condition may be achieved even when the inoculant and misch metal are added simultaneously. For the inoculant to have the required effect it is necessary to add sufficient, in the case of silicon-manganese-zirconium or ferro-silicon, to give a definite increase in silicon of not less than about 0.2 per cent, but no useful purpose is served by using amounts giving silicon yields in excess of 0.5 per cent.

Results given in Table 3 illustrate the effect of adding varying amounts of 80 per cent ferro-silicon for the same addition of misch metal in each case. These figures are average values, obtained from four 1.2-in. bars cast in each case from 50 lb of metal treated as follows: Tap 1—70 grams misch metal; Tap 2—70 grams misch metal + 2 oz 80 per cent ferro-silicon; Tap 3—70 grams misch metal + 4 oz 80 per cent ferro-silicon; Tap 4—70 grams misch metal + 8 oz 80 per cent ferro-silicon.

Transverse rupture stress, tensile strength and impact strength show a marked increase with the first addition of inoculant (compare bars 1 and 2). The impact results are particularly striking—these figures are obtained on a 0.798-in. diameter unnotched test piece which rarely gives a value in excess of 30 ft-lb for high-duty gray cast iron, and only occasionally a value as high as 50 ft-lb for the acicular gray cast iron.

Fig. 8 (top)—Untreated remelted hematite pig iron cast in 3-in. diameter bar. Etched in picric acid.  $\times 60$ .

Fig. 9 (center)—Remelted pig iron with 0.040 per cent Ce, cast in 3-in. diameter bar. Picric acid etch.  $\times 60$ .

Fig. 10 (bottom)—Remelted hematite with 0.101 per cent Ce, cast in 3-in. diameter bar. Picric acid etch.  $\times 60$ .

TABLE 3—EFFECT OF 80 PER CENT FERRO-SILICON ADDITIONS

Tap No.	Composition, per cent						Transverse Rupture Stress, psi	Deflection, in.	Tensile Strength, psi	BHN	Impact, ft-lb
	T.C.	Si	Mn	S	P	Ce					
1	3.67	2.67	0.86	0.005	0.051	0.040	127,200	0.60	58,200	221	54
2	3.52	2.73	—	—	—	0.051	138,000	0.60	71,000	233	91
3	3.49	2.89	—	—	—	0.058	138,900	0.60	74,800	238	84
4	3.50	3.03	—	—	—	0.061	142,000	0.77	75,700	231	70

Microstructures of typical bars from these four taps are shown in Figs. 11, 12, 13 and 14. The bar from Tap 1 (Fig. 11) is seen to have partially aggregated quasi-flake graphite; the bar from Tap 2 (Fig. 12) has many spherulitic nodules with only a little quasi-flake graphite; the bar from Tap 3 (Fig. 13) has an almost entirely nodular structure; the bar from Tap 4 (Fig. 14) has only a little quasi-flake graphite. The matrix structure does not differ greatly through this series.

Similar results can be obtained when the metal is treated simultaneously with the cerium addition and the inoculant, as is shown by the following example. A mixture of 70 grams of misch metal and 6 oz of silicon-manganese-zirconium was placed in the bottom of a ladle and 60 lb of metal run onto it. Four bottom-run 1.2-in. bars were cast and their average mechanical properties and the analysis of one bar are:

Chemical Analysis, per cent	
Total Carbon	3.71
Silicon	2.96
Manganese	0.54
Sulphur	0.010
Phosphorus	0.033
Cerium	0.051
Mechanical Properties	
Transverse Rupture Stress, psi	140,900
Deflection, in.	0.60
Tensile Strength, psi	72,100
Brinell Hardness Number	241
Impact Strength, ft-lb	71

The microstructure of one of these bars is shown in Fig. 15, and is seen to have spherulitic nodules with only a trace of quasi-flake graphite in a matrix of pearlite and ferrite.

Application of the double treatment process tends to offset the carbide stabilizing influence of the cerium and hence to reduce the danger of chilling. This effect automatically permits the use of higher percentages of cerium in the solidified casting, a factor which in addition helps to improve the structure of the material.

To illustrate the reduction in chilling tendency with the double treatment technique, the following example may be cited. A wedge test piece having the dimensions of one inch at base, 2 in. from apex to base, and 6 in. long was cast in a dry sand mold. This wedge was fractured transversely and gave a chilled zone approximately  $\frac{3}{4}$ -in. deep. The analysis of this wedge was: total carbon, 3.77; silicon, 2.69; manganese, 0.56; sulphur, 0.011; phosphorus, 0.024; and cerium, 0.061 %.

In this case no double treatment was applied, and the white iron structure of the apex of the wedge is shown in Fig. 16. Only a trace of hypereutectic graphite can be seen. The structure  $\frac{3}{8}$ -in. from the apex is shown in Fig. 17 and, while rather more graphite can be seen, the structure still is predominantly white.

On the other hand, when a similar wedge of the composition total carbon, 3.70; silicon, 2.73; manga-

nese, 0.59; sulphur, 0.008; phosphorus, 0.055; and cerium, 0.061 per cent was cast after the application of the double treatment process, it showed no visible chill when fractured. The structure of the apex of this wedge is shown in Fig. 18, where only a very small amount of carbide can be seen. At a point  $\frac{1}{8}$ -in. from the apex only traces of free cementite existed, as is shown in Fig. 19.

It appears that an optimum amount of cerium is necessary, even with the double treatment technique, to cause the structure to be entirely nodular. When the cerium content is below this value quasi-flake graphite may persist in the structure. This is shown by the results given in Table 4, and the structures shown in Figs. 20, 21, 22 and 23. Table 4 gives the analyses and mechanical properties of two sets of test bars prepared by treating two 50-lb taps of metal poured from the same melt with the following additions: Set 1—40 grams misch metal followed by 5 oz silicon-manganese-zirconium; Set 2—60 grams misch metal followed by 5 oz silicon-manganese-zirconium.

Microstructures of the 1.2-in. bars are shown in Figs. 20 and 21, and those of the 0.6-in. bars in Figs. 22 and 23. The 1.2-in. and 0.6-in. diameter bars of Set 1 clearly have some quasi-flake graphite, whereas this form of graphite is entirely absent and the structure is wholly nodular in the bars from Set 2.

#### Various Examples of Nodular Cast Irons

*Example 1*—With silicon contents below about 2.3 per cent and in the absence of other graphitizing elements, the chilling tendency of cerium-treated nodular irons, even when the double treatment process is applied, is so great that very careful control is necessary in order to avoid chilling in sections of less than  $\frac{3}{4}$ -in.

TABLE 4—ANALYSES AND MECHANICAL PROPERTIES—CERIUM AND SILICON-MANGANESE-ZIRCONIUM ADDITIONS

Composition, per cent						
	T.C.	Si	Mn	S	P	Ce
Set 1	3.90	2.96	0.51	0.006	0.024	0.016
Set 2	3.90	3.10	0.52	0.007	0.023	0.031

Set No.	Bar Size, in.	Transverse Rupture	Deflection, in.	BHN	Impact strength, ft-lb	Tensile Strength, psi
		Stress, psi				
1	1.6	104,500	0.51	179	32	54,800
	1.2	114,100	0.68	189	46	62,700
	0.875	122,900	0.39	213	76	65,100
	0.6	142,000	0.35	239	—	75,900
2	1.6	133,900	0.90	190	89	73,500
	1.2	143,000	1.02	199	>120*	73,900
	0.875	153,900	0.66	231	>120*	77,300
	0.6	164,000	0.43	252	—	98,100

\* Full capacity of the Izod impact machine used is 120 lb.

\* Full capacity of the Izod impact machine used is 120 lb.



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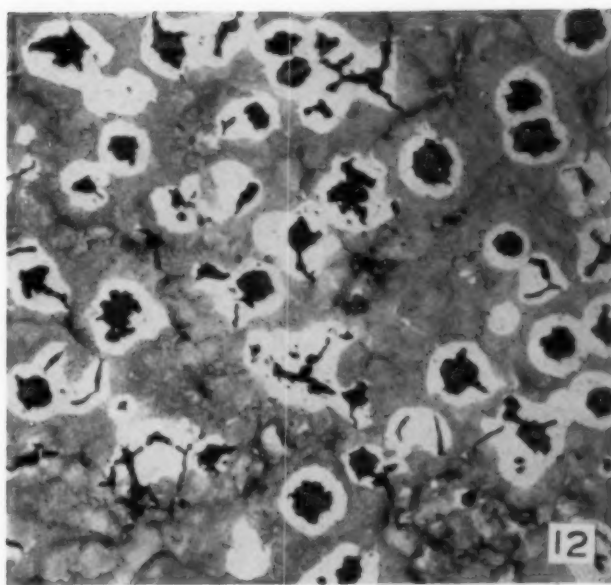
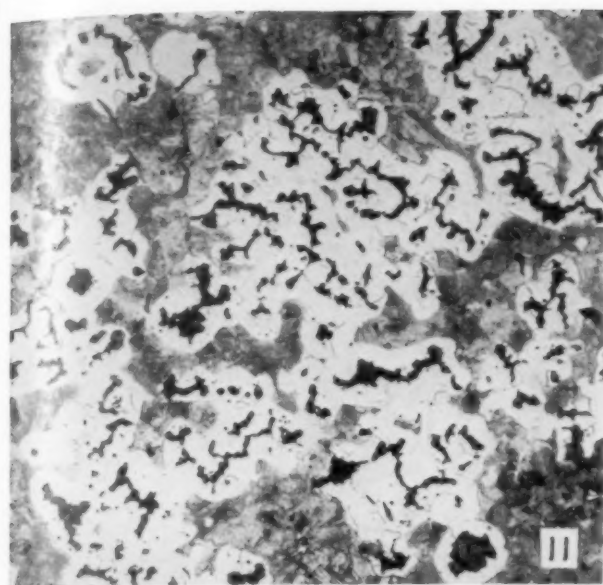


Fig. 11—General structure of 1.2-in. bar with Ce addition (No. 1, Table 3). Etched in picric acid.  $\times 100$ .

Fig. 12—Microstructure of typical 1.2-in. bar with cerium and ferro-silicon addition (No. 2, Table 3). Picric acid etch.  $\times 100$ . Note many spherulitic nodules.

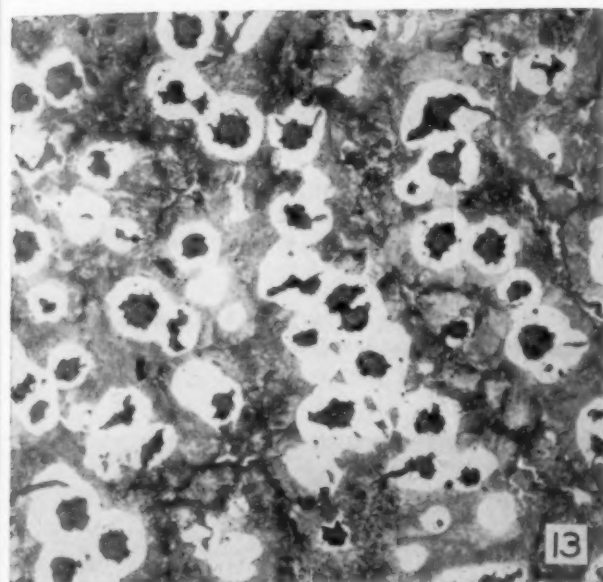


Fig. 13—Structure of 1.2-in. bar with cerium and ferro-silicon addition (No. 3, Table 3) is almost entirely nodular. Etched in picric acid.  $\times 100$ .

Fig. 14—Photomicrograph of 1.2-in. bar with cerium and ferro-silicon addition (No. 4, Table 3) shows small amount of quasi-flake graphite. Picric acid etch.  $\times 100$ .

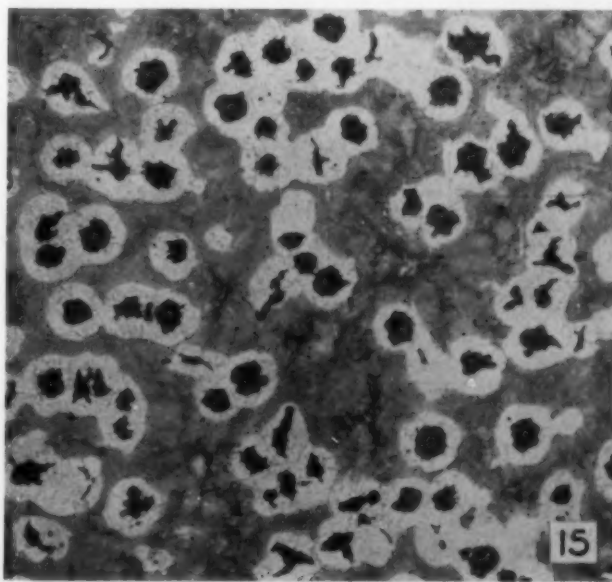
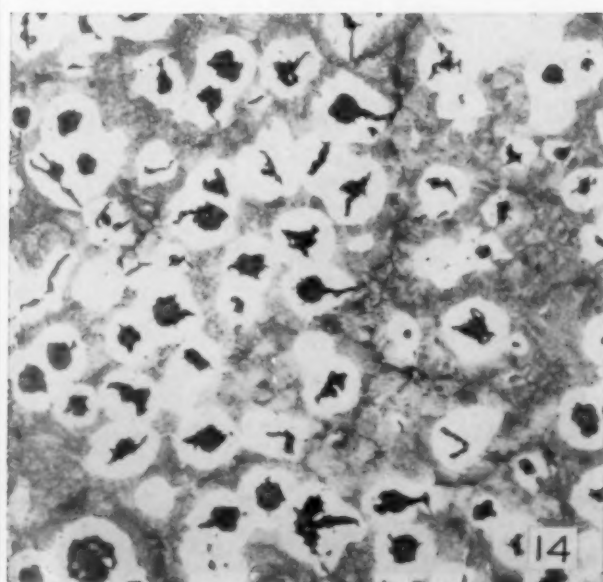


Fig. 15—Metal treated simultaneously with misch metal and silicon-manganese-zirconium shows spherulitic nodules with only a trace of quasi-flake graphite in pearlite and ferrite matrix. Etched in picric acid.  $\times 100$ .

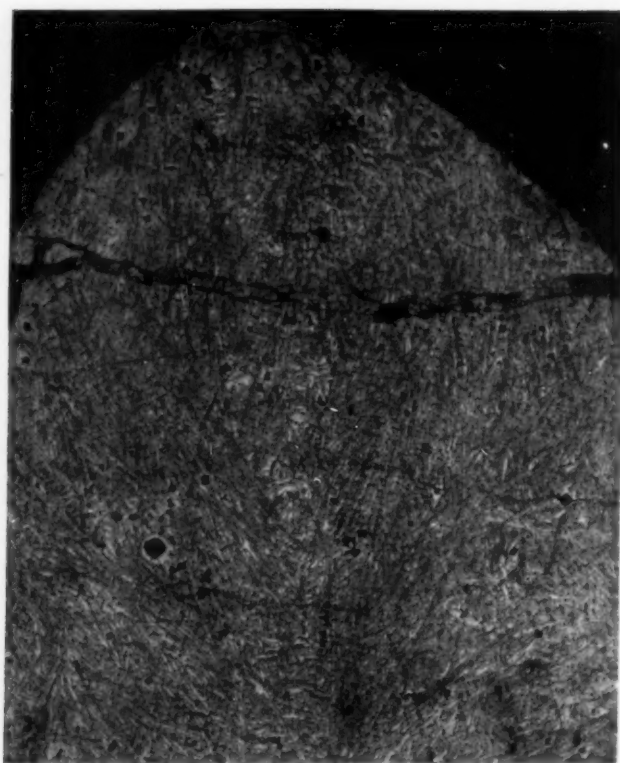


Fig. 16—White iron structure at apex of chill wedge made from metal treated with cerium alone shows only a trace of hypereutectic graphite. Picric acid etch.  $\times 60$ .

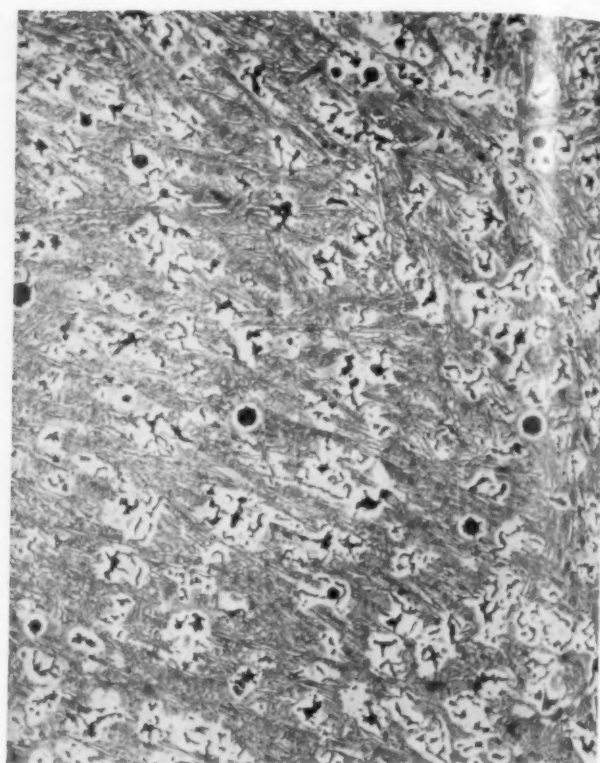


Fig. 17—Structure of chill wedge,  $\frac{3}{8}$ -in. from apex, treated with cerium alone, shows rather more graphite but still is predominantly white. Picric acid etch.  $\times 60$ .

Fig. 18—Apex of chill wedge, cast from double-treated metal, shows only a very small amount of carbide in the structure. Etched in picric acid.  $\times 60$ .

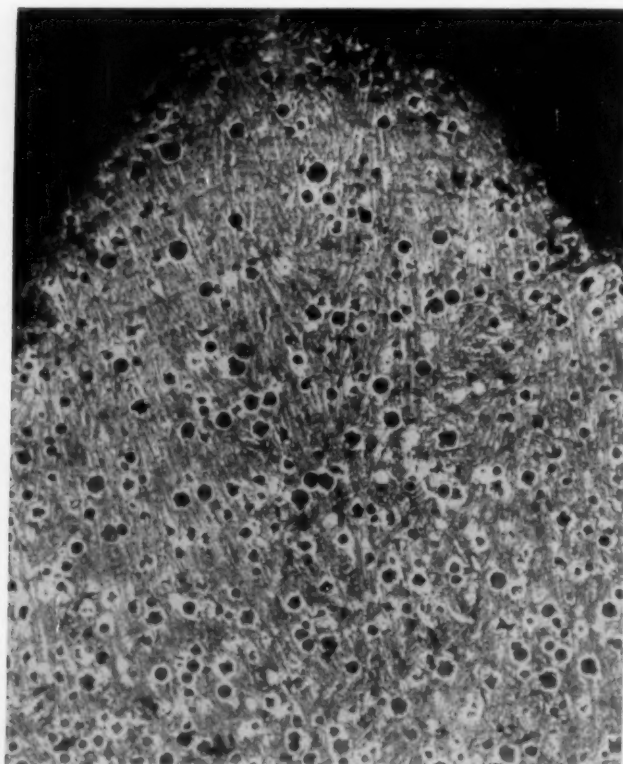
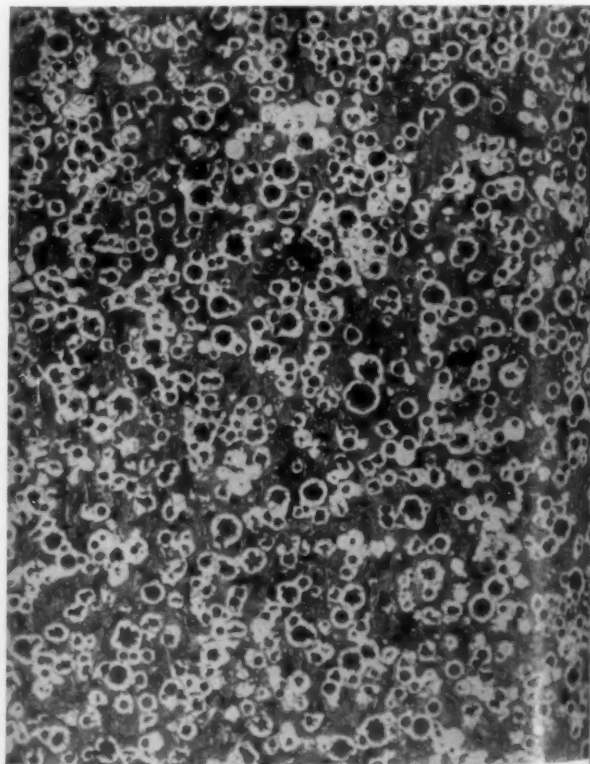


Fig. 19—Chill wedge structure,  $\frac{1}{8}$ -in from apex, cast from double-treated metal, shows presence of only small amount of free cementite. Picric etch.  $\times 60$ .





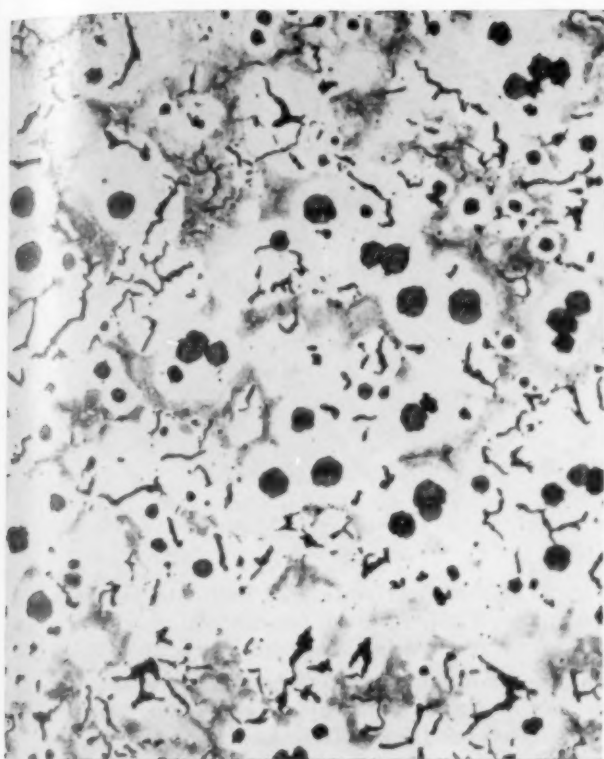


Fig. 20—Structure of 1.2-in. bar (No. 1, Table 4) with 0.016 per cent Ce, produced by double treatment process. Some quasi-flake graphite. Picric acid etch.  $\times 100$ .

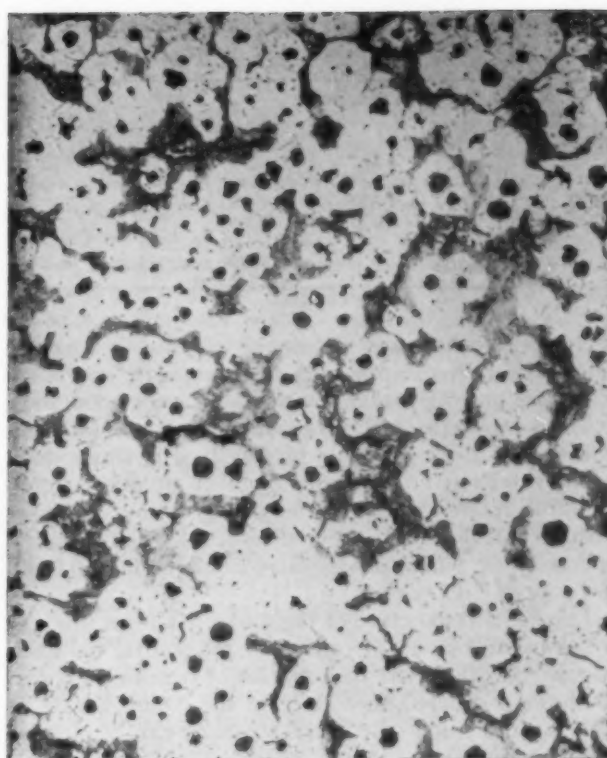


Fig. 21—Wholly nodular structure of 1.2-in. bar (No. 2, Table 4) with 0.031 per cent Ce, produced by double treatment process. Etched in picric acid.  $\times 100$ .

Fig. 22—Microstructure of 0.6-in. bar (No. 1, Table 4) with 0.016 per cent Ce, double treatment process, shows some quasi-flake graphite. Picric acid etch.  $\times 100$ .



Fig. 23—Structure of 0.6-in bar (No. 2, Table 4) with 0.031 per cent Ce, produced by double treatment process, shown to be wholly nodular. Picric acid etch.  $\times 100$ .

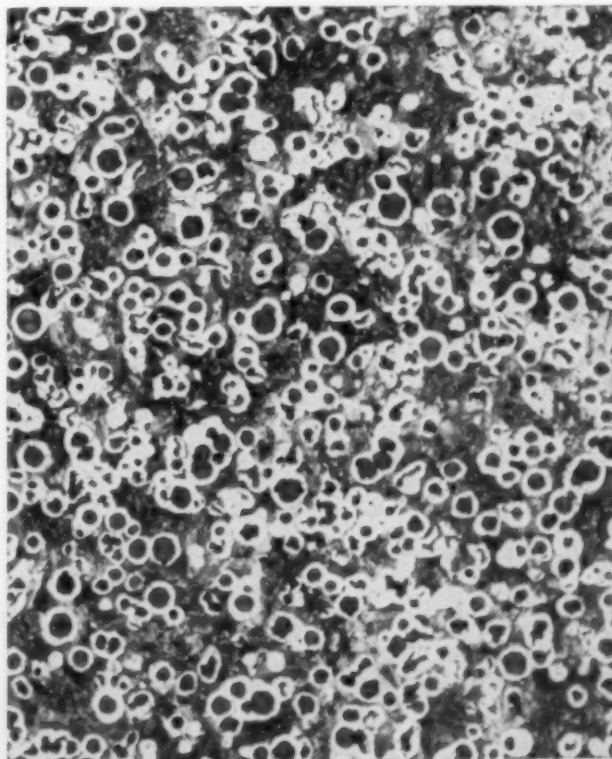




TABLE 5—MECHANICAL PROPERTIES OF DOUBLE TREATED NODULAR IRON

Bar Size, in.	Transverse Rupture Stress, psi	Deflection, in.	Tensile Strength, psi	BHN	Impact Strength, ft-lb
1.6	150,900	0.30	89,000	260	88
1.2	157,000	0.40	79,200	288	92
0.875	181,000	0.30	96,500	302	94
0.6	206,500	0.27	114,000	333	—

However, when any deficiency of silicon is compensated for by the presence of appropriate amounts of copper and/or nickel a material of very good mechanical properties can be obtained, even in relatively thin sections. An example of this type is illustrated by the results given in Table 5, which are taken from a double treated nodular iron of the following composition: total carbon, 3.80; silicon, 2.17; manganese, 0.92; sulphur, 0.015; phosphorus, 0.052; cerium, 0.054; and copper, 2.17 per cent.

The general structure of the 1.6-in bar from this set is shown in Fig. 24 to consist of spherulitic nodules associated with ferrite in a matrix of pearlite. Figure 25 shows a spherulitic nodule in the 0.6-in. bar at a high magnification—the radial structure can be clearly seen. Occasionally in such samples the duplex structure of the nodules is revealed in a quite interesting manner. A typical instance of this is shown in Fig. 26, which shows a duplex nodule in the 1.6-in. bar. The central hypereutectic nucleus can be clearly seen with the remainder of the graphite, still in the spherulitic form, attached to the nucleus at only a few points.

*Example 2*—The double treatment can be applied and useful mechanical properties obtained with silicon contents as high as 6 per cent. With increasing silicon

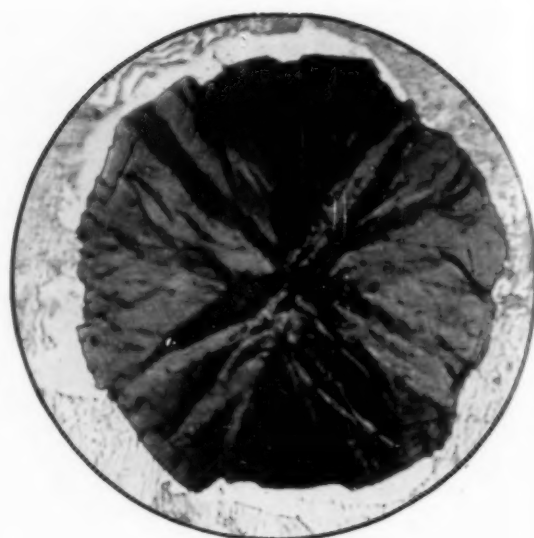
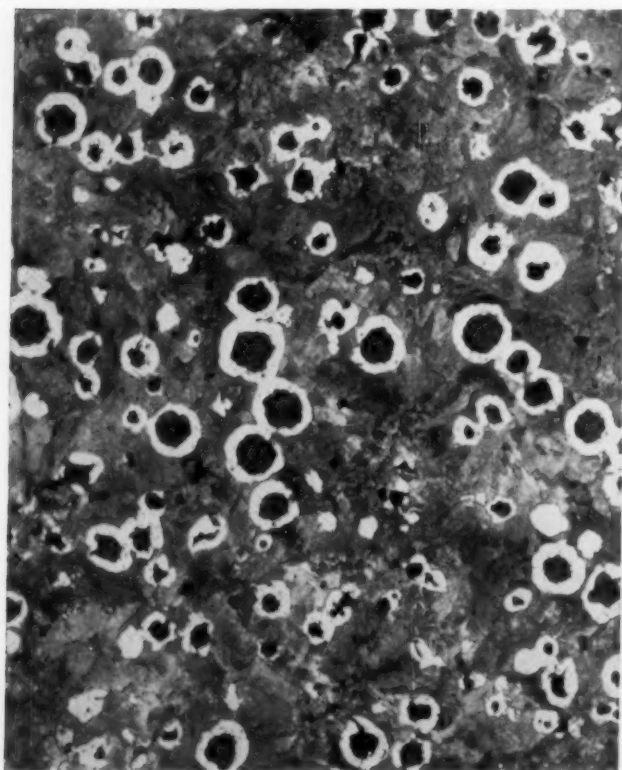


Fig. 25—Spherulitic nodule at high magnification in 0.6-in. bar (Table 5) in copper-containing double-treated iron. Note radial structure. Picric etch  $\times 1250$ .

Fig. 26—Photomicrograph showing duplex structure of nodule in 1.6-in. bar (Table 5) in copper-containing double-treated iron. Etched in picric acid.  $\times 500$ .



up to about 5 per cent the transverse rupture strength, tensile strength and Brinell hardness number do not change appreciably, but the shock-resistance drops progressively. The following figures, obtained on a fairly high silicon iron, illustrate the type of results which can be obtained:

**Chemical Analysis**—Total carbon, 3.14; silicon, 4.13; manganese, 0.85; sulphur, 0.004; phosphorus, 0.046; cerium, 0.051 per cent.

**Mechanical Properties** (1.2-in. diameter bar)—Transverse rupture stress, 141,500 psi; deflection, 0.38 in.; tensile strength, 77,500 psi; Brinell hardness, 234; impact strength, 29 ft-lb.

The microstructure of this sample consisted of spherulitic nodules in a matrix of ferrite with a little pearlite,

Fig. 24—Structure of 1.6-in. bar (Table 5) in copper-containing double-treated iron, shows spherulitic nodules in matrix of pearlite. Etched in picric acid.  $\times 100$ .

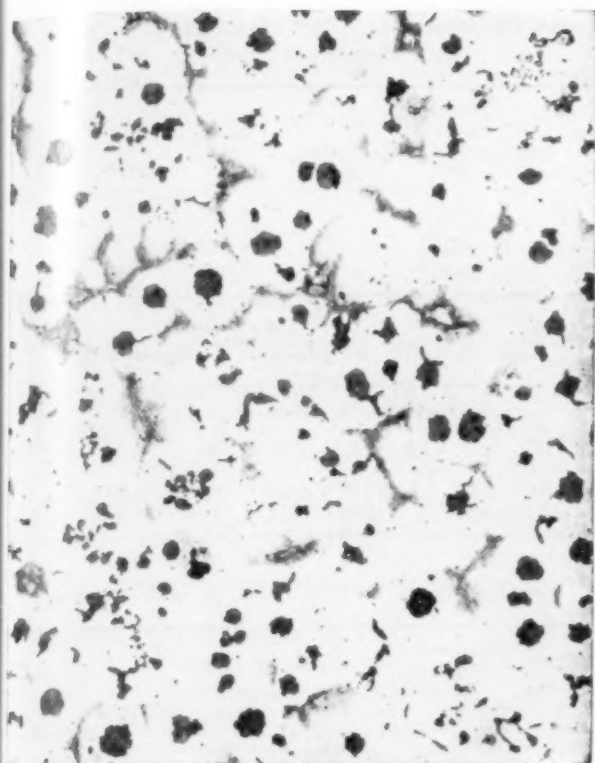


Fig. 27—Microstructure of nodular iron with 4.13 per cent silicon content. Etched in picric acid.  $\times 100$ .

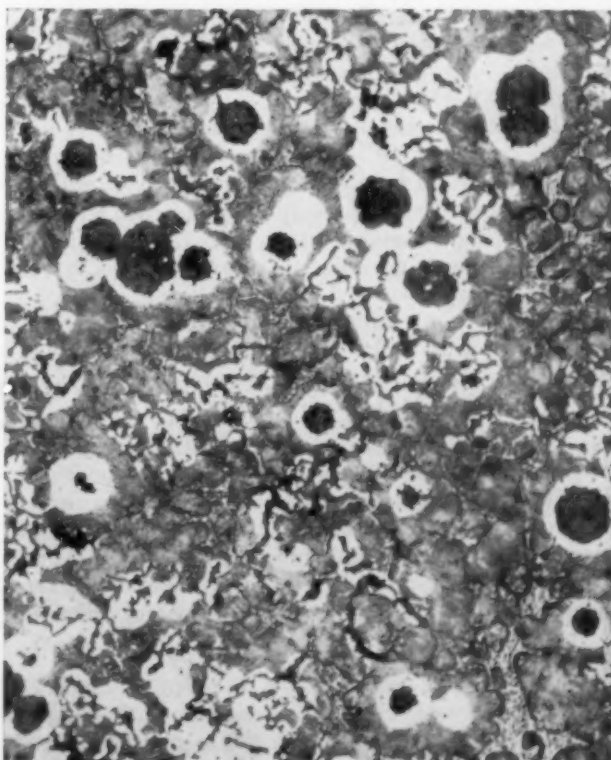
this element. The results shown in Table 6 give the mechanical properties of a fairly high phosphorus iron of the following composition: Total carbon, 3.60; silicon, 2.67; manganese, 0.54; sulphur, 0.007; phosphorus, 0.50; cerium, 0.031 per cent.

This metal was treated with misch metal alone—the double treatment was not applied. The results obtained may therefore be compared with those in Table 1. The microstructure of the 1.2-in. bar from this series had large hypereutectic spherulites and quasi-flake graphite in a matrix of pearlite with phosphide eutectic. This structure is shown in Fig. 28.

### High Manganese Irons

**Example 4**—Provided that the iron will solidify gray before treatment the manganese content of a nodular cast iron may have any value. The general effect of increasing amounts of manganese is to inhibit the formation of ferrite, even with fairly high silicon contents, and ultimately to give entirely pearlitic structures and, if sufficient of the element is present, to give martensitic structures. The results given in Table 7 show the tensile strength and hardness of two relatively high manganese nodular irons produced by the double treatment process. These high manganese irons were cast in the form of 1.2-in diameter bars.

Bar No. 1, containing 2.01 per cent manganese, had a microstructure consisting of uniformly distributed spherulitic nodules in a matrix of about 70 per cent pearlite and 30 per cent ferrite. The graphite structure of bar No. 2 was very similar to that of bar No. 1, but the matrix in this case was almost entirely pearlitic with a small amount of martensite and a little carbide. The structure of bar No. 2 is illustrated in Fig. 29.



as is shown in the photomicrograph of Fig. 27 at  $\times 100$ .

**Example 3**—The use of cerium for the production of nodular cast irons cannot be extended to phosphorus contents in excess of about 0.5 per cent because this element appears to lower the solubility of cerium in the melt. At relatively high temperatures (in excess of 1450 C) the cerium dissolves readily, but on cooling it is deposited from the melt at a temperature of about 1380 C in the form of a cerium-rich alloy.

For instance, 50 lb of a phosphoric pig iron was melted in a crucible furnace and treated with 100 grams of misch metal when at a temperature of 1470 C. The addition appeared to dissolve readily, but as the metal cooled prior to pouring a thick scum developed on the surface of the metal. This was skimmed off and the rest of the metal poured into test bars. The scum and the test bars gave the following figures on subsequent analysis:

	Composition, per cent	
	Scum	Metal Poured
Total Carbon .....	3.33	3.46
Silicon .....	2.37	2.61
Manganese .....	0.72	0.84
Sulphur .....	0.076	0.036
Phosphorus .....	1.20	1.44
Cerium .....	4.80	Not detected

Obviously, there is a considerable segregation of cerium and sulphur in the material skimmed from the surface of the liquid metal.

Although nodular structures can be obtained with up to 0.50 per cent phosphorus, the mechanical properties fall off progressively with increasing amounts of

Fig. 28—Photomicrograph showing structure of 1.2-in. bar (Table 6) in single treatment nodular iron with 0.50 per cent phosphorus. Picric acid etch.  $\times 100$ .

The tensile strength and hardness values of the irons are seen to increase as the structure becomes pearlitic.

*Example 5*—High manganese nickel-containing austenitic irons may be treated with cerium to give nodular structures with a considerable improvement in mechanical properties. This is illustrated by the results given in Table 8, which show the mechanical properties of an untreated and a cerium-treated nickel-manganese austenitic iron. The double treatment was not applied in this case. The untreated material had a structure of coarse flake graphite in a matrix of austenite, and the treated material had a nodular structure in a similar matrix with a little carbide.

#### Treating Nickel-Containing Irons

*Example 6*—When the nickel content of nodular cast iron exceeds about 10 per cent it is not necessary for the iron to be hypereutectic according to the formula: per cent C =  $4.3 - (\frac{1}{3} \text{ per cent Si} + \text{per cent P})$ . Therefore the treatment can be readily applied to ordinary nickel-containing austenitic irons. In Table 9 the chemical analyses and mechanical properties of treated and untreated nickel-containing iron are given for material cast in the form of 0.875-in. diameter test bars. The structure of the untreated bar is shown in Fig. 30 to consist of undercooled graphite in a matrix of austenite, together with a small amount of carbide.

The structure of the treated bar is illustrated in Fig. 31, and is seen to have nodular graphite in a matrix of austenite with rather more carbide than the untreated bar. The treated material was, in this case, produced by treatment of the melt with misch metal alone—the double treatment technique was not employed. It is apparent that the nodular nickel-containing iron has mechanical properties considerably in excess of those of the untreated material.

*Example 7*—With relatively high sulphur contents the yield of cerium depends upon the extent of the desulphurization produced by the addition. In such cases, if desulphurization by the cerium is allowed to take place, the yield of cerium will be low, but it is imperative for the success of the process that this desulphurization should occur.

If desulphurization by the cerium addition is not effected, the apparent yield of cerium will be high, but this cerium, as found by chemical analysis, will be largely combined with the sulphur and hence will not operate to produce nodular structures. As an instance of the desulphurizing influence of cerium, the figures in Table 10 may be quoted for the treatment of cupola-melted nickel-containing iron. These figures clearly indicate the importance of the cerium not combined with the sulphur in determining the graphite structure.

#### High Carbon and Low Sulphur

*Example 8*—The production of nodular cast irons by the process described in this paper presents no special metallurgical difficulties with batch-type melting units if high-carbon and low-sulphur charges are used. The material has been successfully produced in direct- and indirect-arc furnaces, high-frequency induction furnaces, crucible furnaces and oil-fired rotary furnaces

TABLE 6—MECHANICAL PROPERTIES OF 0.50 PER CENT PHOSPHORUS IRON

Bar Size, in.	Transverse Rupture Stress, psi	Deflection, in.	Tensile Strength, psi	BHN	Impact Strength, ft-lb
1.6	82,500	0.20	43,400	241	—
1.2	81,800	0.20	45,300	242	—
0.875	94,100	0.14	43,900	256	11
0.6	81,300	0.09	40,300	285	—

TABLE 7—HIGH MANGANESE NODULAR IRONS

No.	Composition per cent						Tensile Strength, psi	BHN
	T.C.	Si	Mn	S	P	Ce		
1	3.63	2.91	2.01	0.009	0.044	0.067	72,800	256
2	3.51	3.00	2.92	0.017	0.034	0.057	81,100	319

TABLE 8—NICKEL-MANGANESE AUSTENITIC IRON

Bar Size, in.	Transverse Rupture Stress, psi	Deflection, in.	Tensile Strength, psi	BHN	Impact Strength, ft-lb
1.6	32,300	0.60	11,960	86.8	—
1.2	37,000	0.80	13,210	90.7	—
0.875	37,600	0.38	15,000	102	105
0.6	42,100	0.32	16,350	102	—
Untreated: T.C., 3.28; Si, 3.08; Mn, 5.8; S, 0.022; P, 0.072; Ni, 11.87 per cent					
1.6	103,600	1.75	39,000	158	—
1.2	105,300	>2	37,400	156	—
0.875	106,100	1.60	36,700	172	120
0.6	93,200	0.65	35,800	174	—
Treated: T.C., 3.03; Si, 3.18; Mn, 5.9; S, 0.018; P, 0.053; Ni, 12.93; Ce, 0.023 per cent					

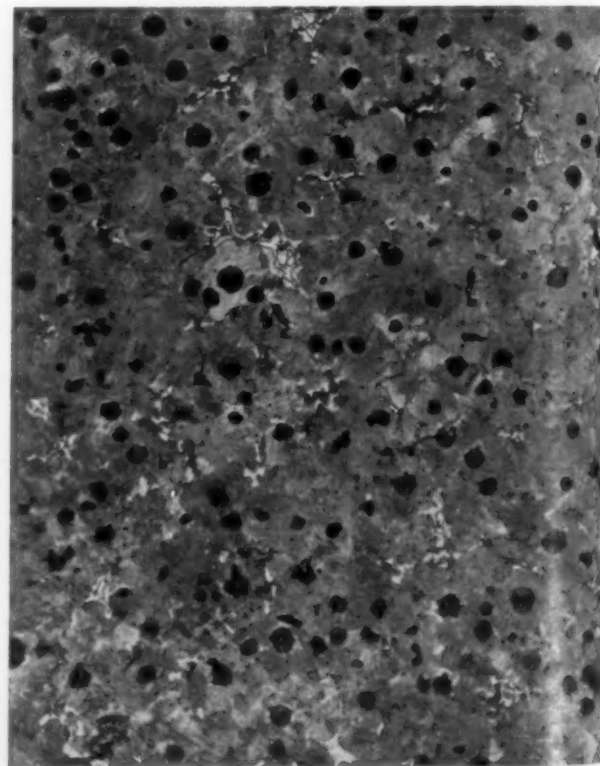


Fig. 29—Microstructure of 1.2-in. bar (No. 2, Table 7) produced in double-treated nodular iron with 2.92 per cent Mn content. Etched in picric acid.  $\times 100$ .



TABLE 9—MECHANICAL PROPERTIES OF NICKEL-CONTAINING IRON

Composition, per cent									
	T.C.	Si	Mn	S	P	Cr	Cu	Ni	Ce
Untreated	2.76	2.68	0.45	0.027	0.024	1.74	6.74	13.24	—
Treated	—	—	—	0.007	—	—	—	—	0.045
	Transverse Rupture Stress, psi		Deflection, in.		Tensile Strength, psi		BHN		
Untreated	65,300		0.47		27,300		146		
Treated	106,500		0.86		46,100		185		

TABLE 10—DESULPHURIZING INFLUENCE OF CERIUM

Addition of Ce as Misch metal, per cent	S, per cent	Ce (by Analysis), per cent	Structure
0.1	0.099	0.066	Flake graphite
0.2	0.051	0.071	Flake graphite
0.3	0.014	0.063	Nodular graphite

TABLE 11—MECHANICAL PROPERTIES OF CUPOLA-MELTED CAST IRON DESULPHURIZED WITH SODIUM CARBONATE

Composition: T.C., 3.69; Si, 2.77; Mn, 0.63; S, 0.005; P, 0.045; Ce, 0.057 per cent

Test Bar Diameter, in.	Transverse Rupture Stress, psi	Deflection, in.	Tensile Strength, psi	BHN
3.0	106,000	0.51	85,000	241
2.1	142,700	2.14	—	—
1.6	152,000	1.30	—	—
1.2	137,500	0.95	80,400	218
0.875	174,300	0.83	76,500	244
0.6	179,500	0.53	—	—

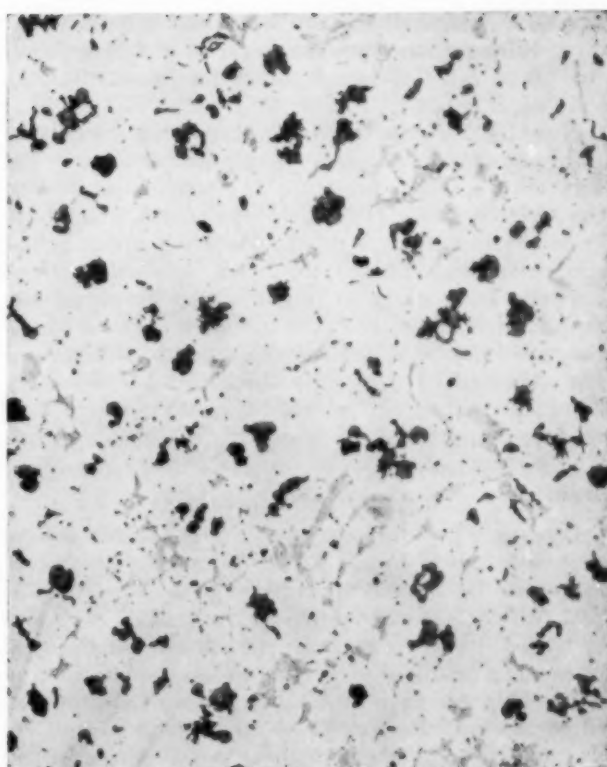


Fig. 31—Microstructure of 0.875-in. bar in nodular nickel-containing iron (Table 9). Picric etch.  $\times 100$ .

by melting low sulphur (of the order of 0.03 per cent sulphur) hematite pig iron and making the appropriate additions of cerium in the ladle.

With the cupola melting furnace the production of molten iron suitable for the application of this process becomes more difficult and, in general, special precautions need to be taken to ensure that the metal is of the correct composition. The production of high-carbon or hypereutectic cast irons from the cupola is not a difficult matter provided the average carbon content of the charge is reasonably high.

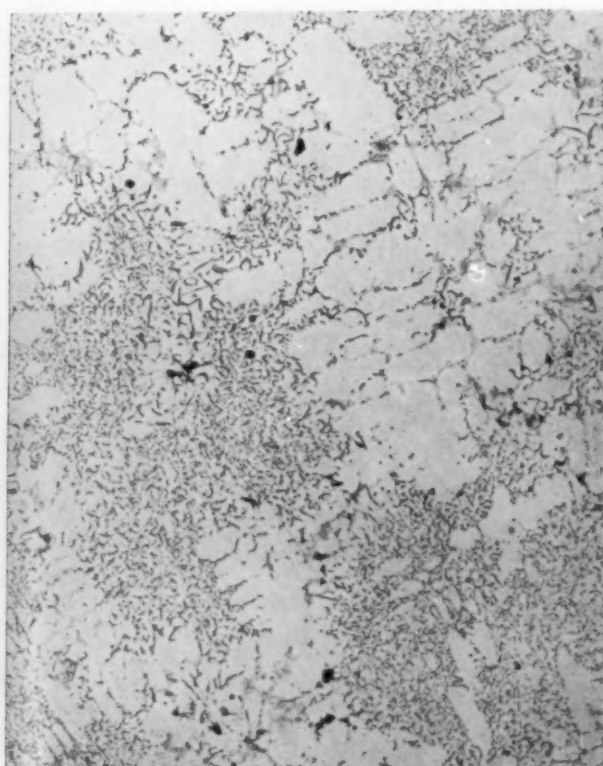
It is, however, contrary to normal practice to aim to produce high-carbon material for the purpose of making a high-duty cast iron. For many years the trend has been in the direction of obtaining lower rather than higher carbon contents.

#### Desulphurizing in the Ladle

Production of satisfactory low-sulphur contents presents much more difficulty. Even when low-sulphur charges are used there generally is a sulphur pick-up from the coke and the sulphur content of the metal tapped will be too high to allow immediate treatment with misch metal, and it therefore becomes necessary to interpose a preliminary desulphurizing treatment of the metal in the ladle.

If the sulphur content of the metal tapped from the cupola is within the range of 0.05-0.08 per cent, it is possible to desulphurize the metal to a satisfactory low-sulphur content by one treatment with sodium carbo-

Fig. 30—Structure of 0.875-in. bar in untreated nickel-containing iron (Table 9). Picric acid etch.  $\times 100$ .



nate, preferably using a ladle lined with a basic refractory. With sulphur contents in excess of 0.08 per cent the "double ladle" technique of desulphurizing may have to be applied.

Nodular cast irons can be produced from the cupola, using the sodium carbonate desulphurizer prior to the addition of the misch metal. The mechanical properties given in Table II illustrate the results which have been obtained from a cupola operating under industrial conditions. These results were obtained from test bars cast from a cupola run in which the metal was tapped directly onto sodium carbonate placed in the bottom of the ladle, the slag thickened with limestone and skimmed off, and then misch metal and silicon-manganese-zirconium added. The microstructure of all bars of this set had spherulitic nodular graphite. The structure of the 1.2-in. bar is shown in Fig. 32.

### Summary and Conclusions

Production of nodular graphite structures by the process described in this paper represents a new development in the field of the metallurgy of cast iron. The past two decades have witnessed the gradual evolution of high-duty gray cast iron, particularly in the field of castings for the engineering and automobile industries. This evolution began with the realization of the importance of the control of carbon and silicon and the relative adjustment of these two elements in cast iron. This phase was accompanied by the development and application of the hot-mold process and the utilization of high-steel mixtures in cupola practice.

Next, the inoculation technique was introduced and developed to a process of indispensable commercial utility. The aim of this technique has been to give random flake graphite structures in irons of relatively low carbon and silicon contents and to inhibit and control the chilling tendencies of such irons. All of these developments have been accompanied by advances in the use of special alloying elements, such as chromium, nickel, copper, molybdenum, etc., to strengthen the metallic matrix of cast iron.

These advances have culminated in the production of the nickel-molybdenum acicular cast irons which, in commercial production, represent a combination of most of the earlier developments—acicular irons are essentially irons with controlled low carbon contents; they are usually inoculated and have a combination of alloying elements which, after an appropriate heat-treatment, confer the maximum strengthening effect on the metallic matrix.

At this point it was reasonable to enquire in what direction further developments in cast iron metallurgy were likely to lead. For practical foundry reasons there appeared to be a lower limit to the carbon content of cast iron used under general production conditions. The inoculation technique had been developed to a state of precise control and no immediate hope could be held out that new elements or combinations of elements could easily raise the strength of the metallic matrix beyond that already achieved (this latter is, however, a field which must be continuously and painstakingly explored).

At this formidable barrier it is to be expected that attention should be focused on the possibility of producing nodular graphite structures in cast irons without the necessity for a heat-treatment application.

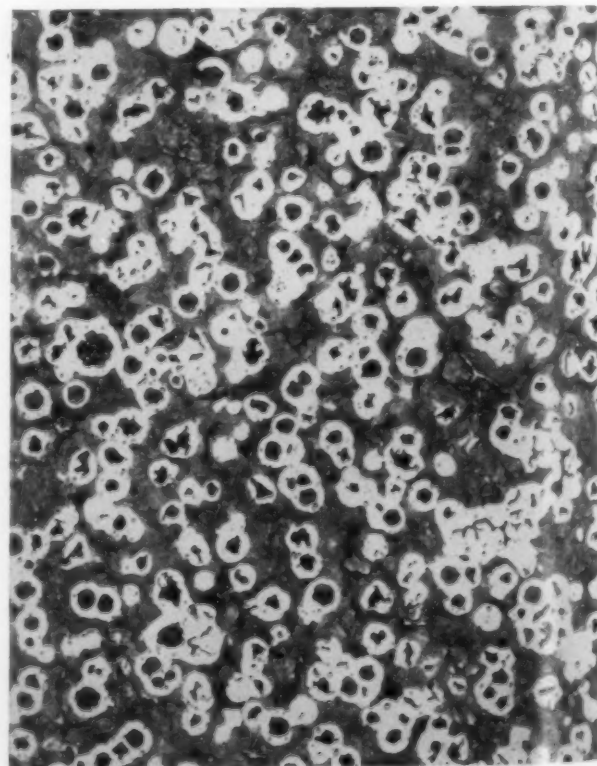
In the opinion of the author it can now be claimed that this development, once seemingly remote, has now been demonstrated to be a possibility and, furthermore, that a process has been developed which is capable of application on an industrial scale. Nodular cast irons produced by the process described in this paper can be readily recognized and differentiated from ordinary gray or malleable cast irons.

Nodular graphite structures are usually associated with malleable cast iron, but it should not be assumed that the properties of the nodular cast irons described in this paper in any way resemble those of malleable irons. The nodular irons represent a material which lies midway between a normal high-duty gray cast iron, on the one hand, and malleable cast iron on the other. Nodular cast irons still are relatively non-ductile and brittle, although they may have a shock-resistance many times better than that of normal gray cast irons.

In this paper it has been shown that the graphite nodules in nodular cast irons are of the spherulitic variety. In studying features of this type erroneous conclusions may be reached unless the metallographic aspect of the work is conducted with great care. The present author and his colleagues have discussed the formation of graphite spherulites more fully on previous occasions, references to which have been given.

The spherulitic form of graphite nodule is probably more commonly known in Europe than in the United States because the manufacture of whiteheart malleable iron is mainly conducted in Europe. Nevertheless, the author would like to draw attention to this interesting form of graphite crystallization which forms the

Fig. 32—Microstructure of 1.2-in. bar (Table II) in cupola-melted double-treated iron, shows spherulitic nodular graphite. Etched in picric acid.  $\times 60$ .



nucleus upon which her present process for the production of nodular graphite in cast irons has been built.

Also in this paper a new graphite arrangement has been recognized and described as "quasi-flake graphite." This form of graphite does not weaken the iron to the same extent as ordinary flake graphite, and may itself be an important structural constituent in some nodular irons.

No attempt has been made to deal with the heat-treatment or with the special mechanical properties of nodular cast irons. These will form the subject of future papers. Nevertheless, it can be stated that nodular irons can be subjected to all the usual forms of heat-treatment, without influencing the graphite structure, for the purpose of obtaining special effects. Nodular cast irons may also be alloyed with elements such as chromium, nickel, copper, molybdenum, etc., to obtain special strength effects in the metallic matrix.

Nodular cast irons may be run and cast in a variety of sections, but problems relating to porosity and fluidity are no greater than those inherent in the production of normal gray cast iron.

It is impossible at this stage to predict the future of nodular cast iron, but the possibility exists of the material playing an important part in the field of engineering and related applications where high strengths are required, and redesign to take advantage of these strengths may be an important factor in the economy of nodular cast irons. The field has now been revealed and other and more convenient processes may be developed to achieve the same end, but it is fairly safe to assume that any alternative process will still demand a high degree of technical skill and control in the producing iron foundry.

#### Acknowledgments

The author wishes to thank the Council and Director, J. G. Pearce, of the British Cast Iron Research Association, for permission to publish this paper. In addition he is indebted to his co-workers, J. W. Grant and W. J. Williams, for their enthusiastic co-operation in much of the work which is described, to W. Westwood and A. Mayer of the association's chemical laboratory, and to D. Marles, metallographic laboratory.

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#### DISCUSSION

*Chairman:* J. T. MacKENZIE, American Cast Iron Pipe Co., Birmingham, Ala.

*Co-Chairman:* W. W. LEVI, Lynchburg Foundry Co., Radford, Va.

A. I. KRYNITSKY (*Written Discussion*):<sup>1</sup> The author should be congratulated for his splendid work. The theoretical and practical importance of this most interesting and valuable paper cannot be overemphasized.

I only want to remark that a rather pronounced globular graphite structure was observed during one of the investigations carried out at the Experimental Foundry of the National Bureau of Standards and published in the *Journal of Research of the National Bureau of Standards*, vol. 28, p. 73, January 1942, and in the *TRANSACTIONS, A.F.A.*, vol. 50, p. 451 (1942).

This structure observed in a 1.2 in. transverse test bar, cast in green sand of a nickel-chromium-molybdenum iron (T.C., 2.88; Graph. C., 2.16; Si, 2.70; Mn, 0.99; P, 0.12; S, 0.04; Ni, 1.24; Cr, 0.21; Mo, 1.17 per cent heated to 1700 C and poured at 1350 C is shown in Fig. 33.

From microscopic examination of this bar it seemed that most of the free carbon was present in the form of globules (Fig. 33A). A row of graphite globules is shown in Fig. 33B. Some of them were associated with graphite flakes, as may be seen in Fig. 33C. It appeared as if the graphite nodule originated from a nucleus situated at the end of the flake which extends into the nodule.

The modulus of rupture of this iron was 106,900 psi; total deflection, 0.34 in.; and relative modulus of elasticity as determined in transverse tests was  $18.60 \times 10^6$  psi.

This nodular graphite was best developed in a bar cast from metal which had been heated to 1700 C but was also present, although to a lesser extent, in bars cast from metal which had not been heated above 1400 C. Consequently, in these experiments, a high maximum heating temperature may have increased the amount of nodular graphite formed but was not essential. It is unfortunate that the pressure of other activities prevented further investigation of this phenomena at that time.

MR. MORROGH: With regard to the comments by Col. Krynitsky, I have been aware of the structures to which he has referred. There obviously is some unknown factor playing a part in this case. Further information which he has presented tempts me to suggest an explanation for those structures since I have produced structures similar to that in Fig. 33.

When irons are cooled under certain critical conditions, we have, just after solidification, eutectic graphite and saturated austenite. Normally that austenite gets rid of its excess carbon by deposition upon that already existing graphite flakes. However, if the cooling rate is at a certain critical value a graphite Widmanstätten structure can be obtained, and if on subsequent cooling the cooling is delayed slightly, spheroidization and aggregation of this Widmanstätten structure can result to give the type of structure depicted by Col. Krynitsky in Fig. 33.

I should like to ask Col. Krynitsky whether he has been able to reproduce the structures which he obtained.

COL. KRYNITSKY: Unfortunately due to the pressure of other activities, we did not continue this investigation, so I do not know if we could reproduce this structure. It is possible that there may have been present in the iron some minor element or there was some other unknown factor which brought about the formation of the nodular graphite.

CHAIRMAN MacKENZIE: What kind of crucible did you melt it in?

<sup>1</sup> National Bureau of Standards, Washington, D. C.



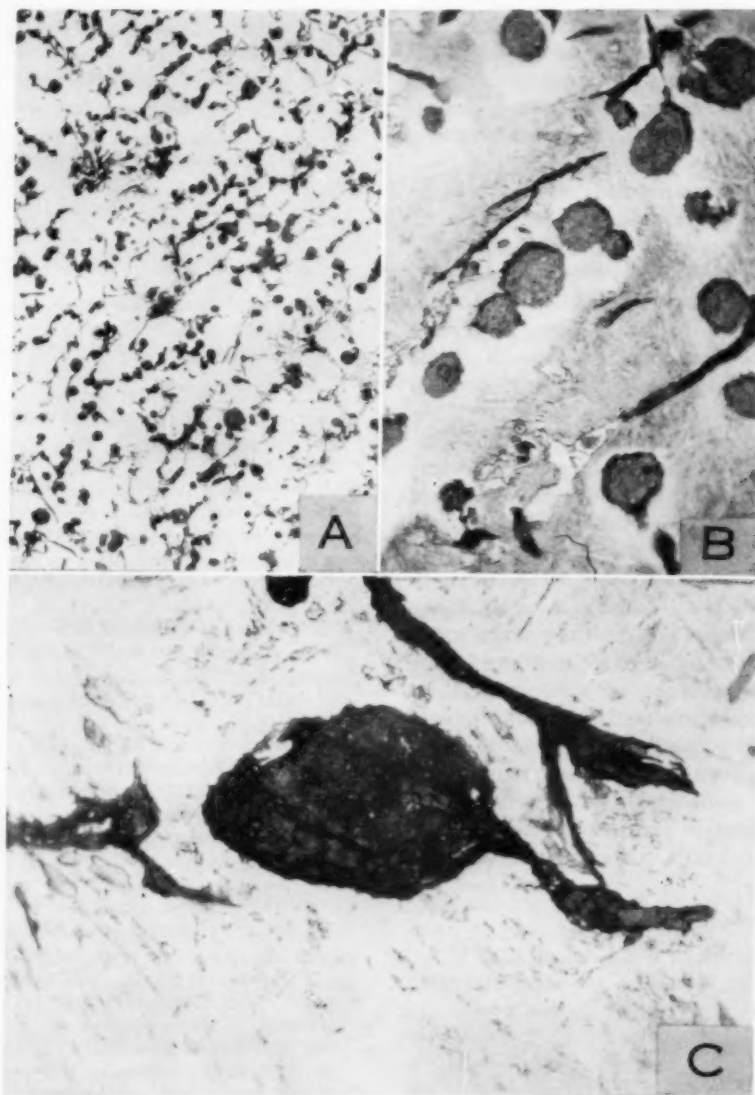


Fig. 33—A 1.2 in. transverse test bar cast in green sand mold of a nickel-chromium-molybdenum iron (T.C., 2.88; Graph. C., 2.16; Si, 2.70; Mn, 0.99; P, 0.12; S, 0.04; Ni, 1.24; Cr, 0.21; Mo, 1.17%) heated to 1700 C and poured at 1350 C. Modulus

of rupture 106,900 psi; total deflection 0.34 in.; relative modulus of elasticity  $18.6 \times 10^6$  psi. A, unetched, X100; B, etched, X500; C, etched, X1500. Etching reagent was 1 per cent solution of nitric acid in ethyl alcohol.

COL. KRYNITSKY: Melting was done in a magnesia crucible, high frequency induction furnace. With reference to the structure which is shown in Fig. 33C, I questioned myself which was formed first, flake or globule? If the graphite flake was first formed, then the globule was formed later. It is interesting to note what the author has to say about this.

MR. MORROGH: I have no further comments to make on Col. Krynitsky's further remarks beyond saying that there is obviously some unknown factor operative there of which we have no knowledge at the moment.

T. H. WICKENDEN:<sup>2</sup> Mr. Morrogh and his colleagues are to be warmly congratulated on the progress they have made toward the metallurgical goal of producing a nodular graphite cast iron during the casting and solidification period. A great many people have been working toward the same goal. Among them is the staff of the International Nickel Co. Research Laboratory, Bayonne, N. J. It is a subject on which they have been working for many years. We had not intended to make a formal announcement of our development until at a later date. But in

view of Mr. Morrogh's paper telling about the progress in Great Britain, I feel that we owe it to the foundry industry to let it be known that we have been successful in developing a process which will produce a spheroidal graphite structure in cast iron in the as-cast condition. It is being tested out on a commercial scale and plans are being formed for its commercial development. Further work is going on which we expect to cover in a paper at a later date.

The Inco process, as I shall call it, is based on the introduction into the iron of a small but effective amount of magnesium or a magnesium-containing addition such as nickel-magnesium alloy. A small but effective addition produces a partial conversion of graphite to the spheroidal form and the remaining graphite takes on a compact form. A larger addition insures that all of the graphite is converted to the spheroidal form, and as the amount of spheroidal graphite is increased in the iron the strength will also increase from the base level to a value several times that of the untreated product.

The impact properties are improved over this range. A low-phosphorus iron with the graphite in the fully spheroidal condi-

<sup>2</sup> International Nickel Co., Inc., New York

tion will show some elongation in the as-cast condition. This can be greatly improved by a controlled cooling rate or subcritical heat treatment.

Our process we find is effective on both hypo- and hyper-eutectic irons. The properties, of course, vary with the base material but a base iron which as-cast shows 25,000 to 30,000 psi tensile strength will probably come out when treated with a tensile strength of 65,000 psi or higher depending on the graphite content, and with low phosphorus, it may show in the as-cast condition between 1 and 2 per cent elongation. With controlled cooling or subsequent tempering or subcritical annealing the elongation will be around 5 or 6 per cent.

CHAIRMAN MACKENZIE: What carbon equivalent do you use?

MR. WICKENDEN: Either hypereutectic or hypoeutectic.

MR. MORROGH: I wish to offer my congratulations to the staff of the International Nickel Co. for accomplishing this development. We must patiently await the subsequent disclosure of the details of the process. It does serve, however, to amplify the remarks which I have made toward the end of my presentation, that nodular cast irons are going to come into the iron foundry industry and they are going to come by several different processes. The future holds that in store for us.

R. A. FLINN: Mr. Morrogh and his colleagues of the British Cast Iron Research Institute have made a very important fundamental contribution to our knowledge of gray iron metallurgy. The point that delights us most is that the goal was obtained not by a completely empirical set of experiments but by a basic approach. Not one, but several steps and complex reasoning were involved, such as a study of the analogous nickel-carbon and cobalt-carbon system, the recognition of the importance of under-cooling in producing nodular graphite, and the review of a number of rare ladle additions.

In the same spirit we wish to devote our questions and discussions to the fundamental rather than the commercial aspect of the problem. Several questions arise from a brief study of the data. We wonder what this new knowledge on the production of nodular graphite can contribute toward an understanding of the mechanism of inoculation, producing flake graphite. This has been such a fundamental study that I think the author may be able to amplify that for us as well as the question of nodular graphite.

Another fundamental question relates to carbide stabilizers and graphitizers. Does this new work shed any light on why an element is a carbide stabilizer? It is easy enough to answer for elements like vanadium and chromium, but what can we say about elements like selenium, sulphur, lead, and even these small amounts of cerium? I think if we can solve this problem of carbide stabilization with the help of the graphitization data we will have accomplished even more.

I should like to discuss the impact test. We have done a fair amount of thinking about impact testing as a method of rating materials. It seems to us that a relatively slow-speed unnotched impact test is really just a kind of fast transverse test. It is a good control test and an interesting test, but we think that you can get the information in a more exact form from a complete stress-strain curve. We would be interested to have the author comment on this opinion.

What is the effect of holding time in the ladle upon the sub-graphitization? Do type of scrap and type of melting furnace have an effect upon the formation of nodular graphite?

MR. MORROGH: With regard to the mechanism of inoculation, I would say that, broadly speaking, this process adds very little to our knowledge of the mechanism, and as it were, makes the field we have to cover more difficult and the problems we have to explain more complex. However, the influence of the inoculant in the double treatment process does perhaps throw some rather indirect evidence upon the mechanism of that type of inoculation. An American investigator, Eash, suggested that ferro-silicon or similar inoculants function by inhomogeneous solution in the metal giving silicon concentrations at certain points, causing the material to be hypereutectic with the deposition of small particles of graphite which remain and function to nucleate the solidification process. [See "Effects of Ladle Inoculation on An Austenitic Cast Iron", by J. T. Eash, TRANSACTIONS, A.F.A., vol. 50, p. 815 (1942)] The evidence that we have obtained from this double treatment process is that

it is highly probable that such an inoculation mechanism does take place.

Dr. Flinn raised some extremely pertinent comments on the subject of carbide stabilization. That is still largely a virgin field. It can be said, however, that when we have metals which form stable carbides themselves and which are capable of entering the iron carbide lattice vicariously with the iron atoms, in general, they are carbide stabilizers, as is the case with manganese, chromium, and vanadium. However, when the elements form unstable carbides, as, for instance, nickel, and they enter the iron carbide lattice vicariously with the iron atoms, then we have a graphitizing influence.

However, the problem of carbide stabilization becomes much more complex when it is recalled that under-cooling in cast irons, even when the result is a gray iron, involves the solidification of a white cast iron which graphitizes subsequent to solidification. Normally we associate with iron carbide less stability or less persistence at high temperatures and increasing persistence at low temperatures. When we have under-cooling we have, seemingly, an instance of higher persistence at higher temperatures, that is, shortly after solidification, than at lower temperatures. Perhaps this indicates that carbon and iron are intimately associated in the melt as some form of iron carbide.

With regard to impact testing, we are well aware of the limitations of the single-blow unnotched impact test. I would not agree that the transverse test gives all the information which the impact test gives as I have, on several occasions, had materials of ostensibly the same properties as judged by transverse testing and tensile testing but which under the impact test were quite different.

With regard to holding time in the ladle and its effect upon this process, with normal foundry practice and with holding times in the ladle up to about 20 min there is little influence. In fact, the longer the holding time in many ways the better is the final result because it permits the segregation desulphurization by the cerium to be more effective.

Dealing with the question of the melting furnace, broadly speaking, provided the compositions are the same, melting furnace type has little effect. We find that composition for composition, the same amount of cerium is required, for instance, in the cupola furnace as in the high-frequency induction furnace. With batch type melting furnaces it is, however, possible to control the sulphur to a low figure, given satisfactory starting charge materials. With the cupola, of course, there is always an inevitable sulphur pickup and more cerium may be required on that account unless the desulphurization is carried out beforehand.

With regard to scrap, I presume that Dr. Flinn is inquiring about the remelting of cerium-treated nodular cast irons. On remelting it is usual for the effective cerium to be lost by oxidation. On remelting an iron containing 0.05 per cent cerium in a crucible furnace, the final cerium content may be as low as 0.005 per cent or lower. Scrap containing cerium cannot therefore be regarded as a potential source of cerium for the process, although its inclusion in the charge may assist in achieving a low final sulphur content.

J. E. REHDER: I should like to add my contribution to the congratulations to Mr. Morrogh and the British Cast Iron Research Association on this excellent work. I think it is not too much to say that a landmark in the history of cast iron metallurgy has been delineated in this technical paper. Mr. Morrogh's metallographic work in particular I think is some of the finest I have seen anywhere.

I have one question to ask Mr. Morrogh. These irons are hypereutectic irons, and I am curious as to what has been found with respect to fluidity of these hypereutectic cerium-treated irons.

MR. MORROGH: With regard to the question raised by Mr. Rehder on fluidity, it is rather difficult to give a precise answer to that, because there is no precise method of measuring fluidity. From actual measurements and in the course of actual foundry experiments, we find that at comparable pouring temperatures the fluidity of this iron is better than that of a low-phosphorus iron with, say 3 per cent carbon and 2

<sup>1</sup> American Brake Shoe Co., Mahwah, N. J.

<sup>2</sup> Bureau of Mines, Ottawa, Ontario, Canada

per cent silicon. Comparing the treated hypereutectic iron with the untreated hypereutectic iron, at the same temperature the fluidity is slightly impaired by the cerium treatment. The precise reason for that is not understood, but cerium-treated irons do tend to be covered with a thin oxide film, such as one sees occasionally in high-titanium iron and in aluminum-containing irons. That may be responsible for this slight reduction in fluidity.

W. W. AUSTIN, JR.:<sup>\*</sup> I should like to reiterate the compliments to Mr. Morrogh and his associates for this outstanding achievement in the field of cast iron metallurgy. I cannot resist this opportunity, however, to inquire as to whether or not any data have been made available on the relative machinability of nodular cast iron in comparison with that of the normal flake graphite structure. Also, we would be interested, I am sure, in hearing some cost data on the utilization and application of this treatment in the foundry.

MR. MORROGH: With regard to machinability, I am unable to quote any precise figures. The machinability of this material will vary between very wide limits, depending upon the composition and the specific type of matrix structure which is obtained. It should be noted that properties equivalent to those

<sup>\*</sup> Southern Research Institute, Birmingham, Ala.

of pearlitic gray cast irons can be obtained with this material in the partially ferritic condition. Under those conditions, an improvement in machinability is obtained. We have in England substituted nodular cast irons of this type for ordinary pearlitic gray cast irons for a variety of castings, and they have been put through the machine shops without any comments at all.

With regard to the cost of this process, again it is rather difficult to give any concrete figures, since it is closely tied up with the type of furnace that is being used, the type of charge materials that are being used, the sulphur content of the metal which has to be treated and also the cost of the cerium. I am not sure what is the cheapest cerium available in this country but it is a relatively expensive process, and the direction in which we are conducting large scale experiments in England is in that of reducing the cost by reducing the amount of cerium being used.

Some preliminary idea of costs can be ascertained in respect to cerium when you know the cost of Misch metal in this country by remembering that yields of cerium from Misch metal will vary between about 50 per cent and 30 per cent, depending upon the way in which the addition has been made and the sulphur content of the metal being treated. In general, treated irons will contain between 0.03 and 0.06 per cent of the element.

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# DELAYED QUENCH FOR STEEL CASTINGS\*

By

S. L. Gertsman \*\*

## ABSTRACT

A "delayed quench" process is described for the heat treatment of castings. The process involves austenitizing at temperatures usually employed; a timed cooling from the austenitizing temperature to a temperature sufficiently above the  $A_{r3}$  to ensure that in plant operation no casting would be below the  $A_{r3}$  temperature; and a quench in water, followed by a draw to produce the required mechanical properties. Cracking susceptibility is lessened by quenching from the lower temperature. A considerable saving of furnace time is effected over the normalize-reheat-quench-and-draw procedure usually employed. Results presented demonstrate that the saving is made at no sacrifice of mechanical properties. Since accurate control is required, the process is recommended only where volume production is involved.

HEAT TREATMENT in steel foundries has evolved gradually. It was found that ductility and machinability of the castings were improved by full annealing, i.e., heating them up to a temperature of 1650-1750 F and allowing them to cool slowly in the furnace. It was then observed that by allowing the castings to cool in air, rather than in the furnace, the yield point and tensile strength were increased, particularly in alloy steels. In high carbon and alloy steel castings of certain designs, however, objectionable cooling stresses were induced by this treatment unless the castings were drawn immediately. Cast steel of all compositions, however, had better ductility, higher yield point, higher tensile strength, and much higher impact and fatigue properties than were possessed by castings given a full annealing treatment.

At first, liquid quenching was avoided because of the possibility of cracking. Such treatment, however, is rapidly gaining favor for both carbon and alloy steels since it produces maximum mechanical properties. The quenching is usually followed by a draw to reduce cooling stresses, which may cause brittleness, and also to produce desirable combinations of mechanical properties. The quenching medium may be either oil or water; the latter is more widely used.

As experience accumulates on the quenching of steel

castings, precautionary measures are being worked out to minimize the danger of cracking. It has been noted that the water being used should be kept preferably at 125-150 F, and that the castings should be removed from the quench tank while still warm and be placed immediately into the draw furnace. Attention is also being focused on the quench temperature. In this connection, the fact that the upper critical temperature on cooling ( $A_{r3}$ ) is lower than the upper critical temperature on heating, makes possible the quenching of a casting from a fairly low temperature while still retaining the fully hardened structure.

The temperature gradients caused by the quenching are then less severe than if the steel were quenched from more elevated temperatures. Distortion, internal stresses, and the tendency to crack are thus reduced. The quenching from this lower temperature is most conveniently achieved by allowing the steel to cool in air from the austenitizing temperature to the temperature of quenching. As a consequence, the process is referred to as a "delayed quench."

As higher carbon (0.45 to 0.52 per cent) and low alloy steels have higher hardenability than mild steel, they also are more likely to crack on quenching. This danger can be minimized, especially on intricate castings, by the use of the delayed quench.<sup>1</sup> The method, however, requires the setting up of accurate controls.

This paper describes the technique that was used to set up the delayed quench in a Canadian steel foundry so that it could be operated by unskilled personnel. The problem was undertaken in the metallurgical laboratories of the Bureau of Mines, Ottawa, in January, 1944. The foundry was producing ball sockets for the four-wheel-drive army trucks. Some of the preliminary work is also included on establishing the analysis (low-molybdenum steel) and the delayed quench treatment that were used for the American M-24 Cadillac tank track shoe, Fig. 3, which was produced in volume at a later date.

Avoidance of cracking was not the only object, however, as the heat treating facilities of the foundry were inadequate to take the flow of castings and constituted a production bottleneck. It was thought that in setting up the delayed quench the entire heat treatment opera-

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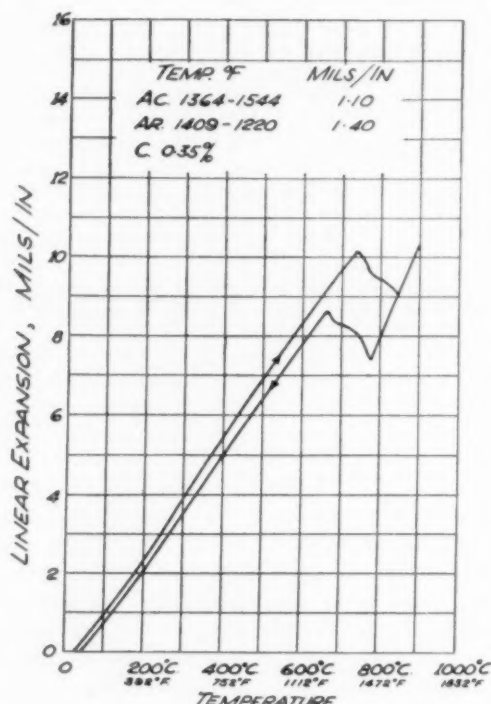


Fig. 1—Dilatometric curve of 0.35 per cent carbon steel. The steel was cooled at a rate of 275 F per hr in obtaining the curve.

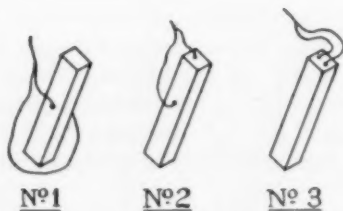
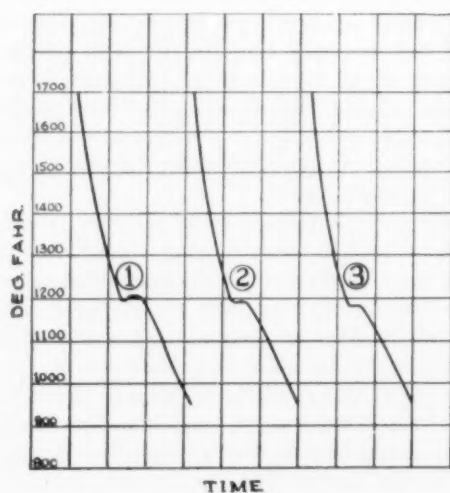


Fig. 2—Variations in AR due to small changes in cooling rates.

tion might be accelerated. This was successfully carried out with no reduction in the mechanical properties of the steel.

An outline of the process that was adopted, and of the mechanical properties obtained, may be of aid to foundrymen in indicating a method by which small and medium-sized steel castings required in volume can be heat treated more economically.

Medium-carbon steel was used and the following heat treatment was given to the ball sockets, all the temperatures quoted representing furnace pyrometer settings:

1.  $3\frac{1}{2}$  hours at 1750 F; air cooled.
2. Reheated to 1575 F for  $2\frac{1}{2}$  hours; water quenched.
3. Drawn at 1225 F for  $3\frac{1}{2}$  hours.

The 1750 F treatment<sup>1</sup> is given to change the "as cast" structure (coarsely ferritic area in a matrix of coarse pearlite) Fig. 9 to a refined structure (ferrite and pearlite finely distributed), and is also supposed to relieve the tendency toward segregation at points where opposing dendrites meet (such segregation forms a line of weakness which would not be eliminated by subsequent heat treatment). Reheating to 1575 F refines the grain if previous coarsening has occurred, and it confers the beneficial effects of a lower quenching temperature, e.g., reduces the tendency to crack or warp. The draw, of course, brings about the desired combination of mechanical properties.

After careful testing of a series of heats of steel the heat treatment was modified as follows to include a "delayed quench" procedure:

1. Soaking for  $3\frac{1}{2}$  hours at the austenitizing temperature (approximately 1750 F).
2. Cooling in air to the lowest practicable point (1350 F, approximately) above the critical temperature ( $Ar_3$ ) of the steel prior to quenching.
3. Quenching in water.

4. Tempering, or drawing, at 1225 F for  $3\frac{1}{2}$  hours.

It is noticed that the reheat-treatment at 1575 F has been eliminated. This would appear to eliminate a grain refining treatment. For most castings, however, grain size is not a critical factor. Indeed, grain coarsening is advantageous, both from the standpoint of increased hardenability and as an aid to machinability, provided its effects on certain mechanical properties are not serious.

### Critical Points, AC and AR

It seems advisable at this stage to discuss certain fundamentals of steel transformation involved in the

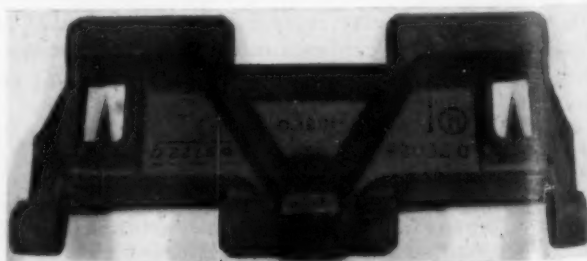


Fig. 3—M-24 tank track shoe.

use of the delayed quench. Iron exists in two forms, namely, the high-temperature, face-centered gamma form which can hold carbon in solution, and the body-centered, low-temperature form which has relatively low solubility for carbon. Because equilibrium conditions are not achieved in commercial operation, the change from one structure to another occurs at a higher temperature on heating than on cooling.

These changes can be followed by dilatometer readings. Heating within a single phase range produces general expansion, but a change from the body-centered form produces contraction. Temperature arrests also occur in the transformation. In using the delayed-quench procedure, advantage is taken of these phenomena by using them to determine the correct temperature at which the steel is completely face centered and austenitic, and then quenching from reasonably close to this temperature.

Figure 1 shows a dilatometer curve for an 0.35 carbon steel. The differences between the AC and AR critical points is shown. The  $Ar_3$  temperature is that which corresponds to the reversion of the original downward slope of the curve and the  $Ar_1$  temperature is the one at which the downward trend is resumed again. The reverse, of course, holds for the  $Ac_1$  and  $Ac_3$  points. In obtaining the curve the cooling rate was kept constant at 275 F per hour.

There is a tendency for metallurgists to consider the critical points of steel as static values independent of the corresponding rates of temperature change. Consequently, the literature often gives the critical point values unaccompanied by cooling rates. The lack of sufficient emphasis of the relation between these two factors is best illustrated by the scarcity of published data on the subject.

Figure 2 illustrates to some extent the effect of cooling rates on the critical temperature. Chromel-alumel thermocouple wires were attached by the method herein described to 6 x 1 x 1/2-in. bars of 0.45 per cent carbon steel, in the position shown in Fig. 2. These bars were heated up to 1700 F for one hour in a neutral atmosphere. The ends of the thermocouple were attached to an L & N "Speedomax" machine and curves were obtained, using the slow speed recorder, after removing the bars from the furnace and allowing them to cool in still air. This gives a qualitative picture of the depressing effect of accelerated cooling rates on the change point. It is observed that with the thermocouple wires attached to the middle (slowest cooling section) of the bar as shown by No. 1 in Fig. 2, the  $Ar_3$  is approximately at 1200 F whereas when they are attached to the end (fastest cooling section) of the bar as in No. 3 arrangement, the  $Ar_3$  is reduced to approximately 1180 F.

It has been shown that, for SAE 4340 steel<sup>2</sup>, varying the cooling rates in degrees per hour from 10 F per hour to 400 F per hour changed the  $Ar_3$  point from 1340 F to 1200 F.

### Chemical Analysis

Table 1 gives the chemical analyses of the various heats of steel castings and keel block test coupons made, for the experimental work, at the foundry in

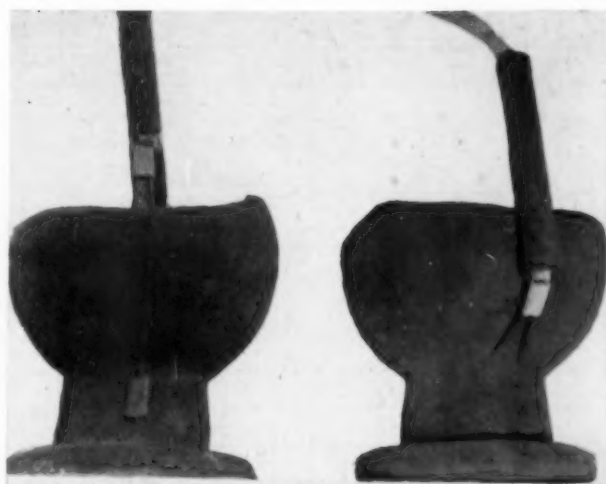


Fig. 4—Ball socket casting. Left, thermocouple attached to the top (fastest cooling section); right, thermocouple attached to the neck (slowest cooling section).

TABLE 1—ANALYSES OF EXPERIMENTAL HEATS OF STEEL

Heat Numbers	Composition, per cent				
	Carbon	Man-ganese	Silicon	Sulphur	Phos-phorus
8866	0.48	0.81	0.31	0.024	0.021
8868	0.43	0.93	0.33	0.025	0.021
8870	0.43	0.85	0.30	0.025	0.020
8873	0.37	0.93	0.25	0.028	0.018
8874	0.47	0.88	0.37	0.032	0.019
8876	0.53	0.88	0.21	0.020	0.029
8877	0.42	0.86	0.25	0.028	0.032
8878	0.42	0.82	0.28	0.031	0.022
8879	0.39	0.85	0.32	0.032	0.028

TABLE 2—ANALYSES OF HEATS MADE AT BUREAU OF MINES LABORATORY

Heat Numbers	Composition, per cent					
	Carbon	Man-ganese	Silicon	Sulphur	Phos-phorus	Molybdenum
224	0.45	0.83	0.31	0.027	0.016	
228	0.29	1.47	0.21	0.049	0.016	
229	0.26	0.93	0.25	0.022	0.017	0.38
233	0.46	0.78	0.21	0.055	0.033	
234	0.52	0.84	0.28	0.055	0.030	
235	0.40	0.85	0.28	0.069	0.041	
236	0.25	0.98	0.28	0.055	0.033	0.52
255	0.40	0.88	0.19	0.035	0.010	
256	0.38	0.80	0.14	0.033	0.010	
258	0.29	0.87	0.35	0.030	0.010	0.31
260	0.54	0.82	0.53	0.031	0.010	

basic-lined, electric arc furnaces. Table 2 lists the analyses of the supplementary heats which were cast in the Bureau of Mines laboratories, using a basic-lined induction furnace. Besides carbon steel, several molybdenum steel heats were cast for the track shoe work.

### Establishing Time Cycle in Air Prior to Quenching

In routine operation of most heat treating furnaces it can be expected that the temperature of the work will usually be somewhat below the temperature of the



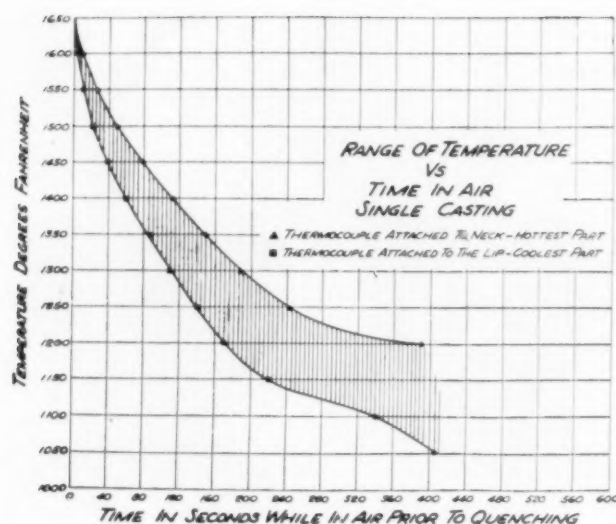


Fig. 5—Cooling curves for ball socket casting.

furnace pyrometer. It was obvious that the temperatures of the work prior to removal from the furnace would have to be established if the proper time cycle in air was to be determined. In order to carry this out, a chromel-alumel thermocouple was attached to the castings in the following manner; an electric arc welder having a graphite electrode was used to melt a small puddle of steel in the casting. One wire of the thermocouple was then rapidly immersed in this puddle about  $\frac{1}{8}$  in. below the surface and allowed to freeze into the steel. This was repeated with the other wire. To perform this operation satisfactorily it was necessary to grind a point on the electrode and to apply it to a freshly ground part of the casting. To protect the thermocouple from breakage during the process of handling, a pipe was attached to the casting and the thermocouple wires were run through it. Asbestos was then packed around the thermocouple to prevent shorting it and to keep it rigid in the pipe. The whole thermocouple was protected by insulating beads. Figure 4 illustrates the thermocouple assembly with the wires attached to the fastest cooling (left casting at the top) part of the ball socket and to the slowest cooling section (right casting at the center or neck). These two points were determined after thermocouples were attached at different parts of the casting and cooling curves were obtained.

A casting with a thermocouple attached to the top was placed in the rotary oil-fired furnace at the plant (Fig. 4, left). The point of attachment was the thinnest part of the casting ( $\frac{1}{2}$ -in. wall thickness) and consequently the most sensitive to furnace temperature variations. The leads of the thermocouple were attached to a portable potentiometer. The furnace was operating under production conditions. After the casting reached 1700 F during the cycle, it cooled down to 1650 F just prior to removal from the furnace. Therefore, although the furnace pyrometer was reading 1750 F, the starting temperature of 1650 F had to be used to standardize the time cycle in air prior to quenching.

TABLE 3—SOME DATA ON DELAY QUENCHING

Time in Air Prior to Quenching, sec.	Quenching Temperature Range of Casting, °F		Temperature of Water, °F.	Brinell Hardness	
	Fastest Cooling Section (top)	Slowest Cooling Section (neck)		Heat 8873	Heat 8876
26	1487	1550	71	534	627
49	1425	1495	73	534	627
85	1350	1437	75	534	627
115	1295	1398	74	534	653
140	1245	1358	76	555	627
151	1230	1347	77.5	534	653
188	1175	1300	79.5	555	653
230	1138	1257	72	415	555
243	1135	1250	74	415	375
332	1104	1207	76	217	269

Two castings, having thermocouples attached at the top and center (neck) as shown in Fig. 4, were placed in a neutral atmosphere and held for  $1\frac{1}{2}$  hours at 1650 F. When the socket was removed from the furnace, the time of cooling was recorded at 50 F intervals, using a potentiometer. The curves plotted for the slowest and fastest cooling parts of the casting are shown in Fig. 5.

#### Delayed Quench Hardness Experiments

Ball sockets taken from two production heats (Heats 8873 and 8876) representing the outside ranges of carbon content, were heated up in a furnace at 1650 F for  $3\frac{1}{2}$  hours. These castings were taken out individually and allowed to cool in air for various lengths of time and then quenched in a tank ( $6\frac{1}{2} \times 3 \times 3$  ft.) with a water flow of 360 gallons per minute. After the quench, the Brinell hardness of the castings were taken at the center (neck) of the casting, the section which cooled slowest. Table 3 lists the hardness, the times in air prior to quenching, the temperature ranges of the castings at the moment of quenching (taken from Fig. 5), and the water temperatures during the quench operations.

#### Standard Heat Treatment vs. Delayed Quench

The main consideration in setting up the time cycle in the plant is the determination of the lowest temperature from which quenching can be carried out satisfactorily. The cooling rate, as was shown previously, greatly influences the  $A_{r3}$  temperature. Figure 1 illustrates that the  $A_{r3}$  point of an 0.35 per cent carbon steel when cooled at a rate of 275 F per hour is 1409 F. The  $A_{r3}$  point of the ball socket (0.37-0.53 per cent carbon) was between 1257 F to 1300 F while cooling in air from the temperature normally prevailing in the plant furnace. (Cooling rate was 90 F per minute measured at 1300 F.) Table 3 shows that the as-quenched Brinell hardness falls off only after the socket is held in air for 230 seconds before quenching, after which time a temperature range of 1138 to 1257 F is reached.

Since the hardness readings are taken close to the slowest cooling section, or neck, of the socket, it can be assumed that the casting has to be quenched from a point somewhere above 1257 F. In order to get a uniformly martensitic structure throughout the casting,

the fastest cooling part, or top, should be above this temperature.

To care for the variations that normally occur in production, it was felt that the temperature at the top should be well above the  $A_{r3}$ ; consequently, 1350 F was chosen. Figure 5 shows that the top of the socket is at 1350 F when the casting has been held in air for 85 seconds after being taken from the furnace at 1650 F. The temperature at the neck of the casting after the same length of time in air is 1435 F. The temperature range within the casting for an 85-second delay is therefore 1350-1435 F.

### Mechanical Properties

In order to compare the various mechanical properties of ball sockets given the delayed quench with those of the sockets which were normalized and reheated prior to quenching, the following series of experiments were carried out:

Cooling curves were obtained for bars from which tensile test pieces were to be prepared after delayed quenching. The dimensions of these bars were:

Type A =  $6 \times 1\frac{1}{8} \times 1\frac{1}{16}$  in. (representing the thicker sections of the casting).

Type B =  $6 \times 1 \times \frac{1}{2}$  in. (thinner sections).

Thermocouples were attached to the bars in the

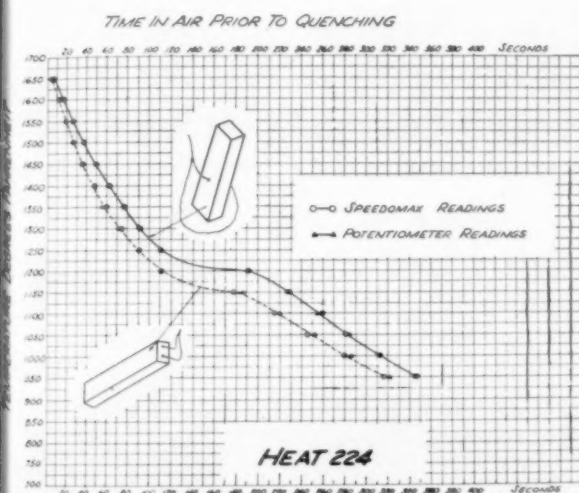


Fig. 6—Cooling curves for Type A bar,  $6 \times 1\frac{1}{8} \times 1\frac{1}{16}$  in.

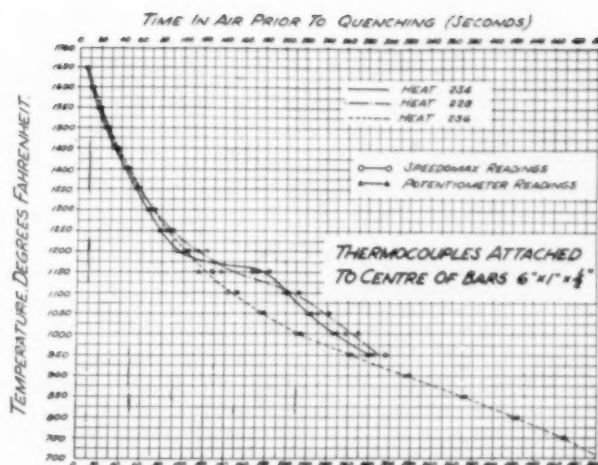


Fig. 7—Cooling curves for Type B bars. These curves show minor variations in cooling rate down to 1350 F for steels of different analysis.

manner previously described. A portable potentiometer was used to determine the temperatures. The times were taken at 50 F intervals, using a stop watch. This method was checked by an L & N Speedomax machine, using the high-speed recorder.

Figure 6 illustrates the curves obtained for bars of Type A above, cut from Heat 224, which has 0.45 per cent carbon. These bars were heated to 1700 F, since the castings reached that temperature during the heat-treat cycle at the plant. The broken line is the curve for the fastest cooling part of the bar, whereas the solid line represents the cooling curve through the center section, which has the slowest cooling rate.

Tensile test bars and Izod impact bars were machined in such a manner that the break occurred close to the center section of the bar, as it was felt that the cooling rate at this point would be most representative. Therefore, in all the delayed-quench experiments for both types of bars, A and B, the cooling curve representing the cooling rate through the center was used in order to determine the time the bar had to be held in air to reach any required temperature prior to quenching.

Figure 7 gives the cooling curves for bars of Type B, taken from three heats of steel having different

TABLE 4—MECHANICAL PROPERTIES, TYPE A BARS \*

Heat No.	Heat Treatment	Max. Stress, psi	Yield Stress, psi	Elongation in 1 in., per cent	Reduction of Area, per cent	Brinell Hardness
224	Delayed Quench	105,000	89,000	24	54.8	197
224	Normalize-Reheat-Quench	103,000	88,400	25	57.4	183
233	Delayed Quench	106,500	89,700	19	41.8	202
233	Normalize-Reheat-Quench	107,500	88,000	21	45.9	205
234	Delayed Quench	110,300	96,800	20	31.7	223
234	Normalize-Reheat-Quench	113,800	97,400	20	35.0	225
235	Delayed Quench	105,600	88,800	17	31.1	217
235	Normalize-Reheat-Quench	105,500	87,200	21	44.5	210
229	Delayed Quench	103,800	86,000	23	57.0	197
229	Normalize-Reheat-Quench	105,600	93,600	23	56.0	201
228	Delayed Quench	92,400	76,000	15	30.0	174
228	Normalize-Reheat-Quench	93,200	80,600	16	33.0	174
236	Delayed Quench	116,300	101,000	20	42.0	222
236	Normalize-Reheat-Quench	117,500	102,000	20	42.0	225

\* All the delayed-quench bars were held in air for 75 sec., after which time they reached 1350 F, prior to quenching.

TABLE 5—MECHANICAL PROPERTIES, TYPE B BARS

Heat No.	Heat Treatment	Time in Air Prior to Quench, sec.	Temperature of Specimen When Quenched °F.	Maximum Stress, psi.	Yield Stress, psi.	Elongation in 1 in., per cent	Reduction of Area, per cent	Brinell Hardness
8870	Delayed Quench	57	1350	99,400	86,800	32	54.2	195
8870	Normalize-Reheat-Quench			97,800	80,000	29	38.7	195
8873	Delayed Quench	38	1435	90,600	71,600	26	50.0	195
8873	Normalize-Reheat-Quench			92,200	74,200	24	33.5	195
8877	Delayed Quench	57	1350	97,100	85,500	30	51.6	195
8877	Normalize-Reheat-Quench			97,400	81,300	30	51.6	195
8879	Delayed Quench	38	1435	95,200	80,000	28	50.0	195
8879	Normalize-Reheat-Quench			95,800	82,200	32	55.7	200
8876	Delayed Quench <sup>1</sup>	57	1350	106,100	90,600	27	45.1	222
8876	Normalize-Reheat-Quench <sup>2</sup>			113,200	102,300	17	29.0	244
8878	Delayed Quench <sup>1</sup>	57	1350	112,800	98,400	21	37.0	236
8878	Normalize-Reheat-Quench <sup>1</sup>			103,200	80,600	23	41.9	222
8874	Delayed Quench <sup>1</sup>	57	1350	102,000	84,500	27	51.6	222
8874	Normalize-Reheat-Quench <sup>1</sup>			105,700	89,700	24	49.0	222

<sup>1</sup> Drawn at 1200 F, 1½ hr.<sup>2</sup> Specimen had a longitudinal quench crack.

analyses. Heat 236 has 0.25 per cent carbon and 0.52 molybdenum; Heat 228 has 0.29 carbon and a high manganese content (1.47 per cent); and Heat 234 is 0.52 carbon steel. As would be expected in cooling to 1350 F no greater than a four-second difference exists between the three heats.

Two pieces were cut from each keel block test coupon, taken from different heats of steel. One piece received the delayed-quench-and-draw treatment, whereas the other was used for the standard normalize-reheat-quench-and-draw treatment. All specimens were held at 1700 F for 1½ hours. The normalized specimens were reheated to 1550 F for one hour and then quenched. A 1½-hour draw at 1225 F was used for most specimens, and some were drawn at 1200 F to give somewhat higher hardness values. Table 4 gives the results obtained for both methods of heat treatment on bars of Type A (6 x 1-1/8 x 11/16 in.). The delayed-quench bars were all held in air for 75 seconds prior to quenching. Figure 6, *solid line curve*, shows that this length of time is required to cool the bars to 1350 F. This temperature corresponds to the lowest temperature of the ball socket when it is held in air for 85 seconds prior to quenching.

Table 5 compares the results obtained with bars of type B (6 x 1 x 1/2 in.). Two specimens have been cooled in air to 1435 F prior to quenching. This is the highest temperature at any point in the ball socket when it is given an 85-second delay prior to quenching. Figure 7 shows that for this size of bar a delay of 57 seconds in air (the average reading given by the three curves) is required to bring the temperature at the center of the bar down to 1350 F, and 38 seconds for a temperature of 1435 F.

The bars were all quenched by hand in water. The tensile bars were machined to 0.282-in. diameter tapering 0.005 in. to the center of the specimen, which caused the bars to break at that point in tension.

It can be seen, from the test results, that no significant difference in mechanical properties exists for bars given either type of heat treatment.

### Impact Tests

Tables 6 and 7, for Type A and Type B bars respectively, give the results of Izod impact tests taken on bars given both heat treatments. The delay quench

TABLE 6—TYPE A BARS—6 x 1 1/8 x 11/16 IN.

Heat No.	Heat Treatment Type <sup>1</sup>	Impact, ft.-lb.	Brinell Hardness
224	Delay Quench	48:52	192
224	Normalize-Reheat	50:52	192
233	Delay Quench	30:32	202
233	Normalize-Reheat	32:32	207
234	Delay Quench	24:24	223
234	Normalize-Reheat	24:25	217
235	Delay Quench	26:27	207
235	Normalize-Reheat	26:25	207
236	Delay Quench	26:26	223
236	Normalize-Reheat	26:25	229

<sup>1</sup> Delay Quench: Bars heated to 1700 F for 1½ hr, then taken out, cooled in still air for 75 sec. to 1350 F, water quenched, and drawn at 1225 F for 1½ hr.

Normalize-Reheat: Bars were heated to 1700 F for 1½ hr, then taken out, cooled in air to room temperature, reheated to 1575 F, held one hour at temperature, water quenched, and drawn at 1225 F, for 1½ hr.

TABLE 7—TYPE B BARS—6 x 1 x 1/2 IN.

Heat No.	Heat Treatment Type <sup>1</sup>	Impact, ft.-lb.	Brinell Hardness
8868 <sup>2</sup>	Delay Quench	48:50	220
8868 <sup>2</sup>	Normalize-Reheat	47:47	220
255	Delay Quench	49:43	197
255	Normalize-Reheat	43:43	192
256	Delay Quench	41:37	192
256	Normalize-Reheat	39:41	192
258	Delay Quench	47:45	212
258	Normalize-Reheat	45:45	219
260	Delay Quench	37:36	229
260	Normalize-Reheat	35:36	223

<sup>1</sup> Delay Quench: Bars heated to 1700 F for 1½ hr, then taken out, cooled in still air 57 sec. to 1350 F, water quenched, and drawn at 1225 F for 1½ hr.

Normalize-Reheat: Bars heated to 1700 F for 1½ hr, air cooled to room temperature, reheated to 1575 F, held at temperature for 1 hr, water quenched, and drawn at 1225 F for 1½ hr.

<sup>2</sup> A 1200 F draw for 1½ hr was used for these bars instead of 1225 F.

bars were all held in still air to 1350 F prior to quenching. The 10-mm. square type of specimen was machined from the heat-treated bars. A V-notch was placed one-half inch on each side of the center of the bar. The results indicate that the impact resistance is the same for bars given either type of heat treatment.

### Grain Size

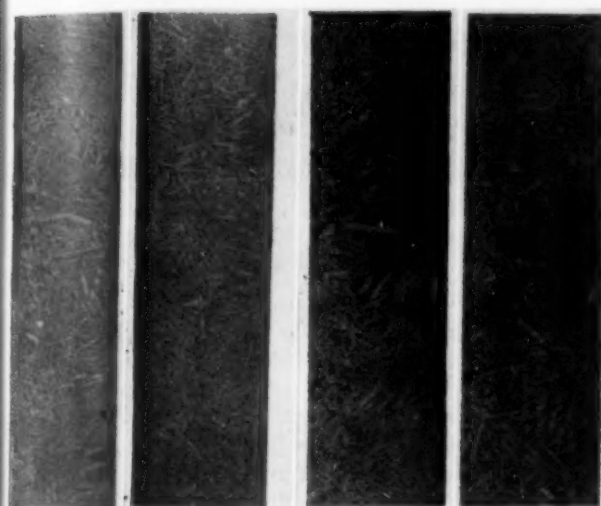
Grain size determinations were made in order to see whether any coarsening occurred. After quenching, all samples were drawn at 600 F for one hour. Vilella's etch was used on the polished microspecimens to indi-



cate the austenite grain boundaries. Table 8 shows that for the heats tested, reheating effects no appreciable refinement.

TABLE 8—GRAIN SIZES OF EXPERIMENTAL HEATS

Heat No.	Type of Bar	Heat Treatment	Grain Size
8866	B	Delay Quench	5-6
8866	B	Normalize-Reheat-Quench	6-7
234	A	Delay Quench	5-6
234	A	Normalize-Reheat-Quench	5-6
229	A	Delay Quench	5-6
229	A	Normalize-Reheat-Quench	5-6



Normalized Delayed Quench  
Heat 224  
Normalized Delayed Quench  
Heat 229

Fig. 8—Macrostructure of bars. Normalized = normalized; reheated; quenched and drawn. Delayed quench = delayed quench and drawn.

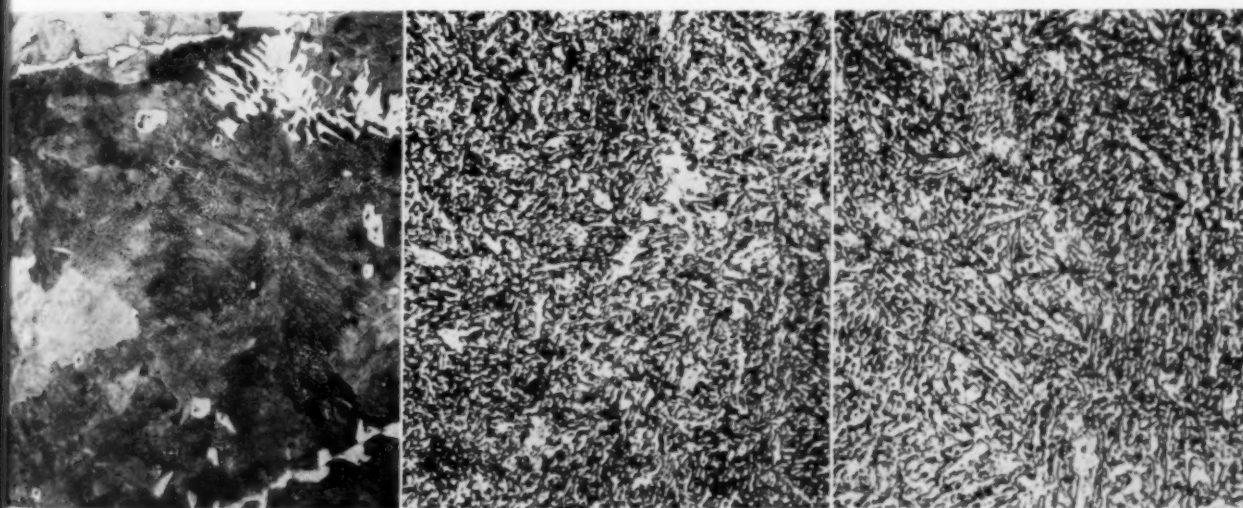


Fig. 9 (left)—As-cast structure of 0.45 carbon steel at  $\times 100$ . Fig. 10 (center)—Same steel normalized at 1700 F, reheated to 1575 F, quenched in water, and drawn at 1225 F.  $\times 1000$ . Fig. 11 (right)—Same steel heated to 1700 F, withdrawn and held in air for 75 sec. to 1350 F, quenched in water, and drawn. Type A bar.  $\times 1000$ . Etch—2% Nital.

### Metallography

Photomicrographs were taken of bars which had been etched in 1:1 hydrochloric acid for 20 minutes at a temperature of 175 F. Type "B" bars were used and were given the complete heat treatment in each case, i.e. normalized, reheated, quenched, and drawn at 1225 F; or delay quench 75 seconds, drawn at 1225 F. Figure 8 illustrates that, for the specimens photographed, the dendrites are, if anything, slightly coarser for the normalized specimens. It would not normally be expected that the reheat at 1575 F would have any appreciable effect on the "as cast" dendritic pattern.

Photomicrographs were taken of the 2 per cent nital-etched specimens. Figures 10 and 11 (1,000  $\times$ ) illustrate the microstructure of Type "A" bars, Heat 224, given both complete heat treatments. Partial spheroidization has been obtained in both cases.

### Production Arrangement for Delayed Quench

At the plant, the following procedure was set up for the delayed quench:

The castings were placed in the rotary oil-fired austenitizing furnace in groups of three, and held for a total of  $3\frac{1}{2}$  hours. After this period, three castings were removed from the furnace and placed on a table equipped with a timing device. When the castings touched the table, the timer was set off and after a total of 85 seconds in still air, the table top was suddenly lifted (mechanically) with the result that the castings were thrust into the quench tank. Figure 12 shows the castings on the time table (two rollers are shown in the two sockets at the right). After the quench, the castings were immediately placed in a second rotary oil-fired furnace for the  $3\frac{1}{2}$ -hour draw at 1225 F. Brinell hardness checks were made twice daily on the "as quenched" castings as a control test. It was found that the time selected always produced fully quenched castings for heats within the proper range of chemical analysis. Statistical analysis showed that the same range of mechan-

ical properties, within normal limits, were produced by the delayed quench as has been obtained with the standard normalize-reheat-and-quench treatment.

Figure 12 (upper right side) shows a batch of ten M-24 tank track shoe castings ready to be put into the austenitizing furnace. It was found that the delayed quench worked very well for the shoes stacked in this manner. For castings having large variations in dimensions, cooling curves may indicate that too great a range in temperature exists within the casting, if the delay in air is to be set up for single castings. Different methods of stacking should be tried in order to narrow down the temperature within the individual castings at the moment of quenching. If the temperature differential cannot be decreased, the delayed quench (in air) heat treatment should not be used as no advantage would be gained for the effort involved.

Cooling curves of various dimension bars having thermocouple wires attached to the fastest cooling parts, as shown by the broken line curve in Fig. 6, will give the foundryman some idea as to whether or not he should consider stacking when contemplating the use of the delayed quench for a casting. The ball socket only has a differential of 85 F in temperature (1350 to 1435 F) when it is delayed in still air for 85 seconds. This is not considered serious, from the viewpoint of cracking due to thermal gradients in the steel. Also, 1435 F is still well below the 1575 F, the temperature used in the reheat treatment method.

After some experience with the delayed quench procedure on different types of castings, it may be possible to establish the time cycle of a new casting merely by obtaining "as quenched" hardness readings after various delays in air, as in Table 3. The point at which the hardness falls off for the fastest cooling section of the casting will then indicate the cycle that should be used.

It should also be mentioned that the delay in air to above the  $A_{r3}$  point can be effected right in the furnace by zoning, i.e., arranging the temperature of the furnace so that the castings on coming out are at the proper temperature for quenching. Although a more uniform temperature range across the casting would be produced by doing this, it is not considered economical as the furnace time taken for the cooling period is much better used for the solution (austenitizing) cycle.

### Summary

1. The setting-up of the "delayed quench" procedure in a foundry and the means for the adoption of this procedure are described. The change in practice includes the following:

a. The actual temperature of the casting in the plant furnace was determined under production conditions.

b. Cooling curves of the casting were obtained as it was held in still air, in order to make sure that the whole casting would be quenched from above the  $A_{r3}$  point, and also to indicate the temperature differential in the casting at the moment of quenching.

c. As-quenched Brinell hardness readings were taken on a series of castings which had had different delays in air prior to quenching. The drop in hardness indicates the critical time factor in air. By correlating with the cooling curve, the lowest temperature from which the casting might be quenched was obtained.

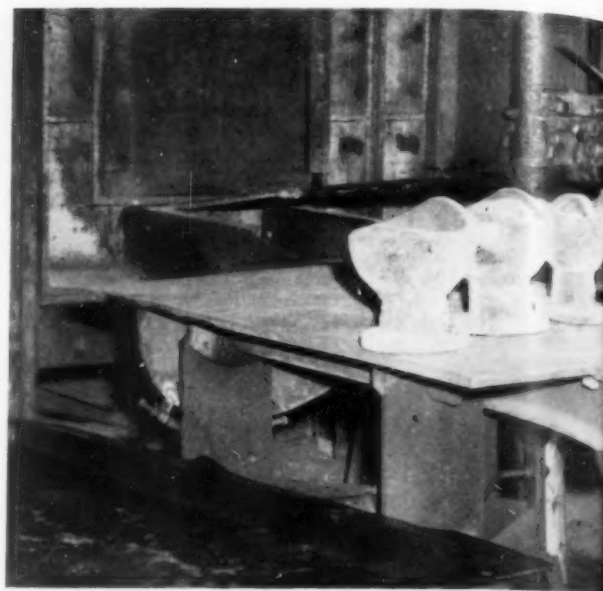


Fig. 12—Three ball sockets on timer table, prior to being quenched.

2. Determination of mechanical properties, of grain sizes, and micro- and macro-structures for both heat treatments indicated that comparable results can be obtained by using the delayed quench procedure. Cooling curves obtained from the middle of the unmachined bars (two types) were used as a guide to establish the time in air which would produce the same temperature at the center of the bars as was obtained on holding the ball socket in air for 85 seconds.

3. The heat treatment that was successfully established for the ball socket was as follows:

a. Holding  $3\frac{1}{2}$  hours in the furnace at the austenitizing temperature, 1700 F (actual temperature of the castings).

b. Cooling in air for 85 seconds to a temperature of 1350-1435 F.

c. Quenching in water.

d. Drawing at 1225 F for  $3\frac{1}{2}$  hours.

Since accurate controls are required and considerable preliminary work is involved, the "delayed quench" procedure is best used for light to medium castings produced in volume. This type of treatment was used for about 75,000 ball sockets and, later, for similarly high numbers of track shoes, rollers, tow hooks, etc. These castings ranged from 10 to 40 pounds in weight and by their design allowed for a proper delay in air without too great a temperature differential being formed within the casting, just prior to quenching.

The heat treatment of castings which differ in size or composition from those described would require modification of the procedure outlined in this paper. It should always be observed that certain castings may not lend themselves to the delayed quench treatment.

### Acknowledgments

The writer wishes to express his gratitude to Mr. C. S. Parsons, Chief of the Bureau of Mines, and to Mr. R. J. Traill, Chief of the Mineral Dressing and

Metallurgical Laboratories, for their authorization of this work. He also wishes to thank the personnel of these Laboratories for their co-operation, and is grateful to Dr. G. S. Farnham of the International Nickel Company of Canada, Limited, for his helpful suggestions.

### References

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2. D. Niconoff, "Critical Points of SAE 4340 Steel as Determined by the Dilatometric Method," *Transactions, ASM*, vol. 37, pp. 337-353, (1946).

### DISCUSSION

Chairman: FRANK KIPER, Ohio Steel Foundry Co., Springfield, Ohio

Co-Chairman: V. E. ZANG, Unitcast Corp., Toledo, Ohio

CHAIRMAN KIPER: In our shop we generally subject low-alloy castings to a preliminary normalize before we do any work on them such as welding. Can we eliminate this operation?

MR. GERTSMAN: As I understand the question, low-alloy castings are normalized before any welding is done at Mr. Kiper's plant. Insofar as repair welding is concerned, preliminary normalizing is unnecessary on 0.50 per cent carbon castings or on 0.30 carbon, 0.25 per cent molybdenum castings of 20 to 40 lb weight. A preheat of 300 to 600 F, however, is required just before welding. If the casting is very intricate a post heat at the same temperature may also be necessary in order to avoid cracking. It should be mentioned that the preheat is necessary whether or not the castings have had a preliminary normalizing. At this plant all the small castings manufactured to the above analyses were cut and ground in the green state. If weld repair was required, a preheat was used prior to the welding operation. All the castings were then loaded for the heat treatment process, whether the delayed quench type of treatment was used or not. Satisfactory results were obtained and the necessity of loading and unloading all the castings twice was avoided.

CO-CHAIRMAN ZANG: We generally give similar steels a preliminary normalize before we do any work on them. There are some people who quench castings direct from the first heat.



# ALUMINUM — ZINC — MAGNESIUM — COPPER CASTING ALLOYS

By

L. W. Eastwood and L. W. Kempf\*

## ABSTRACT

The tensile properties and hardness of aluminum-zinc-magnesium-copper alloys containing approximately 0.25 per cent chromium and 0.15 per cent titanium have been investigated over a range of 0 to 1.75 per cent copper, 3 to 13 per cent zinc, and 0 to 1.0 per cent magnesium. The chromium and titanium were added for their specific effects on resistance to corrosion and grain refinement, respectively. Aluminum ingot which contained approximately 0.15 per cent iron, 0.08 per cent silicon, and 99.75+ per cent aluminum was used as a base. In sand castings, approximately 0.4 per cent copper, 6.6 per cent zinc, 0.33 per cent magnesium, 0.25 per cent chromium, and 0.15 per

cent titanium appear to give a good combination of strength and ductility together with satisfactory resistance to corrosion. Such an alloy ages at room temperature without any previous heat treatment and attains high tensile properties, endurance limit, resistance to failure by impact, and good resistance to corrosion in the accelerated tests utilized in this investigation. Castings of this type of alloy, however, have the disadvantage of being somewhat "hot short." Its tensile properties at elevated temperatures are relatively low, and it overages with the consequent loss of tensile strength and hardness when exposed for a few months at temperatures as low as 212° F. At 300° F this overaging effect is rapid with consequent marked deterioration of the tensile properties and hardness.

ALUMINUM-BASE ALLOYS containing a high zinc content were among the first aluminum alloys used for castings. Because they were used in the early years of the aluminum casting industry, particularly in Europe, their history presents an interesting chapter in the commercial development of this metal. The principal early investigation of binary and more complex aluminum-zinc alloys was done by Rosenhain and Archbutt<sup>1</sup> in 1912. An excellent review including an extensive bibliography of the early development of aluminum-zinc alloys was published by the Bureau of Standards<sup>2</sup> in 1927.

The high zinc-aluminum alloys most widely used in Europe contained 10 to 14 per cent zinc and 2 to 3 per cent copper, largely as a result of the work of Rosenhain and Archbutt. In the United States, Zay Jeffries and William A. Gibson<sup>3</sup> developed an alloy containing 10 to 12 per cent zinc, approximately 2 per cent copper, and 1 to 1.75 per cent iron, which was used quite extensively during the decennium 1920-30. In addition to aluminum-copper-iron-zinc alloys, these investigators in 1919 also patented aluminum-copper-

iron-magnesium-zinc alloys, a preferred composition of which was given as 7 per cent zinc, 3 per cent copper, 0 to 1.5 per cent iron, and 0.1 to 0.3 per cent magnesium.

Early in 1921, the production of heat-treated castings began<sup>4</sup> in the United States, and their continued development and use has been accompanied by the gradual displacement of the high zinc-aluminum alloys. This replacement by the heat-treated alloys occurred because of their higher tensile properties, lower specific gravity, better casting characteristics, and greater resistance to corrosion.

Aluminum alloys containing zinc and magnesium as the principal alloying ingredients have also been in commercial use. At an early date, William Guertler and Wilhelm Sander<sup>5</sup> investigated and proposed the use of aluminum alloys containing magnesium and zinc in the proportions in which they occur in the compound  $MgZn_2$ . Alloys of this type have been used commercially for a number of years<sup>6</sup>. For example, an alloy known as "Constructal 8" containing 7 per cent zinc, 2.5 per cent magnesium, and 1 per cent manganese was first produced in Germany about two decades ago. A similar alloy developed by William Guertler for castings, containing 7.6 per cent zinc and 1.7 per cent magnesium, has been designated G.W.32. Another containing 6 per cent zinc, 1.2 per cent magnesium and 2 per cent iron was developed by T. S. Fuller and David Basch<sup>7</sup> in the United States.

In 1937 and 1938 a series of United States patents was issued to Yonosuke Matuenaga<sup>8</sup> on aluminum

\* The research investigation that provided data for this paper was conducted by Dr. L. W. Eastwood and Mr. L. W. Kempf while with the Cleveland Research Division, Aluminum Research Laboratories, Aluminum Company of America. This work was completed in 1941, and the report on the results was submitted to the National Advisory Committee for Aeronautics through which restricted circulation was controlled. Mr. Kempf is deceased, and Dr. Eastwood is now a Supervising Metallurgist at Battelle Memorial Institute, Columbus, Ohio.

alloys containing zinc, magnesium, and copper as the principal alloying constituents. Aluminum-zinc-magnesium alloys having a preferred composition of 8.0 per cent zinc, 0.5 per cent chromium, 0.80 per cent magnesium, 0.10 per cent titanium, less than 0.3 per cent copper, silicon or manganese and less than 0.8 per cent iron have been proposed by G. F. Comstock<sup>9</sup>. There have been many other investigations of aluminum alloys containing zinc and magnesium as the principal alloying constituents, but a complete review here is unnecessary.

This paper contains data on the mechanical properties of sand-cast test bars having a considerable range in zinc, magnesium, and copper content and more detailed data on the properties and foundry characteristics of an aluminum-zinc-magnesium-copper alloy having a preferred composition for high strength and ductility. The principal portion of the data is on alloys which also contain approximately 0.15 per cent of titanium and 0.25 per cent chromium. The titanium and chromium were added because they were found to effect grain refinement and improve resistance to corrosion, respectively.

### Experimental Procedure

#### PREPARATION OF THE ALLOYS

Aluminum ingot containing 99.75 per cent aluminum and the alloy components were carefully weighed. The magnesium and zinc were added as such, but the chromium, titanium, and copper were added in the form of aluminum-base riches prepared with 99.8 per cent aluminum. The melts were made in plumbago crucibles in gas-fired furnaces. First the aluminum was melted down, and then the alloying material, except the magnesium, was added and melted.

The melt was fluxed with chlorine for 10 to 15 minutes, and then the magnesium was added. If a variation in the content of magnesium, zinc, or copper was desired, these metals were successively added in the proper amounts after each set of the test castings had been poured. Each set of castings in the series was then given the same heat number, with the letters A, B, C, etc., attached in the order of pouring.

#### TEST CASTINGS

Cast-to-size test bars were made using a 6-bar casting, a photograph of which is shown in Fig. 10. These test bars are  $\frac{1}{2}$ -in. in diameter at the test section. The test-bar castings used for corrosion tests were cast somewhat oversize and machined to  $\frac{1}{2}$ -in. diameter at the test section. A casting having heavy sections was used to determine the effect of section size. This casting is shown in Fig. 11 with the gates and risers attached. Each section is  $\frac{3}{4}$  in. long, 4 in. wide, and 1, 2, and 4 in. thick. Test bars 0.505 in. in diameter at the test section were machined from this casting. All castings were poured from 1350° F unless otherwise noted, and all were made in green sand molds.

#### AGING TREATMENT

After the castings were made, they were aged as indicated by the data in the tables or figures. Usually the castings were aged for 30 days at a room tempera-

ture maintained at about 85° F. In some instances an equivalent aging treatment was effected by using a shorter time at a slightly elevated temperature. This subject is treated more fully elsewhere in this paper.

#### CORROSION TESTS

Corrosion tests were made on separately cast test bars. The investigation of the corrosion of test bars under an applied stress was conducted by using the experimental procedure described by E. H. Dix, Jr.<sup>10</sup>

The general corrosion characteristics of unstressed bars were determined in salt-spray exposures in a manner described by E. H. Dix, Jr. and J. J. Bowmann.<sup>11</sup> The equipment used is illustrated by Fig. 1 and 3 of their paper "Salt Spray Testing." Hard rubber spray nozzles were employed, and the salt-spray boxes contained six vents each. The air at 40 psi was passed through a cleaning tower and then through a water column maintained at 85° F to saturate it at the test temperature before it entered the boxes.

The salt-spray exposures, maintained at 85° F, were of two types, continuous and intermittent. A 20-per cent-salt solution was used for the continuous exposure, and a  $3\frac{1}{2}$ -per cent-salt solution for the intermittent exposure, Morton's Flake Butter Salt being used for both types. The intermittent cycle comprised 16 hr. with the box closed and the spray operating, and 8 hr. with the box open and the spray not operating.

Six cylindrical test bars made in green sand and machined to 0.5 in. at the test section were suspended from glass rods for each type of exposure. In order to remove dust and grease which might have interfered with the test, the bars were cleaned in petroleum ether before starting the exposures.

#### TENSILE AND HARDNESS TESTS

Except those made on heavy sections, all tensile tests were made on separately cast test bars without machining the gage section. Yield strength values were obtained at the point of 0.2-per cent deviation from the modulus line.

Brinell hardness was obtained by using a 500-kg. load and a 10-mm ball.

Elongation values were determined on a 2-in. gage length.

#### Experimental Results

The experimental results obtained are presented in the accompanying tables and figures. The effects of zinc, magnesium, and copper content on the tensile and hardness properties of cast test bars have been investigated over a considerable range. The resistance to corrosion of these alloys under an externally applied stress also has been investigated. A relatively narrow concentration range of zinc, magnesium, and copper has been more thoroughly investigated in respect to aging characteristics, effects of exposure to elevated temperatures, high-temperature tensile properties, tensile properties in heavy sections, foundry characteristics, and physical properties.

#### EFFECT OF ZN, MG AND CU CONTENT

The effects of the magnesium content on the tensile strength, per cent elongation in 2 in., yield strength, and Brinell hardness of alloys containing 3 to 13 per

cent zinc and approximately 0.4 per cent copper, 0.25 per cent chromium, 0.15 per cent titanium, 0.15 per cent iron, and 0.08 per cent silicon are shown by Fig. 1a, 1b, 1c, and 1d, respectively. Data on similar alloys containing 1.0 per cent copper instead of 0.4 per cent are graphically represented by Fig. 2a, 2b, 2c, and 2d, respectively. A third set of data on a series containing 1.75 per cent copper is represented by Fig. 3a, 3b, 3c, and 3d.

By the use of curves which represent the tensile strength and per cent elongation, it is possible to determine the maximum and minimum amounts of magnesium at each zinc content which will give desired minimum values of tensile strength and elongation. This has been done for certain values, and the results are represented graphically by Fig. 4. This figure shows the range in zinc and magnesium content at which minimum tensile properties of 36,000 psi. and 10 per cent elongation, and 34,000 psi. and 7 per cent elongation were attained under the experimental conditions utilized. The former minimum values of tensile properties are represented by the inside area bounded by full lines, and the latter values by the entire area bounded by the dashed lines.

Examination of Fig. 4 shows that the higher the zinc content, the lower the magnesium content for maximum combinations of strength and ductility. The shape of the areas representing the zinc and magnesium concentrations for minimum tensile properties of 36,000 psi. and 10 per cent elongation is quite similar at each of the three values of copper content represented. However, increasing the copper content decreases the size of areas representing the range of magnesium and zinc for these minimum properties; and for a given zinc content, increasing the copper requires decreasing magnesium concentration.

Table I shows, for three series of alloys containing 0.4, 1.0, and 1.75 per cent copper and 4 to 13 per cent zinc, the magnesium content required for the attainment of 36,000 psi. tensile strength and 10 per cent elongation.

As indicated in Fig. 5, increasing the zinc content increases the tensile and yield strengths and hardness, but decreases the elongation of alloys containing approximately 0.35 per cent copper, 0.15 per cent iron, 0.08 per cent silicon, 0.13 per cent titanium, 0.25 per cent chromium, 0.27 and 0.29 per cent magnesium.

#### EFFECT OF FE AND SI CONTENT

Iron and silicon may be regarded as impurities since they do not improve the mechanical properties. These elements invariably occur as impurities in aluminum. Accordingly, it is desirable to know the effects of these impurities on the properties. The available data listed in Table 2, though not extensive, show the effects of iron and silicon separately and in combination. Fig. 6 also shows that increasing concentrations of silicon

TABLE 1—MAGNESIUM CONCENTRATION AT VARIOUS ZINC AND COPPER CONTENTS FOR ATTAINMENT OF 36,000 PSI. TENSILE STRENGTH AND 10% ELONGATION

Zn., %	0.4% Cu.	1.0% Cu.	1.75% Cu.
	Optimum Magnesium Content		
	Mg., %	Mg., %	Mg., %
5	0.54	—	—
6	0.42	0.41	—
7	0.31	0.29	.16
8	0.20	0.15	.07
9	0.12	0.06	.02
10	0.07	0.02	.01
11	0.04	0.005	.005
12	0.005	—	—

TABLE 2—EFFECT OF IRON AND SILICON CONTENT ON TENSILE PROPERTIES OF TEST BARS OF AL-ZN-MG-CU-TI-CR ALLOYS

Heat No.	Aging Time	Cu.	Fe.	Si.	Zn.	Ti.	Mg.	Cr.	Y.S.	T.S.	% El.	B.H.N.
Section A												
Effect of Silicon Content												
555	30 days at 85°F.	0.32	0.14	0.08	6.65	0.13	0.33	0.24	21700	38200	12.8	73
556	30 days at 85°F.	0.37	0.14	0.19	6.58	0.13	0.32	0.20	21600	34900	10.2	70
557	30 days at 85°F.	0.37	0.15	0.33	6.51	0.13	0.33	0.22	23100	33950	6.2	74
063A	30 days at 85°F.	0.003	0.24	0.08	6.80	0.18	0.31	0.00	21800	35200	9.5	74
B	30 days at 85°F.	*	*	0.18	*	*	*	0.00	20000	31800	8.2	69
C	30 days at 85°F.	*	*	0.30	*	*	*	0.00	19800	30000	6.3	70
Section B												
Effect of Iron Content												
062A	30 days at 85°F.	0.003	0.29	0.10	6.91	0.15	0.28	0.00	22000	36200	10.2	74
B	30 days at 85°F.	*	0.46	*	*	*	*	0.00	21600	36600	10.7	74
C	30 days at 85°F.	*	0.56	*	*	*	*	0.00	20400	35300	9.0	73
Section C												
Effect of Ingot Purity												
067	30 days at R.T.**	0.01	0.11	0.12	7.06	0.16	0.37	0.00	23100	36300	9.0	74
069	30 days at R.T.**	0.01	0.33	0.16	7.13	0.16	0.36	0.00	22300	34100	7.3	74
409	30 days at R.T.**	0.30	0.16	0.07	6.98	0.18	0.26	0.23	20300	36600	13.3	69
413	30 days at R.T.**	0.27	0.23	0.10	7.00	0.18	0.26	0.25	20200	35350	11.7	69
410	30 days at R.T.**	0.32	0.30	0.14	6.96	0.19	0.27	0.24	20300	34900	9.8	69
412	30 days at R.T.**	0.32	0.32	0.14	6.95	0.23	0.30	0.24	21000	35200	9.2	70

\*The chemical analysis may be assumed to be approximately the same as the A sample except as noted.

\*\*Room temperature.



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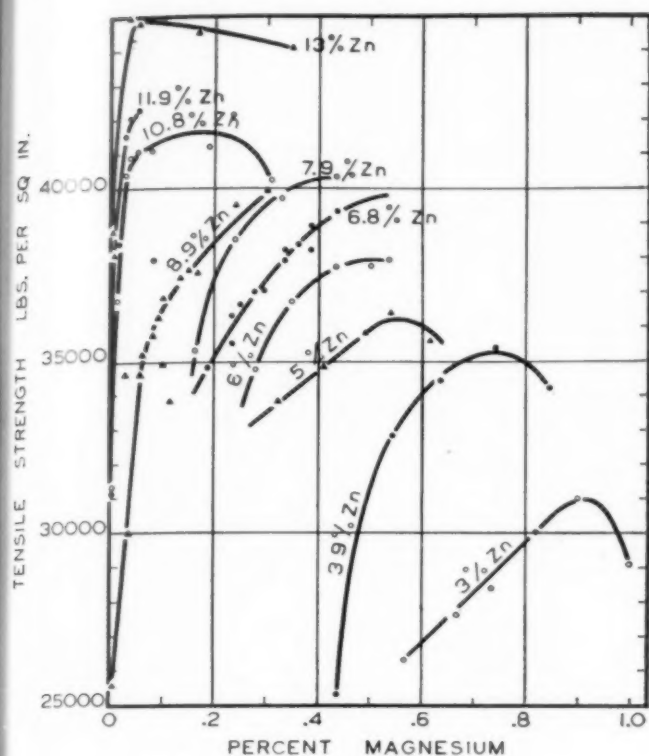


Fig. 1a—Effect on tensile strength

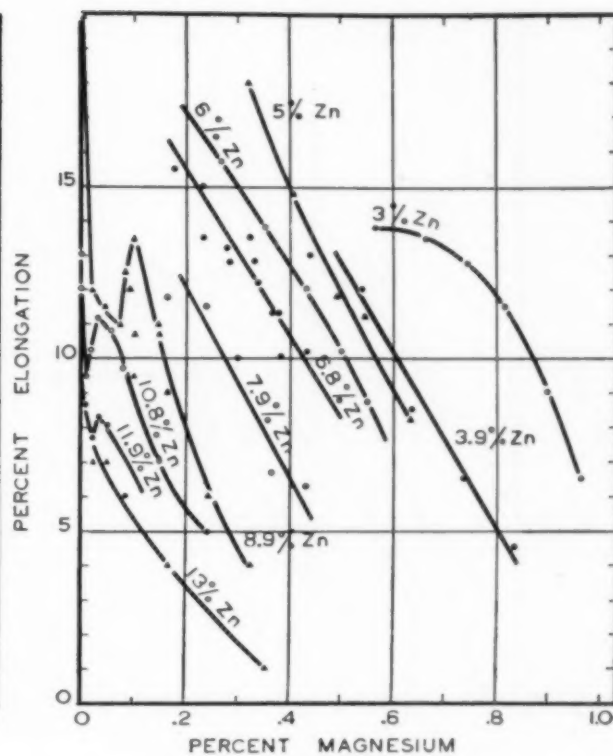


Fig. 1b—Effect on per cent elongation in 2 in.

Fig. 1—Effect of magnesium content on mechanical properties of alloys containing approximately 0.4% Cu, 0.15% Fe, 0.08% Si, 0.25% Cr, and various amounts of Zn. All alloys as cast and aged 30 days at 85° F.

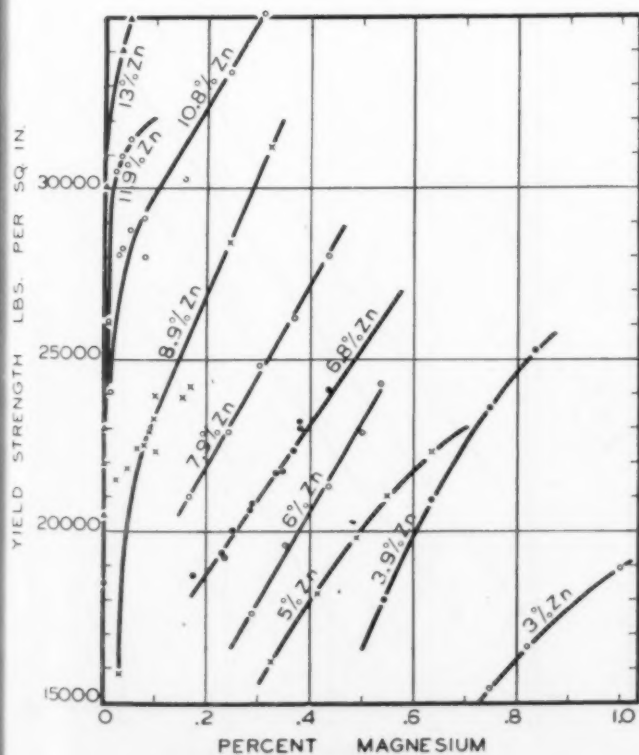


Fig. 1c—Effect on yield strength

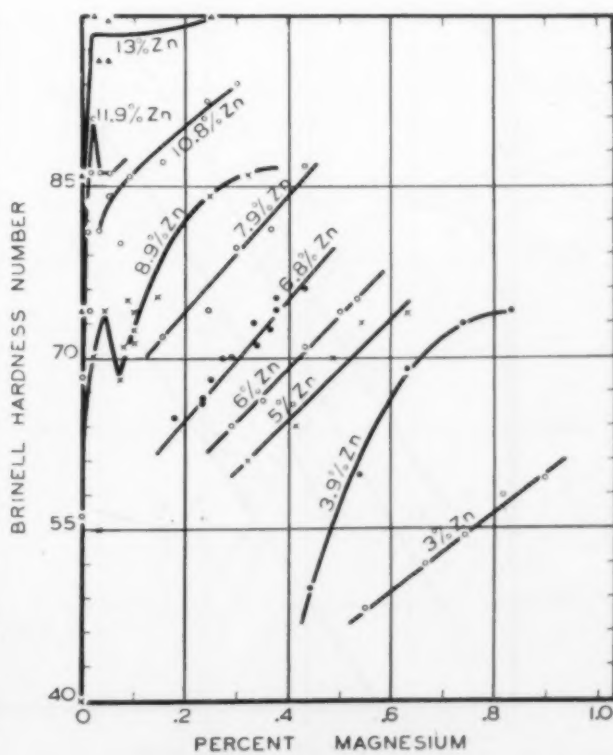


Fig. 1d—Effect on Brinell hardness

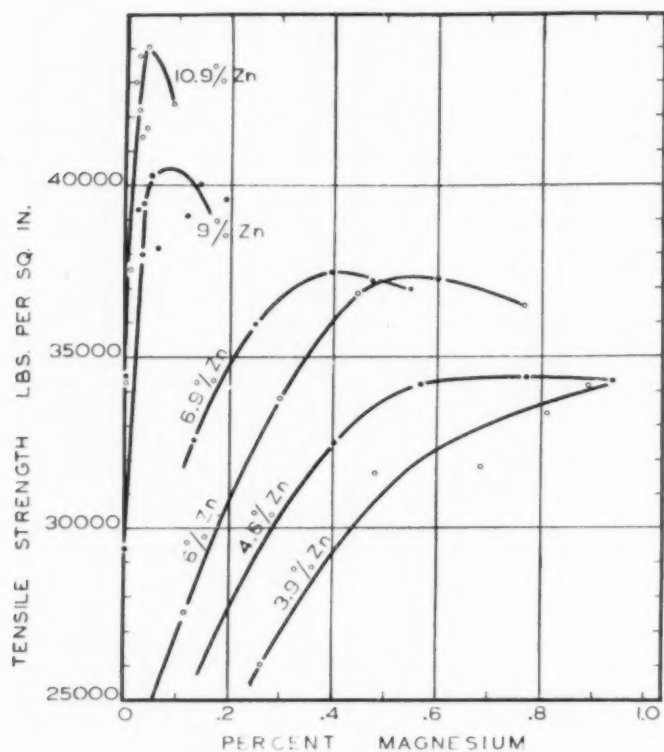


Fig. 2a—Effect on tensile strength

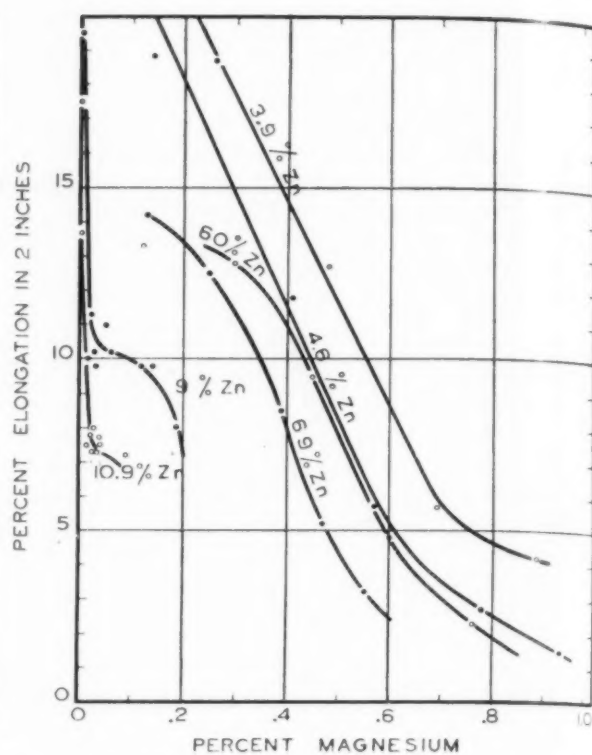


Fig. 2b—Effect on per cent elongation in 2 in.

Fig. 2—Effect of magnesium content on mechanical properties of alloys containing approximately 1.0% Cu, 0.15% Fe, 0.08% Si, 0.25% Cr, and various amounts of Zn. All alloys as cast and aged 30 days at 85° F.

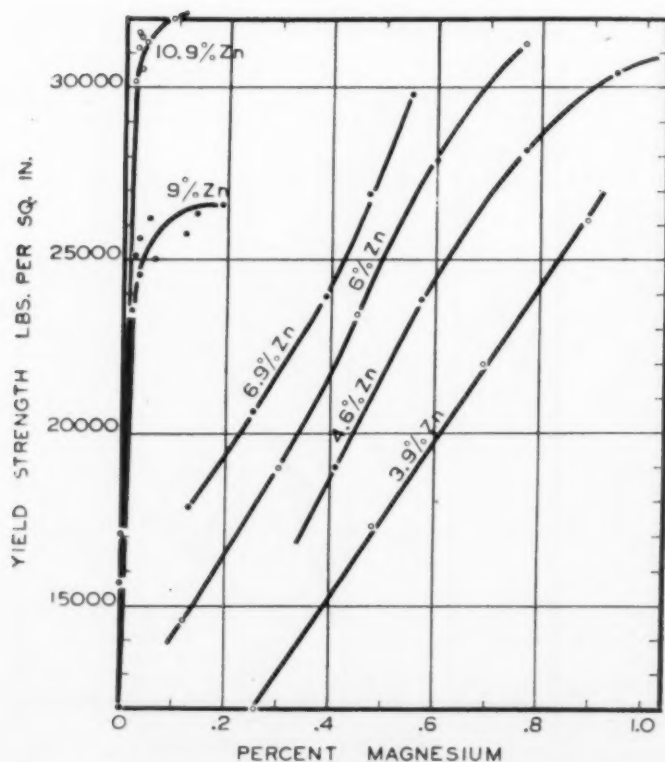


Fig. 2c—Effect on yield strength

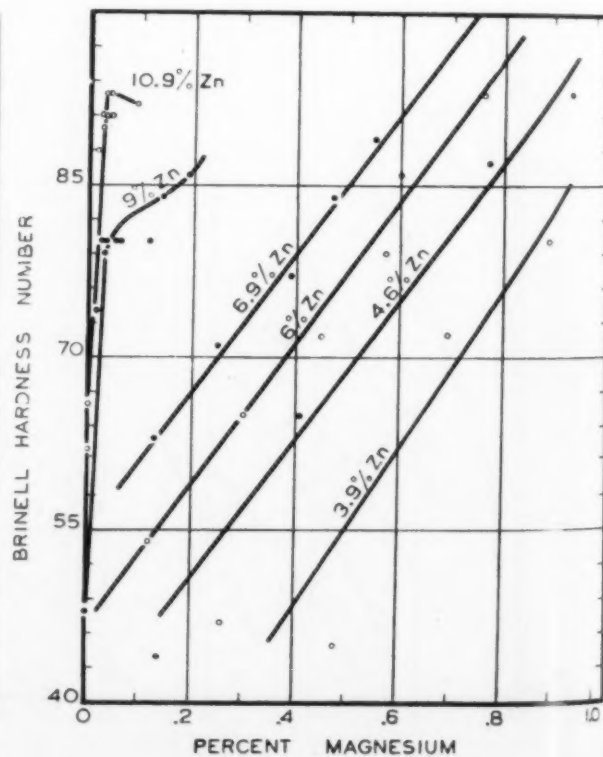


Fig. 2d—Effect on Brinell hardness

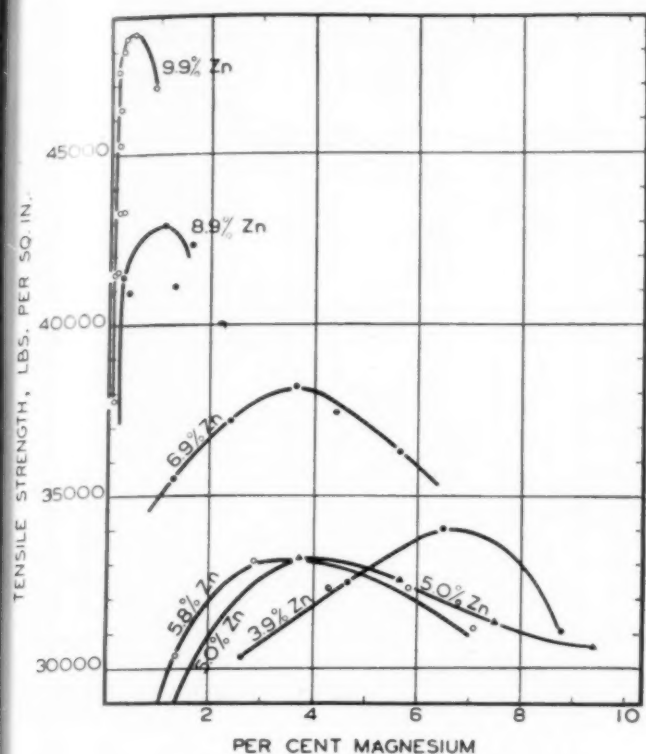


Fig. 3a—Effect on tensile strength

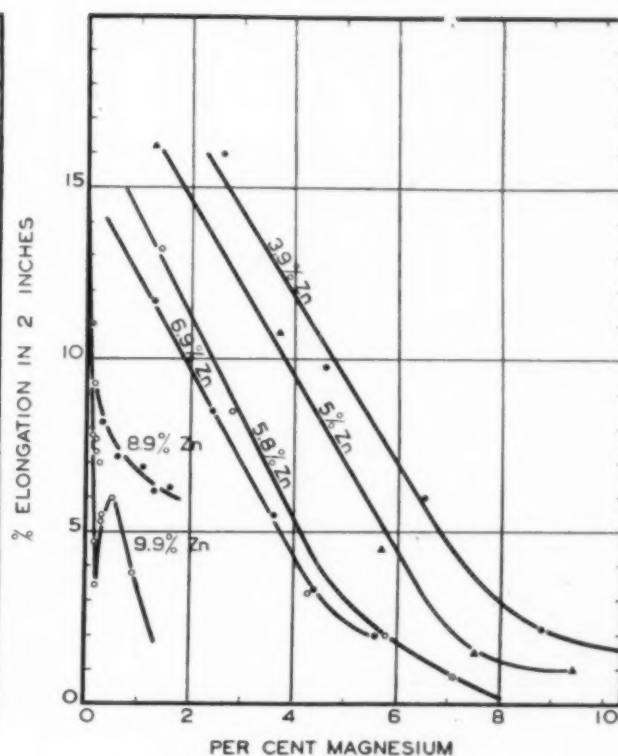


Fig. 3b—Effect on per cent elongation in 2 in.

Fig. 3—Effect of magnesium content on mechanical properties of alloys containing approximately 1.75% Cu, 0.15% Fe, 0.08% Si, 0.25% Cr, and various amounts of Zn. All alloys as cast and aged 30 days at 85° F.

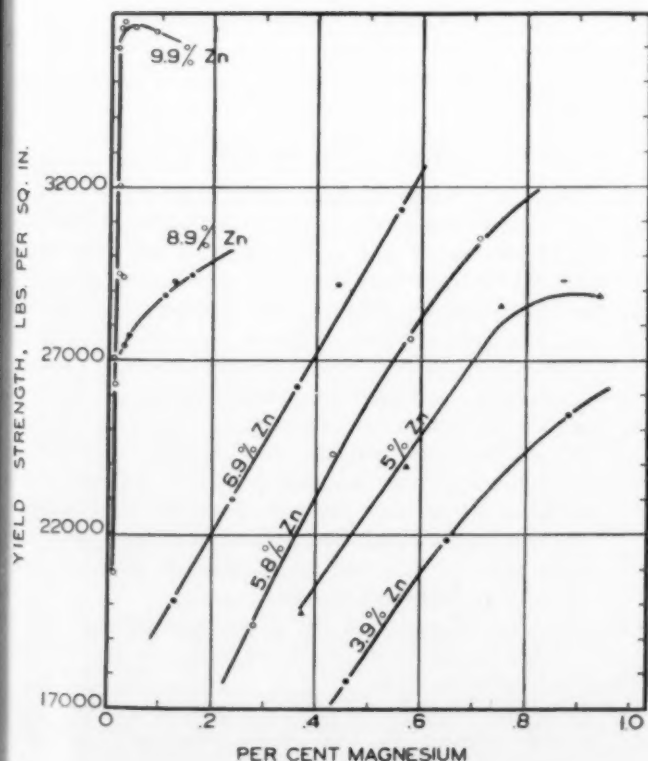


Fig. 3c—Effect on yield strength

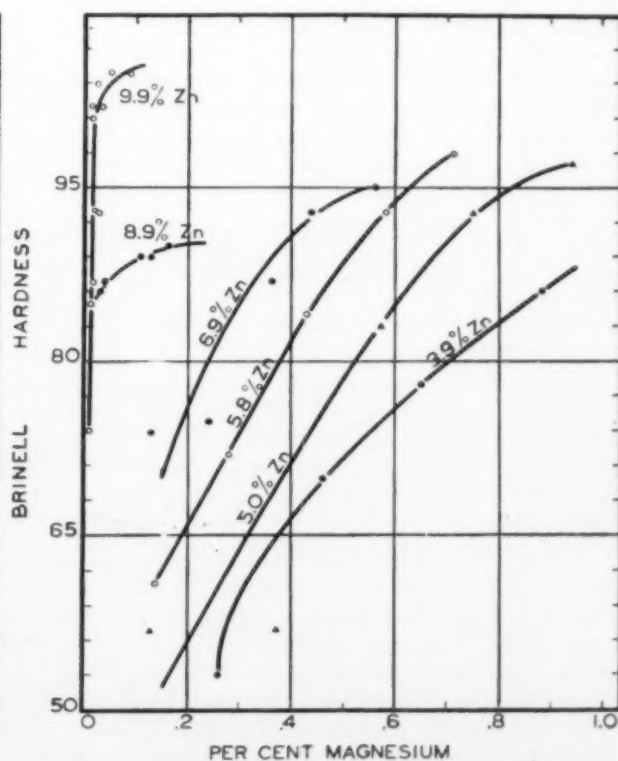


Fig. 3d—Effect on Brinell hardness



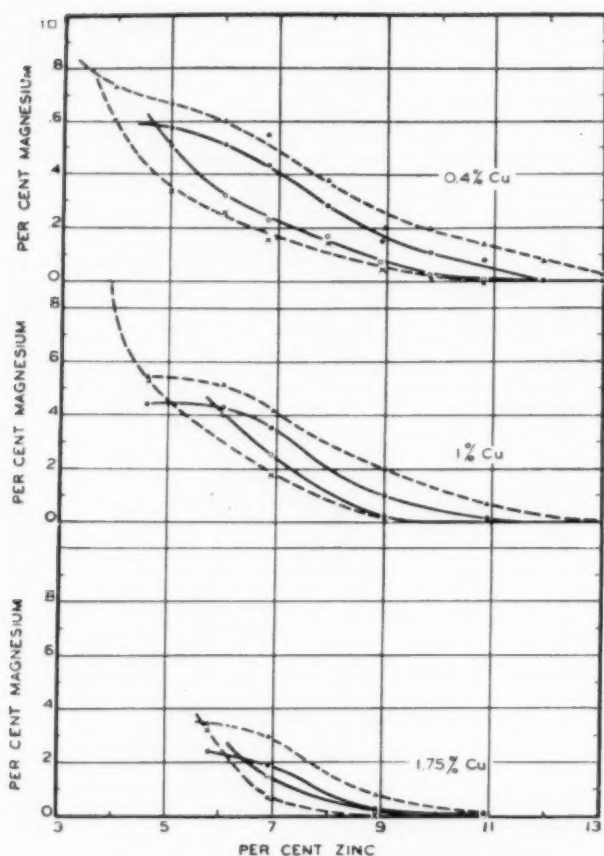


Fig. 4—Inside area represents range in Mg and Zn contents of cast test bars which will generally attain, after aging 30 days at 85° F, a min. tensile strength and elongation of 36,000 psi. and 10 per cent, respectively. Composition represented by outside dashed lines will generally produce a min. of 34,999 psi. and 7 per cent elongation. These data are for aluminum alloys containing 0.15% Fe, 0.08% Si, 0.15% Ti, and 0.25% Cr and the Cu, Mg, and Zn contents indicated.

have a very adverse effect on the tensile strength and ductility of alloys containing 0.38 per cent copper, 0.15 per cent iron, 6.6 per cent zinc, 0.13 per cent titanium, 0.2 per cent chromium, and 0.08, 0.19, and 0.33 per cent silicon.

Iron alone has only a slightly adverse effect on the tensile properties even when 0.5 per cent is present. When iron and silicon are increased simultaneously, the tensile strength and ductility are reduced to about the same extent as they would be if the silicon alone were increased.

The adverse effect of silicon is probably due to the formation of  $Mg_2Si$  which depletes the effective magnesium content and forms a brittle grain boundary constituent. With the aluminum at present commercially available, it probably is not practical to specify silicon concentrations lower than about 0.25 per cent. The mechanical properties of commercial castings might be expected to be lower than those obtained in this investigation on alloys containing about 0.08

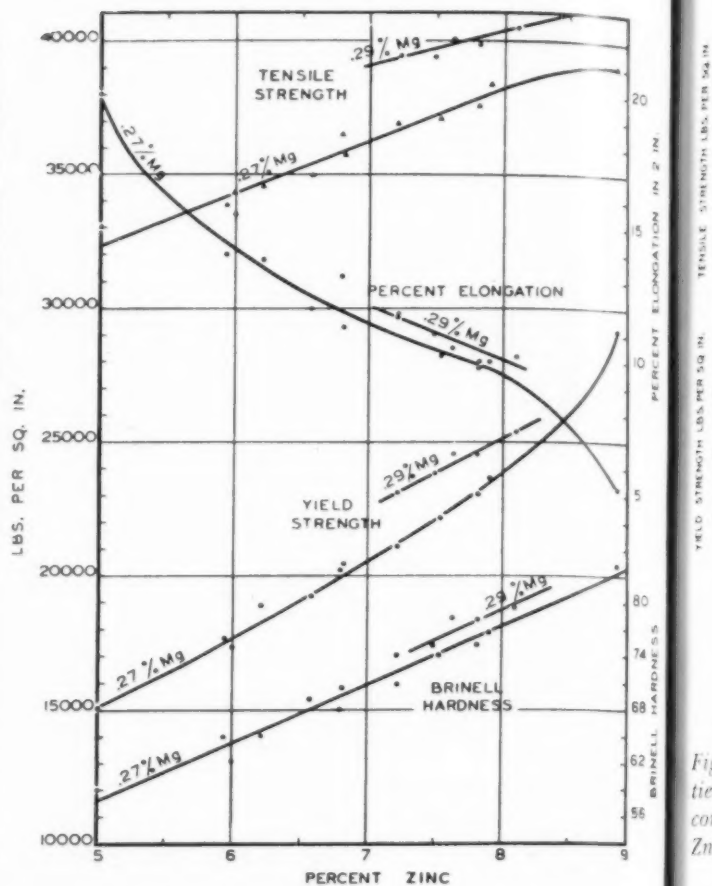


Fig. 5—Effect of zinc content on the tensile properties and hardness of cast test bars of aluminum alloys containing approximately 0.35% Cu, 0.18% Fe, 0.08% Si, 0.13% Ti, 0.28% Cr, and 0.27% and 0.29% Mg. All alloys aged 30 days at 85° F.

per cent silicon in about the ratio indicated in Table 2.

#### RESISTANCE TO CORROSION

Four alloys were exposed in salt spray in the manner described above. Two of these alloys contained approximately 7 per cent zinc, 0.3 per cent magnesium, 0.15 per cent iron, 0.08 per cent silicon, 0.15 per cent titanium, 0.35 per cent copper, and 0.00 per cent chromium. One of these was prepared from 99.99+ per cent zinc, and the other from 99.5 per cent zinc. The other two alloys were similar in composition to the first two described, but both were made with 99.99+ per cent zinc and both contained 0.25 per cent chromium. One of these two contained 0.35 per cent copper and the other, 1.0 per cent copper. The results obtained after two years\* of exposure of unstressed cast test bars to continuous and intermittent salt sprays are as follows:

1. The resistance of the alloys containing 0.25 per cent chromium is of the same order as that of the chromium-free alloy.

2. The alloy containing 1.0 per cent copper appeared inferior to that containing 0.35 per cent copper.

\* Text has been revised after preprinting of this paper.

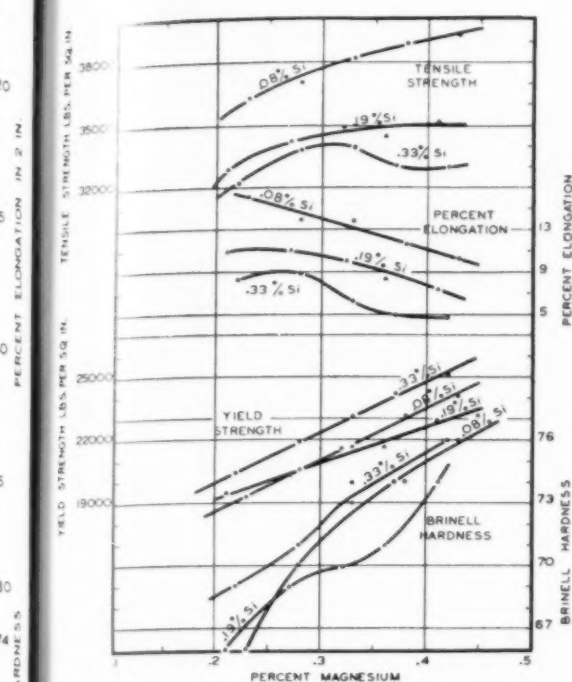


Fig. 6—Effect of magnesium content on tensile properties and hardness of cast test bars of aluminum alloys containing approximately 0.38% Cu, 0.15% Fe, 6.6% Zn, 0.13% Ti, 0.22% Cr, and 0.08, 0.19, and 0.33% Si. All alloys aged 30 days at 85° F.

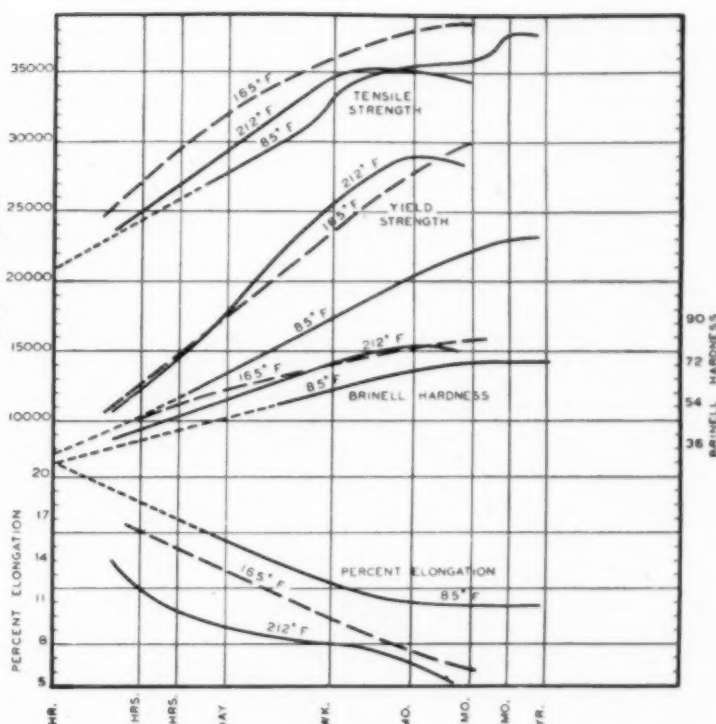


Fig. 7—Effect of aging time at 85° F, 115° F, and 212° F on tensile properties and hardness of cast test bars of an aluminum alloy containing 0.38% Cu, 0.17% Fe, 0.08% Si, 6.88% Zn, 0.12% Ti, 0.27% Mg, 0.23% Cr.

3. The resistance to this type of corrosion is not noticeably affected by the degree of purity of the zinc.

4. An alloy containing approximately 0.35 per cent copper, 0.15 per cent iron, 0.08 per cent silicon, 7.0 per cent zinc, 0.15 per cent titanium, and 0.25 per cent chromium has good resistance to this type of corrosion; it is about equivalent to Alcoa No. 43 alloy consisting of aluminum with 5 per cent silicon.

It has also been found that bars stressed at 75 per cent or less of the yield strength are not subject to intergranular corrosion when continuously immersed in a solution of NaCl and H<sub>2</sub>O<sub>2</sub>, provided the zinc does not exceed about 7.0 per cent and the copper is not less than 0.25 per cent or more than 0.6 per cent.

#### PREFERRED COMPOSITION

On the basis of the results on tensile properties and resistance to corrosion previously discussed, an alloy containing approximately 0.4 per cent copper, 0.15 per cent iron, 0.08 per cent silicon, 6.6 per cent zinc, 0.33 per cent magnesium, 0.15 per cent titanium, and 0.25 per cent chromium was selected for more detailed investigation.

#### AGING CHARACTERISTICS

Table 3 and Fig. 7 show the changes in tensile properties and hardness of cast test bars of an alloy containing 0.38 per cent copper, 0.17 per cent iron, 0.08 per cent silicon, 6.88 per cent zinc, 0.12 per cent titanium, 0.27 per cent magnesium, and 0.23 per cent chromium. Three different aging temperatures were

used, that is, 85°, 165°, and 212° F. The curves of Fig. 7 show that the best combination of tensile strength and ductility is attained by aging at 85° F. It will be noted that one week at 165° F is approximately equivalent to one month at 85° F. This alloy overages even at 212° F when the time at temperature is of several months' duration. Overaging manifests itself by a greatly reduced ductility and some drop in yield strength and hardness. After three months at 165° F, there is no softening noticeable, but the ductility as measured by the per cent elongation is greatly reduced. This might be compensated by starting out with a lower magnesium content and higher initial ductility.

#### TENSILE PROPERTIES IN HEAVY SECTIONS

Table 4 contains data on the tensile properties in heavy sections of an alloy having about the preferred composition referred to above. These data clearly show that a very high percentage of the properties contained in separately cast test bars is obtained in bars machined from this 18-lb. casting having sections 1, 2, and 4 in. thick.

#### HIGH-TEMPERATURE TENSILE PROPERTIES

The tensile strength and elongation of an alloy containing 1.06 per cent copper, 0.16 per cent iron, 0.08 per cent silicon, 6.89 per cent zinc, 0.13 per cent titanium, 0.26 per cent magnesium, and 0.27 per cent chromium at 200°, 300°, 400°, 500°, and 600° F are

TABLE 3—CHANGE IN TENSILE PROPERTIES OF CAST TEST BARS OF AN AL-ZN-MG-CU-TI-CR ALLOY WITH AGING TIME AT VARIOUS TEMPERATURES

Heat No.	Aging Treatment	Y.S.	T.S.	% El.	B.H.N.
559*	1 hour after casting...	6550	20875	21.2	34
559*	3 days at 85°F.....	15375	30425	13.3	56
559*	7 days at 85°F.....	17575	33350	13.8	61
559*	31 days at 85°F.....	20325	35575	10.5	66
559*	60 days at 85°F.....	21250	35800	10.7	73
559*	90 days at 85°F.....	21925	35925	9.5	72
559*	182 days at 85°F.....	23100	37850	11.5	74
559*	307 days at 85°F.....	23200	37800	10.5	72
559*	1 hour after casting...	6550	20875	21.2	34
559*	2½ hours at 165°F.....	10675	24675	16.8	43
559*	19 hours at 165°F.....	17100	31500	13.7	61
559*	68 hours at 165°F.....	20975	33550	10.7	69
559*	6 days at 165°F.....	23675	35850	10.0	70
559*	10 days at 165°F.....	24600	36250	9.0	70
559*	14 days at 165°F.....	26600	37375	9.2	77
559*	21 days at 165°F.....	26600	37700	8.3	77
559*	31 days at 165°F.....	28000	37100	6.7	80
559*	45 days at 165°F.....	28450	37900	7.5	79
559*	60 days at 165°F.....	28800	38500	7.7	86
559*	90 days at 165°F.....	30000	38525	6.2	84
559*	1 hour after casting...	6550	20875	21.2	34
559*	3 hours at 212°F.**...	10825	23875	14.0	41
559*	20 hours at 212°F.....	17125	28725	9.5	56
559*	69 hours at 212°F.....	21525	31975	8.2	66
559*	6 days at 212°F.....	25325	34800	8.7	72
559*	10 days at 212°F.....	26825	34725	8.0	75
559*	14 days at 212°F.....	27550	35425	7.5	77
559*	21 days at 212°F.....	26300	35300	7.8	75
559*	31 days at 212°F.....	29025	35350	5.0	77
559*	36 days at 212°F.....	29050	34900	5.7	81
559*	60 days at 212°F.....	28600	35800	6.5	78
559*	70 days at 212°F.....	28650	34600	5.3	78

\*The analysis of these bars was as follows: 0.38% Cu., 0.17% Fe., 0.08% Si., 6.88% Zn., 0.12% Ti., 0.27% Mg., 0.23% Cr.

\*\*These test bars were placed in boiling water. The resulting extremely slight corrosion after long exposure probably has contributed to the apparent adverse aging effect under these conditions.

shown by the data in Table 5. The elongation at room temperature for the particular lot of test specimens used for the determination of these data and those of Tables 6, referred to in the next paragraph, is lower than normal for some unknown reason. Nevertheless, the conclusion is probably justified in that the high temperature properties of this type of alloy are somewhat inferior to those of many present day commercial aluminum-casting alloys.

#### EFFECT OF PROLONGED EXPOSURE AT 300° AND 400° F

Table 6 shows the effect of prolonged exposure at 300° and 400° F on the tensile and hardness properties of the alloy referred to in the preceding paragraph. These data show that exposure to such temperatures has an adverse effect on the room-temperature tensile properties because of the over-aging effect.

#### EFFECT OF EXPOSURE TO A TEMPERATURE NEAR THE MELTING POINT

Aluminum-zinc-magnesium-copper alloys can be reheated to temperatures near the melting point without a marked adverse effect on tensile properties. Table 7 shows the effects of reheating a specific alloy to temperatures from 900° to 1120° F, air cooling, and re-aging at room temperature. These data show that reheating to 900° F has a slightly adverse effect on the tensile properties, where as reheating to 1050° to 1120° F does not affect the tensile properties. Of course, when this alloy is reheated to such temperatures, the tensile properties are about equivalent to those obtained immediately after casting, and re-aging is necessary to restore them. The amenability of these alloys to reheating to a high temperature makes them attractive for use in furnace-brazed assemblies.

#### MECHANICAL AND PHYSICAL PROPERTIES

Using the experimental conditions outlined, an

TABLE 4—TENSILE PROPERTIES OF AN AL-ZN-MG-CU-CR-TI ALLOY\* IN SEPARATELY CAST TEST BARS AND IN HEAVY SECTIONS POURED FROM 1350° F AND AGED 1 YEAR AT 85° F

Heat No.	Aging Treatment	Casting	Section Thickness, In.	Y.S.		T.S.		% El.		B.H.N. Aver.
				Min.	Aver.	Min.	Aver.	Min.	Aver.	
366	1 year at 85°F.....	6-bar step	0.5	24600	24750	36000	37600	7.5	8.8	74
			1.0	22000	22230	26750	30450	2.5	5.5	70
			2.0	20600	21500	34200	37200	11.5	16.2	74
			4.0	14550	18600	22500	32900	7.5	16.2	69
365	1 year at 85°F.....	6-bar step	0.5	25400	25600	38300	39450	8.5	9.6	80
			1.0	23300	23550	34700	35800	7.5	17.7	74
			2.0	16850	17900	31700	32900	13.0	14.6	68
			4.0	13400	17500	16150	30400	3.5	9.6	68
413	60 days at 85°F.....	6-bar step	0.5	19100	19600	35800	36100	12.5	14.4	72
			1.0	18900	19500	32100	33400	9.5	10.5	70
			2.0	18500	20100	33100	36600	11.0	15.7	71
			4.0	18900	19900	32800	35700	9.0	13.9	70

\*The analyses are as follows:

Heat No.	% Cu.	% Fe.	% Si.	% Zn.	% Ti.	% Mg.	% Cr.
366.....	0.33	0.17	0.08	7.08	0.16	0.30	0.26
365.....	0.34	0.17	0.08	7.02	0.18	0.33	0.26
413.....	0.27	0.23	0.10	7.00	0.18	0.26	0.25

alloy containing approximately 0.35 per cent copper, 0.15 per cent iron, 0.08 per cent silicon, 6.6 per cent zinc, 0.15 per cent titanium, 0.33 per cent magnesium, and 0.25 per cent chromium may be expected to have approximately the following mechanical and physical



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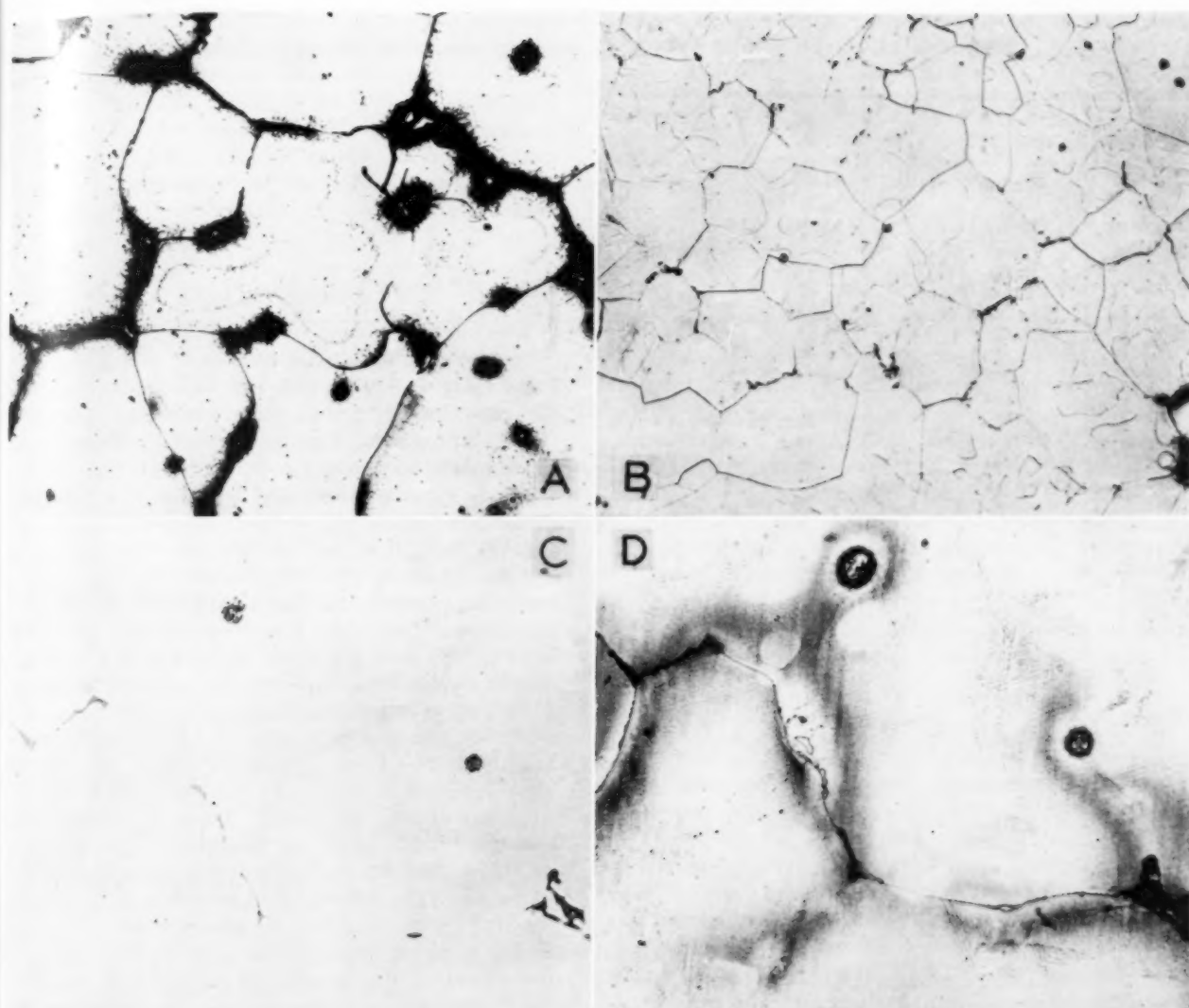


Fig. 8—(A) Photomicrograph at x100 of an alloy containing 1.05% Cu, 0.15% Fe, 0.13% Si, 7.02% Zn, 0.14% Ti, 0.29% Mg and 0.30% Cr. Keller's etch. (B) Photomicrograph at x100 of an alloy containing 0.34% Cu, 0.17% Fe, 0.08% Si, 7.02% Zn, 0.18% Ti, 0.33% Mg and 0.26% Cr. Keller's etch. (C) Photo-

micrograph at x500 showing the elongated irregular gray constituent alpha Al-Fe-Si. The rounded light gray constituent is CuAl<sub>2</sub> rosettes. Unetched. Composition similar to that of (B). (D) Same as (C). Etched with 10% NaOH solution in water. The CuAl<sub>2</sub> rosettes are light and the alpha Al-Fe-Si constituent is black.

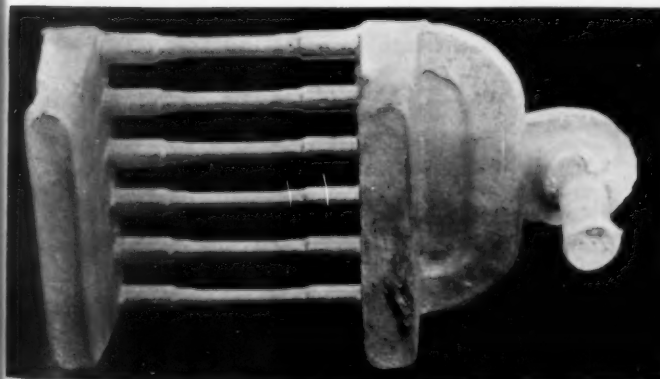


Fig. 9—The six-bar casting with gate and risers attached.

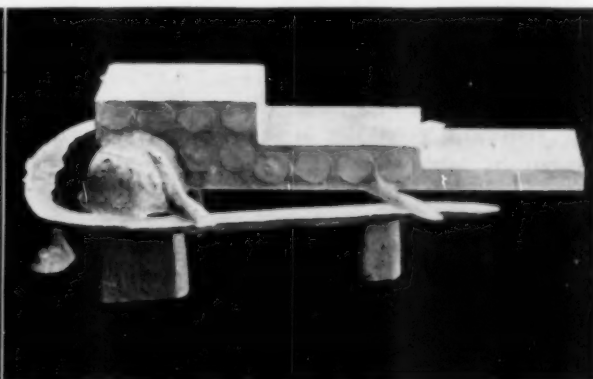


Fig. 10—The step casting with gate and risers attached.

TABLE 5—HIGH TEMPERATURE PROPERTIES OF CAST TEST BARS OF AN ALLOY CONTAINING 1.06% Cu, 0.16% Fe, 0.08% Si, 6.89% Zn, 0.13% Ti, 0.26% Mg AND 0.27% Cr.

Temperature	Time at Temperature	Y.S.	T.S.	% El.	B.H. N.*
Room	6 months	27800	38400	5.5	82
200°F.	1/2 hour		32100	10.5	66
200°F.	3 days		33450	8.5	71
200°F.	10 days		35800	5.0	79
200°F.	25 days		37900	4.5	86
200°F.	50 days		37800	3.5	86
200°F.	100 days		36400	3.5	91
300°F.	1/2 hour		25800	9.5	58
300°F.	4 days		29100	6.5	78
300°F.	9 days		27000	8.0	68
300°F.	25 days		23700	8.0	62
300°F.	50 days		22200	9.5	59
300°F.	100 days		20700	8.0	49
400°F.	1/2 hour		19700	10.5	56
400°F.	2 days		14570	13.5	49
400°F.	5 days		13650	16.5	48
400°F.	10 days		12500	18.5	45
400°F.	25 days		12250	18.0	45
400°F.	50 days		11800	15.5	41
500°F.	1/2 hour		11300	23.0	49
500°F.	3 days		8700	33.0	43
500°F.	5 days		8300	20.0	43
500°F.	10 days		8550	30.0	41
500°F.	25 days		8050	28.0	43
500°F.	50 days		7800	33.0	40
600°F.	1/2 hour		6400	29.0	42
600°F.	1 day		6000	41.0	40
600°F.	5 days		5500	32.0	41
600°F.	10 days		5500	38.0	40
600°F.	15 days		5400	39.0	39
600°F.	25 days		5500	39.0	42

\*Brinell Hardness tests were made at room temperature, after the high temperature treatment indicated in the table.

properties in separately cast test bars poured in green sand and aged 30 days at 85° F.

Yield strength, psi. — 21,000

Tensile strength, psi. — 36,000

Elongation, per cent in 2 in. — 10

Brinell hardness — 66 to 70

Endurance limit, psi.\* — 7500

Charpy impact value, ft.-lb.\*\* — 3.5

Solidification range — 652° C (1266° F) to 610° C (1130° F)

Specific gravity — 2.81

Electrical conductivity, per cent I.A.C.S. — 29.5

Inasmuch as iron and silicon concentrations of 0.15 per cent and 0.08 per cent, respectively, probably cannot be maintained in ordinary foundry practice with aluminum of the purity at present generally available, it is to be expected that minimum specification values for mechanical properties of this type of alloy must be considerably lower than those given in

\* R. R. Moore rotating beam type of machine, 500,000,000 cycles.

\*\* Modified Charpy impact machine, 10 mm x 10 mm keyhole type, drilled and sawed, notched specimens, section back of the notch 5 mm x 10 mm, 5.07-lb. hammer.

the foregoing. The highest-purity aluminum-casting alloys at present in commercial use are produced to maximum silicon concentrations of about 0.25 per cent. Under similar conditions, it is believed that this type of alloy could be produced to minimum tensile specifications of 30,000-psi. tensile strength and 6 per cent elongation. The properties of the alloy are much less sensitive to iron concentration, and a maximum somewhere between 0.5 per cent and 0.75 per cent probably will be found permissible.

#### MICROSTRUCTURE

The microstructure of the aluminum-zinc-magnesium-copper alloys are illustrated by the photomicrographs (Fig. 8, 8b, 8c, and 8d). The compositions of the alloys photographed are given in the captions to these illustrations. Figures 8a and 8b show similar specimens cut from the cope side of 4-in. sections of the step casting illustrated by Fig. 10. The alloy shown in Fig. 8a contained 1.05 per cent copper, while that in Fig. 8b contained 0.34 per cent copper. In Fig. 8a the dark areas brought out by Keller's etch<sup>13</sup> are rich in copper and they also contain light particles of CuAl<sub>2</sub> precipitate. The lower copper alloy shown by Fig. 8b does not have a noticeable amount of copper segregation. Such structures are usually accompanied by superior tensile properties in heavy sections or in castings otherwise subjected to abnormally slow solidification. This type of structure also appears more resistant to corrosion than one exhibiting particles of copper constituent. There is a considerable difference in the grain size between the two specimens of Fig. 8a and 8b, probably due in part to the higher titanium in the finer-grained specimen and in part to the inevitable variations in the structure of castings. However, the specimens illustrated are fairly typical of the effects of the amount of copper content on the microstructure in heavy sections. In chilled sections or light sections where solidification is more rapid, this type of copper segregation is less pronounced.

In general, the high-purity alloys of the composition photographed consist essentially of a solid solution which is subject to precipitation hardening at low temperatures. Only a very small amount of visible undissolved microconstituents occur. The principal microconstituents which form visible particles in alloys of the composition photographed are the Al-Fe-Si

TABLE 6—EFFECT OF EXPOSURE TO 300° F AND 400° F ON ROOM TEMPERATURE TENSILE PROPERTIES OF CAST TEST BARS OF AN ALLOY CONTAINING 1.06% Cu, 0.16% Fe, 0.08% Si, 6.89% Zn, 0.13% Ti, 0.26% Mg AND 0.27% Cr.

Treatment	Properties			
	Y.S.	T.S.	% El.	B.H.N.
6 Mos. at R.T. ....	27800	38400	5.5	82
2 Mos. at R.T.+100 days at 300°F.+2 Mos.* at R.T.**	20200	28050	5.7	60
2 Mos. at R.T.+ 50 days at 400°F.+3 Mos.* at R.T.**	13700	26600	7.7	54

TABLE 7—EFFECT OF EXPOSURE TO BRAZING TEMPERATURES AND REAGING AT ROOM TEMPERATURE. TESTS MADE ON CAST TEST BARS CONTAINING

0.38% Cu, 0.17% Fe, 0.08% Si, 6.87% Zn, 0.11% Ti, 0.26% Mg, 0.23% Cr.

Treatment	Y.S.	T.S.	% EL.	B.H.N.
31 days at 85°F.	20300	35600	10.5	66
26 days at 85°F., 2 hours at 900°F., + 30 days at 85°F.	19500	33600	9.0	64
26 days at 85°F., 2 hours at 950°F., + 30 days at 85°F.	19500	33450	9.1	63
26 days at 85°F., 2 hours at 1000°F., + 30 days at 85°F.	19500	32700	8.0	67
26 days at 85°F., 2 hours at 1050°F., + 30 days at 85°F.	19600	34650	11.0	68
26 days at 85°F., 2 hours at 1100°F., + 30 days at 85°F.	19900	35800	12.9	67
26 days at 85°F., 2 hours at 1120°F., + 30 days at 85°F.	19300	35200	13.1	66

constituents which usually occurs at the grain boundary, but it may not occur in the typical "Chinese script" form, probably because of its small amount. A very small amount of  $\text{CuAl}_2$  particles occur within the grain or at the grain boundaries where the final solidification took place. Some  $\text{Mg}_2\text{Si}$  particles, recognized under the microscope by their bluish color, occur as isolated particles or in conjunction with the other constituents. The constituents in the alloy have been identified by the methods outlined by E. H. Dix, Jr. and F. Keller<sup>13</sup>.

#### FOUNDRY CHARACTERISTICS

Although the foundry experience obtained on an aluminum-zinc-magnesium-copper alloy having the preferred composition mentioned above is not extensive, some estimate of their foundry characteristics can be made.

It has been well established that the tensile properties of test bars or of bars machined from 1-in. sections are not affected by pouring temperatures between 1300° and 1450° F. When the pouring temperature is lowered to 1250° F or raised to 1500° F, a very slight decrease in tensile properties occurs.

The fluidity at 1350° and 1450° F has been determined in the manner formerly described<sup>12</sup> and found to be somewhat inferior to many aluminum alloys now in use. However, this difficulty can be offset by employing a slightly higher pouring temperature since no adverse effect is encountered by this procedure.

Data on tensile properties in heavy sections have already been presented, and it was previously noted that a high percentage of test-bar properties is obtained.

The alloy must be well risered to prevent shrinkage, but in this respect it does not differ from some alloys now in commercial use.

The foundry characteristic which probably would cause the most trouble is hot-shortness. In this respect it is about as subject to hot-cracking as some of the aluminum-copper alloys now in use. Therefore, very intricate types of castings might be expected to be difficult to produce in this alloy.

#### WELDING AND BRAZING CHARACTERISTICS

It has been pointed out that the alloy having the preferred composition is not adversely affected by heating to a temperature near the melting point if it

is allowed to re-age subsequently. Furthermore, the high-tensile properties of this alloy are attained without heat treatment so that, in consequence, it can be welded as readily as the other as-cast alloys and will still retain its high-tensile properties. The alloy also is readily furnace-brazed, since a brazing temperature up to 1100° F can be used. Accordingly, this aluminum-zinc-magnesium-copper type of alloy presents the possibility of utilizing welded and brazed assemblies of castings having exceptionally high strength, toughness, and resistance to corrosion.

#### Summary

The tensile properties of aluminum-zinc-magnesium-copper alloys have been determined over a range of 0 to 1.75 per cent copper, 3 to 13 per cent zinc, and 0 to 1.0 per cent magnesium. An alloy containing 0.4 per cent copper, 0.15 per cent iron, 0.08 per cent silicon, 6.6 per cent zinc, 0.33 per cent magnesium, 0.25 per cent chromium, and 0.15 per cent titanium appears to have the maximum combination of strength, ductility, and resistance to corrosion and has been investigated in greater detail.

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## DISCUSSION

Chairman: HIRAM BROWN, Solar Aircraft Co., Des Moines, Iowa

Co-Chairman: H. R. YOUNGKRANTZ, Apex Smelting Co., Chicago

E. H. HOLZWORTH (Written Discussion):<sup>1</sup> This paper presented by L. W. Eastwood seems to stress copper as a beneficial element in castings of this alloy group. This may be because there is no comparison with copper-free alloy of the same composition. In several thousand tests, we have found that copper has no beneficial effects and tends to embrittle the alloy. Copper cannot be considered by itself, but rather in conjunction with other impurities such as silicon and manganese. We have found that the lower the silicon, manganese and copper content, the higher the elongation and yield strength, and copper does not increase the tensile strength. We have found that copper plus silicon plus manganese should not be over 0.75 per cent maximum.

The proper ratio of zinc to magnesium may be changed so as to allow a small per cent of copper to be substituted for magnesium, but we have found no beneficial effects from this procedure. We find that by proper control of the zinc magnesium ratio, higher combination of physical properties are obtained than can be obtained by the substitution of copper for part of the magnesium.

We first allowed 0.4 per cent copper as a maximum and later found it advisable to reduce this allowance to 0.3 per cent maximum, preferring the alloy as near copper-free as possible.

Some examples are shown below. In this group we prefer the analysis as shown in example 16-52T where copper, silicon and manganese are low.

Heat No.	T.S., psi	Y.S., psi	% Zn	Mg	Cr	Ti	Cu	Si	Fe
			Elong.						
C141	42,000	32,800	3	11.84	0.39	0.41	0.20	0.10	0.22 0.28
C143	39,000	28,800	4	6.44	0.71	0.45	0.21	0.12	0.25 0.29
412	37,200	28,000	3.5	5.10	0.59	0.57	0.20	0.18	0.08 0.22
16-52T	39,000	26,000	8	5.35	0.54	0.53	0.15	0.08	0.10 0.22
118-16	36,900	26,600	5	5.33	0.56	0.55	0.19	0.08	0.13 0.26
S 2	38,000	26,400	5.5	7.84	0.39	0.50	0.18	0.12	0.14 0.41
S 4	37,000	28,600	5	4.38	0.77	0.51	0.17	0.12	0.14 0.41

The above heats, some made for experimental work and some made in regular practice by different foundries, compare with properties shown in the presented paper.

Comparing Fig. 1a, 1b, 1c, 1d with Fig. 2a, 2b, 2c, 2d using the 6 per cent zinc graph, tensile strength is approximately the same, yield strength is higher with larger percentage of copper, and elongation decreases with the increase in copper. Figure 4 illustrates how the increase copper content limits the scope of the maximum physical properties.

On corrosion tests we show the following results.

	T.S., psi	Y.S., psi	Elong., %
Start of test	37,500	23,600	10.2
After 6 months in salt spray	36,500	27,400	4.3

Analyses were as follows:

Zn—5.48; Mg—0.60; Fe—0.58; Ti—0.28; Cr—0.53

Elevated temperature tests were run by the University of Michigan and were for 1000 hr at indicated temperatures:

Temp. °F	T.S., psi	Y.S., psi 2% Offset	Elong., %
85	34,000	29,000	3.5
175	33,800	29,000	3
250	29,250	26,700	3.5
325	19,700	17,000	7
400	13,225	10,900	15

HIRAM BROWN (Written Discussion):<sup>2</sup> First of all, I should like to congratulate the authors on an excellent paper which did much to bring to prominence the very useful aluminum-zinc-magnesium alloys.

<sup>1</sup> Frontier Bronze Corp., Niagara Falls, N. Y.

<sup>2</sup> Solar Aircraft Co., Des Moines, Iowa

I should like to submit a few comments to supplement the data in the paper. The paper states that chromium was added to improve corrosion resistance. Both Walter Borsack and myself in previous papers have pointed out that chromium in alloys of this type also improves strength. In tests performed on alloys of identical composition, except for chromium content, it was found that when 0.4 per cent chromium was added to a chromium-free heat of the following composition:

Zn	Mg	Cr	Fe	Cu	Si	Mn	Ti
5.53	0.57	—	0.37	.08	0.17	—	—

yield strength increased 12.3 per cent, tensile increased 18.3 per cent, and elongation increased 42 per cent.

Subsequent addition of titanium to this heat did not affect physical properties but refined the grain and reduced somewhat the tendency of the alloy to hot crack.

The presence of lead in exceedingly small amounts is very detrimental to physical properties.

Calcium in small amounts has bad effects upon both the physical properties and the casting properties.

It is interesting to point out that in Fig. 1a, 1b, 1c, and 1d alloys containing 8 per cent zinc and 0.4 per cent magnesium and 5 per cent zinc and 0.5 per cent magnesium respectively both have high strength. However, the tensile strength elongation ratio indicates that the proportional ductility is much higher in the lower zinc higher magnesium alloy. By dividing the per cent elongation by the psi tensile shown on the curves mentioned above it is found that for each 1000 psi tensile the 8 per cent zinc, 0.4 per cent magnesium alloy has 0.175 per cent elongation while for each 1,000 psi tensile the 5 per cent zinc, 0.5 per cent magnesium alloy has 0.324 per cent elongation, or 80 per cent greater. Where ductility is a factor, such as in applications involving shock, it might be better to use the lower zinc, higher magnesium alloys.

It is interesting to note that the properties of alloys containing various ratios of zinc and magnesium are so predictable that curves can be drawn which show the ratios of zinc and magnesium necessary to maintain uniform physical properties over a wide range of composition.

The authors may be interested to know that other curves of more linear nature have been drawn to show this relation. One curve was boiled down to an equation several years ago but has only recently been published. The equation is  $Y=KA^x$  where  $Y$ =% magnesium,  $x$ =% zinc,  $A=0.85$  and  $K=1.7$ . If this equation is applied to zinc and magnesium in the ranges of 1.5 to 14 per cent zinc and 0.15 and 1.7 per cent magnesium, the resulting alloys will give in excess of 32,000 psi tensile and 3 per cent elongation with commercial impurities present. If values from this equation are plotted to form a curve, then two additional curves spaced from this main curve by  $\pm 15$  per cent of the zinc values or  $\pm 15$  per cent of the magnesium values will give limits of composition which will give minimum values of 32,000 psi tensile and 3 per cent elongation. These figures are based on copper content of 0.4 per cent or less.

R. A. QUADT (Written Discussion):<sup>3</sup> For the past three years the Federated Metals Division of the American Smelting & Refining Company has been marketing a high-strength as-cast aluminum alloy called "Tenzaloy". This is an aluminum alloy containing nominally 8 per cent zinc, 0.8 per cent copper, 0.35 per cent magnesium. It also contains about 0.1 per cent titanium for grain refinement and usually contains 0.1 per cent chromium.

This alloy has the same self-aging characteristics of the alloy described in the present paper. In arriving at the above composition the various investigations on the aluminum-zinc-magnesium-copper system produced results with regard to mechanical properties that check exactly with those presented by the authors.

The authors are to be complimented for the excellent data presented that prove the mechanical properties that are to be expected from this class of alloy under the various conditions indicated. However, it is most unfortunate that, with regard to corrosion, the paper contains a number of very definite statements, but a disconcerting lack of data are presented to substantiate them. This is particularly unfortunate in the case of these alloys since by tradition and reputation these alloys

<sup>3</sup> American Smelting & Refining Co., Barber, N. J.

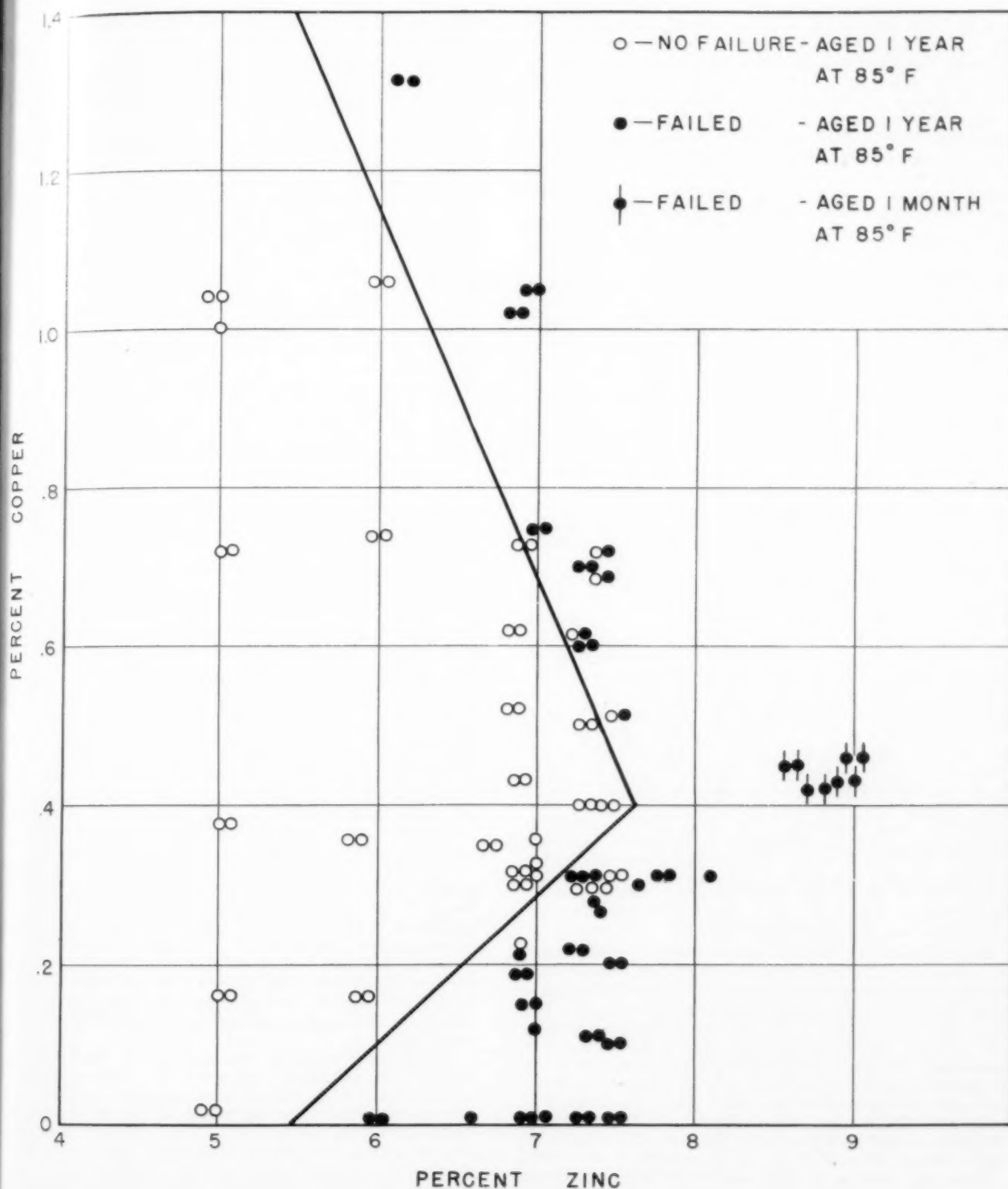


Fig. 11. Effect of copper and zinc contents on the resistance to stress corrosion in NaCl-hydrogen peroxide 14-day test. Alloys contain approximately 0.15% Fe, 0.08% Si, 0.15% Ti, 0.0 or 0.25% Cr, the copper and zinc contents as indicated, and with magnesium contents controlled by the optimum zinc to magnesium ratios.

(Note: Chromium in the range indicated had no measurable effect on the resistance to stress corrosion in these tests.)

have a very unsavory past. There is very little data on the subject although in the past few years there have been some indications that the bad reputation that zinc holds in aluminum alloys is not entirely justified. Very likely the quantities

of other elements are of more significance than the zinc alone.

My only questions with regard to this paper concern corrosion. I would like very much to have answers and data on the following:

1. How significant is the effect of 0.25 per cent chromium on the corrosion resistance of the alloy. Is the improvement one of appearance, weight loss, or improvement in residual mechanical properties, and is this improvement persistent over an extended corrosion period such as several years or, as is often the case, after a long period of exposure does it become impossible to differentiate between the two types of alloys?

2. With regard to the effect of 1 per cent copper as compared to 0.35 per cent copper, again is the difference with re-

gard to initial appearance, weight loss, mechanical properties, and is there a significant difference over an extended period of time. The statement mentioned by the authors is not a definite one and it is important to determine whether or not the difference is of a significant order of magnitude.

My next questions have to do with the stress-corrosion characteristics. In view of the statements made by the authors, would an alloy of their preferred composition but with 8 per cent zinc, stress corrode at 75 per cent of the yield strength? Under the same conditions, would the preferred alloy, except containing no chromium, stress corrode? Also would an alloy of their preferred composition but containing no copper stress corrode when loaded at 75 per cent of the yield strength in their corroding medium? Also would an alloy of their preferred composition but containing 0.8 per cent copper stress corrode if loaded at 75 per cent of the yield strength?

Finally, I would like to offer a few comments with regard to the effect of copper and chromium on the foundry characteristics of the alloy. It is very useful to specify a small quantity of copper, under 1 per cent, in order to permit reduction of the quantity of magnesium that is required to produce the desired mechanical properties after room temperature aging. At the authors show, the higher the copper, the lower the required magnesium to produce equivalent properties. Since there is undoubtedly an upper limit with regard to copper from the corrosion point of view, a small amount of copper appears desirable. As any foundry metallurgist knows who has worked with alloys of this type, the smaller the amount of magnesium in the casting alloys, the fewer the problems with regard to dross inclusions.

Also as any foundryman knows who has attempted to use casting alloys with 0.25 per cent chromium or more, it is difficult to get the chromium into solution and usually stirring has to be employed. In alloys of this type agitation of any kind, and especially that produced by stirring, is likely to produce unhappy results with regards to dross inclusions. Chromium segregation is quite pronounced when it is as high as 0.25 per cent and from the foundry point of view it would be desirable to have the chromium lower than that and perhaps only require as much as can be readily kept in solution without resorting to stirring. In our tests this appears to be in the neighborhood of 0.1 per cent or perhaps 0.15 per cent. It is for this reason that chromium should probably only be specified as a maximum in this alloy since it apparently does no harm, and before it should be required as an important alloying ingredient it should definitely be proved that it has a significant effect on long time corrosion resistance properties.

#### DR. EASTWOOD'S REPLY

DR. EASTWOOD: I do not believe there is anything too controversial in the remarks by Mr. Holzworth, except those on the effect of copper. As I recall it, we actually do get a slight improvement in tensile strength by additions of a few tenths per cent of copper to the optimum composition, namely, one containing about 6.6 per cent zinc and 0.33 per cent magnesium. The main purpose of the copper, however, is not to improve the tensile strength, but to improve its stress-corrosion characteristics in the sodium chloride-hydrogen peroxide test. This test has been described by E. H. Dix, Jr., in his paper entitled "Acceleration of the Rate of Corrosion by High Constant Stresses," *Trans. AIME*, vol. 137, p. 11 (1940).

Using this hydrogen peroxide-sodium chloride stress-corrosion test, we can definitely show an improvement in stress-corrosion characteristics if the copper content is about 0.4 per cent. This is illustrated by Fig. 11, showing the effect of copper and zinc contents on the resistance to stress corrosion in the salt-hydrogen peroxide 14-day test, using specimens stressed at 1600 psi. This figure includes recent data obtained by the Aluminum Research Laboratories. Copper is not added for improvement in tensile properties but, fortunately, at 0.4 per cent copper the tensile properties are not harmed by the copper addition.

With reference to Fig. 11, a few points on the graph at about 9 per cent zinc represent specimens that received less aging than the others. These specimens failed and would have failed also if they had aged as long as the other specimens. Increased aging, at least up to 1 year at 85 F, causes alloys of

this type to become more susceptible to stress corrosion.

Test results on alloys both with and without a 0.25 per cent chromium addition were included in Fig. 11. This procedure was employed because comparable resistance to stress corrosion was obtained on like alloys regardless of whether this amount of chromium was or was not present.

D. BASCH: This 40E alloy as it is commercially known at the present time, was originally based on a patent which the General Electric took out under the name of Fuller and Basch in 1927, the first of a series of aluminum-zinc-magnesium cast alloys now on the market. It had exactly the same composition as the alloy that has been described in the authors' paper except that it did not have chrome and that no definite additions of titanium were proposed. When Frontier Bronze Corp., just before World War II, came out with their alloy, adding to the G. E. composition chrome and titanium, we at General Electric realized that they offered some advantages and we made arrangements with them whereby we dropped all possible patent entanglements in return for shop rights and joined in a further prosecution of the alloy. By the time that Dr. Eastwood and Mr. Kempf had published their paper, General Electric had already done a great deal of development work and if Dr. Eastwood will remember, when the seal of secrecy was placed on the Eastwood-Kempf paper during the war by the National Advisory Board, we (General Electric) advised the Board that we had already obtained all the data before the issuance of the Eastwood-Kempf paper by our own independent research and the National Board never disputed this statement.

I want to bring that out to point out the alloy is of considerably greater age than it would appear from this paper. Of course, a great deal of work has been done since that paper was prepared. For instance, it has been found that we can now get aluminum ingots with better purity than 0.25 per cent silicon, which in that paper was pointed out as the optimum obtainable commercially at that time. The zinc-magnesium ratio in its effect on properties has been studied and optimum values have been established. The remarks that Mr. Holzworth made about the addition of copper I can fully subscribe to because we have also investigated the stress corrosion of that alloy. It has been found that as long as zinc plus magnesium does not exceed 6 per cent, and with the addition of chrome, stress corrosion is practically eliminated. That has been corroborated by investigations of German scientists of which you will find many references in the literature.

One thing that I missed in the presentation of the paper is some of the commercial features of that alloy. The Frontier Bronze Corp. who are handling that alloy at the present time, since General Electric only uses the alloy in its own plant and does not sell it to anybody, have actually produced well over 15 million pounds of castings of that alloy since its inception. An alloy of that sort, with its definite disadvantages of poorer fluidity and castability and its insistence on high purity ingots, must have some definite advantages to warrant such a high production.

I think foundrymen are interested in why they should use the alloy. In addition to high corrosion resistance, ability to regain original high strength after exposure to welding and brazing temperature through simple air aging, advantageous realization of test bar strength in heavy sections, three other distinct advantages should be cited. First, the elimination of the high temperature quench necessary for heat treated high strength aluminum alloys, aside from the commercial advantage of saving time, equipment, space and control required for solution treatment, obviates the danger of warping and distortion especially in spidery castings and castings where section thicknesses vary. Secondly, 40E alloy is outstanding in its machinability. In that respect it compares with the high magnesium content aluminum alloys. Third, in shock resistance it is again outstanding and that is one of the reasons the Navy Department in many instances had castings made out of 40E alloy. I am not referring to shock resistance as evaluated by a Charpy test although even there it is better than alloy 195, 356 or 355. A Charpy test usually runs concurrent with elongation, and it simply determines the foot-pounds at which a certain sample breaks, but it does not bring out the fact

\* Almin Ltd. of Great Britain, Schenectady, N. Y.



of deformation, especially under such applications as the Navy and aircraft are interested in. Deformation of a part before breakage may be worse than actual breakage, because in a final adjusted apparatus the deformation of any part may make it impossible to use the apparatus and may make repair of the apparatus so difficult that it cannot be completed in time for the operation that it is intended.

Now these shock tests that determine the resistance of the alloy to deformation were conducted by actually detonating against certain test samples high explosive mixtures which had a very high and rapid surge of shock forces, such as you would get, for instance, when a mine is exploded under water against a ship, or when a bomb is dropped on the deck of a ship. When I was with the General Electric Co. we were approached at that time by the Navy to investigate that particular phase. Charpy and explosion tests were made and we spent close to \$20,000 in the investigation of that particular feature. The tests showed that the 40E alloy actually is in a class by itself, superior to alloy 195, 356, 43, 108 or 112 alloy; inferior only to alloy 220.

I am not trying to sell you the alloy, because I have no commercial interest in it. I have spent over 20 years handling the alloy, investigating it with funds supplied by the General Electric Co. and afterwards in conjunction with E. H. Holzworth. I really feel that the foundry industry should take cognizance of the alloy and should realize its possible advantages. I have spoken primarily of 40E alloy because that is the composition I am most familiar with, but similar claims can probably be made for other compositions of the aluminum-zinc-magnesium series brought out during recent years.

By no means should it be assumed that 40E alloy is a general substitute for commercial alloys. Alloy 40E has certain disadvantages. It does not cast so well as alloy 356, it costs more than other aluminum-silicon alloys in ingot form. So it would be poor commercial and engineering policy to use it when you can meet the requirements of your job with some of the other alloys, but where the specific properties, the specific advantages of 40E alloy are demanded then 40E alloy may be the engineers' and foundrymen's best choice.

SAM TOUR:<sup>6</sup> Figure 4 shows the area in which these properties could be obtained with different copper content, which led to the statement in the paper that 0.4 per cent copper was the optimum. In the paper Fig. 4 shows that for 1.75 per cent copper there is a very small area. For 1 per cent copper it is still larger. Nowhere in the paper do the authors show that the area decreases in going below 0.4 per cent copper. The conclusion, therefore, as I see it is a bit tenuous that 0.4 per cent copper is better than 0.3 per cent or 0.2 per cent. The data presented show improvement in this area relationship as copper goes down.

If you draw lines encompassing those areas in Fig. 4 you will see that they are spreading lines and nowhere have they

reached a reversal point to indicate that 0.4 per cent is significant of anything. It then seems to me that the final conclusion in the paper is that copper is there only for a supposed increase in resistance to stress-corrosion attack. Admittedly it is not there to improve physical properties nor to make this range of composition easier to hit to get a given physical property, but it is there presumably, and so stated, for increased resistance to stress-corrosion cracking.

It is unfortunate that we know so little about the mechanism of stress-corrosion attack. Just why things do stress corrode and why they do not we know very little about. It would certainly be surprising if copper addition to aluminum alloys and to formation of copper-aluminum compound would be an inhibitor of stress-corrosion cracking. Figure 8D shows little rosettes of copper-aluminum compounds. Those rosettes appeared at the grain boundaries. It would certainly be surprising if segregation of copper-aluminum compound at grain boundaries were an inhibitor to stress-corrosion cracking. There are so many other things that affect stress-corrosion cracking, it seems to me it is entirely possible that the apparent improvement in stress-corrosion cracking was inadvertently due to something else and not to the presence of this copper.

#### AUTHOR'S CLOSURE

DR. EASTWOOD (*Written Reply*): Naturally, during the oral presentation of the paper, I did not have time to go into the historical aspects, but I believe we have done it to at least some extent in the paper. I can vouch for everything that Dr. Basch has said. To the best of my knowledge, Ref. 7 in the paper contains the entire published contribution on this subject by Dr. Basch.

I think that the remarks of Mr. Tour have already been answered. Copper at about 0.35 or 0.4 per cent does not have an adverse effect upon the tensile properties, and if I remember correctly, it produced a slight improvement; therefore, the area in Fig. 4 would be no larger at zero per cent copper than it is at 0.4 per cent copper.

Mr. Quadt has raised several questions. Increasing the copper from 0.4 per cent to 1.0 per cent decreased the resistance to salt-spray corrosion. This would result in increased weight loss, poorer appearance and slightly greater loss in tensile properties as a result of the greater degree of corrosion. His questions on stress corrosion are clearly answered by Fig. 11.

Mr. Brown's comments indicate a considerable improvement in tensile properties produced by the chromium addition. We were unable to note this improvement, but this difference in observation may be caused by the lower ingot purity used by Mr. Brown.

The remarks of the various commentators and their interest in this paper are greatly appreciated.

<sup>6</sup> Sam Tour & Co., Inc., New York

# CAUSES OF RAT-TAIL CASTING DEFECT

A.F.S. Committee on Physical Properties of Iron Foundry  
Molding Materials at Elevated Temperatures

## ABSTRACT

The rat-tail phenomena was studied by making flat gray iron castings in a number of different sands. There were indications that pouring temperature of the metal, green strength and moisture content of the sand were not major cause of rat-tails. Since rat-tails occurred on light casting sections, it was inferred that sand does not have to be heated to a temperature higher than 1200 F to cause rat-tails.

Two sand properties appear to be directly correlated to the tendency of a sand to produce rat-tails. These are expansion and hot strength at 1000 F. The sands which produced the rat-tail defect have higher hot strength and higher expansion than the good molding sands.

The report indicates the effect of various materials upon hot strength and expansion.

## Introduction

THIS REPORT describes work carried out by the committee in the foundry of the University of Michigan from April 7 to April 11, 1947, and in the sand research laboratory of the Harry W. Dietert Co. from April 7 to August 1, 1947, with the object of determining the physical properties of molding sands which will control rat-tail casting defects. Previous foundry tests enabled us to select a sand composition which produced severe rat-tail casting defects. These rat-tails could not be eliminated by changing pouring time, gating or pouring temperature. Thus, when they were eliminated, the pattern or metal conditions could not have affected the phenomena.

## Description of Rat-Tail Defect

The rat-tail defect is usually found on the cope or the drag. The defect may be of small magnitude so that only a faint line on the surface is visible as shown in the upper view of Fig. 1. This defect may also be of considerable magnitude so that a definite objectionable break in the casting surface is apparent as shown in the lower view of Fig. 1.

## Preliminary Work

In order to study the rat-tail defect, it was considered necessary to find or develop a sand that would produce rat-tails consistently. A number of sands were tried out by making flat castings. A particular mixture was found to produce rat-tails on every test casting at will. This sand was composed of 100-mesh

sand bonded with clay and western bentonite. It is described as sand mixture No. 1 in the following report.

## Equipment

Figure 2 shows the test pattern used in these experiments. This pattern was molded in a snap flask 14 in. x 16 in. with a 4-in. drag and 6-in. cope.

## Materials

Previous experience had indicated that a 1-in. facing was sufficient to give the same effect as a complete

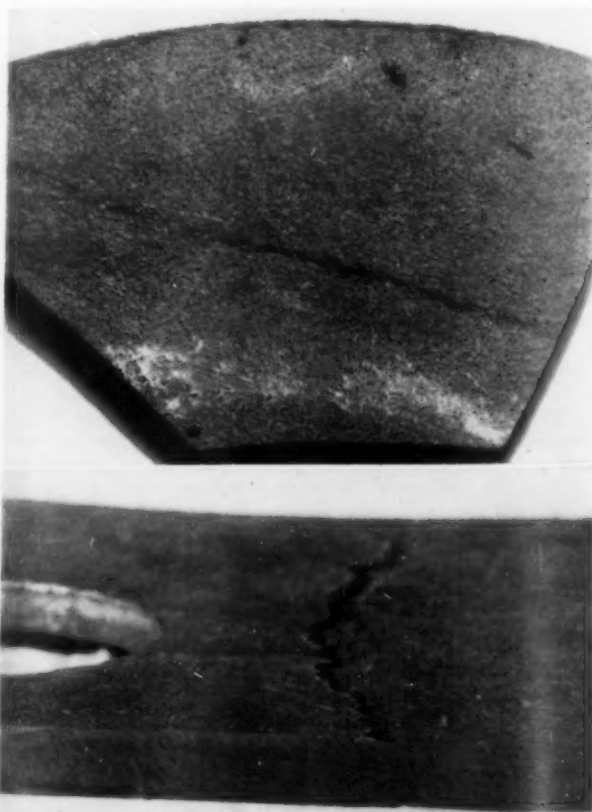


Fig. 1—Examples of a Mild and a Severe Rat-Tail Defect.

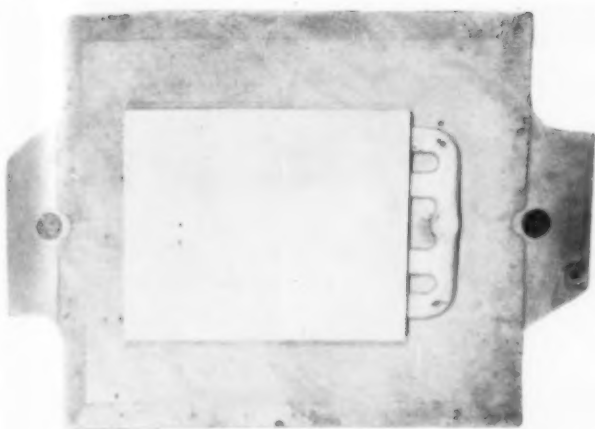


Fig. 2—Pattern Equipment Used in This Investigation.

flask of the facing material; therefore, the pattern proper was covered to a thickness of substantially 1 in. with the sands being studied. The remainder of the flask was filled with heap sand, the properties of which were essentially constant. The properties of sands being studied and the heap sand are given in the body of the report.

#### Test Methods

For most of the experiments, three molds were made for each test condition. As far as possible, each of the following conditions were changed one at a time:

Test A. Thickness of Casting	Fig. 3
B. Pouring Temperature	Fig. 4
C. Pouring Time	Fig. 5
D. Mold Hardness	Figs. 6 and 7
E. Compressive Strength	Fig. 8
F. Moisture	Fig. 9
G. Grain Fineness and Distribution	Figs. 10 and 11
H. Cereal Addition	Fig. 12
I. Combustible Materials	Fig. 13
J. Bond Type	Fig. 14
K. Natural Sands	Fig. 15

#### Molding Program

##### Facing Sand Mixture No. 1

90.5 lb. 100 Mesh Silica Sand  
 8.75 lb. Clay  
 .75 lb. Western Bentonite  
 Mull dry 3 min.  
 Add  
 3.5 lb. Water  
 Mull 5 min.

- A. Effect of Casting Thickness.  
Using mixture No. 1, make three molds from each test pattern. Test patterns are  $\frac{3}{16}$ ,  $\frac{5}{16}$ ,  $\frac{7}{16}$  and  $\frac{9}{16}$  in. thick.
- B. Effect of Pouring Temperature.  
Pour three molds with metal at a temperature of 2500, 2600, 2700 and 2800 F.
- C. Effect of Pouring Rate.  
Make three molds using one, two, three and four ingates. Each ingate is  $1 \times \frac{1}{8}$  in. in cross-section.

- D. Effect of Mold Hardness.  
Make three molds at each of the following mold hardnesses: 25, 40, 60 and 80.
- E. Effect of Green Compression Strength.  
Using sand mixture No. 1, adjust the bond addition to reach approximately 4, 8 and 10 psi green compression and make three molds of each.
- F. Effect of Moisture Content.  
Make three molds with No. 1 facing mixture and dry in core oven. Make three molds with No. 1 sand tempered to 2, 3.5 and 4 per cent moisture.
- G. Effect of Grain Size.

Prepare sands that will:

- a. Pass through 30 mesh and be retained on the 40-mesh screen.
- b. Pass through 40 mesh and be retained on the 50-mesh screen.
- c. Pass through 50 mesh and be retained on the 70-mesh screen.

Add to each of these sands 8.75 per cent clay and 0.75 per cent western bentonite. Mix and temper to best workable moisture.

#### Effect of Grain Distribution.

Prepare sands of each grain size. Make up mixtures of grain sizes so that the sand is evenly distributed over 3, 5 and 7 meshes with the average held at 100 mesh. Make one sand with grains evenly distributed over 5 meshes average 70 mesh. To each of these sands add 8.75 per cent clay and 0.75 per cent western bentonite. Mix and temper to best workable moisture. Make one or more molds with each mixture.

- H. Effect of Cereal Binder.  
Starting with sand mixture No. 1, make up mixtures containing 0, 0.5, 1.0 and 1.5 per cent cereal flour. Make three molds with each sand mixture.
- I. Effect of Combustible Materials.  
Add to mixture No. 1 5 per cent seacoal. To another No. 1 mixture add 2 per cent wood flour. Make three castings with each mixture.
- J. Effect of Clay Bond.  
Make three mixtures using 100-mesh silica sand. To the first add 10 per cent fireclay, to the second add 4 per cent western bentonite. Add 4 per cent southern bentonite to the third mixture. Temper each mixture to optimum moisture content and make three castings with each sand mixture.
- K. Effect of Some Natural Molding Sands.  
Secure representative natural molding sands from the states of Ohio, New York and New Jersey. Make castings in these sands. Also include used molding sand that has been maintained by additions of Ohio sand.

#### Molding Procedure

Every step in the various operations required in the foundry was according to routine and every task was performed in accordance with written instructions. Molding procedure outline is given below.

1. Brush match plate clean.
2. Place cope flask on table of jolt-squeeze machine.
3. Place match plate on cope flask with drag face up.
4. Place drag flask on match plate pins down. Red corners of flask together.



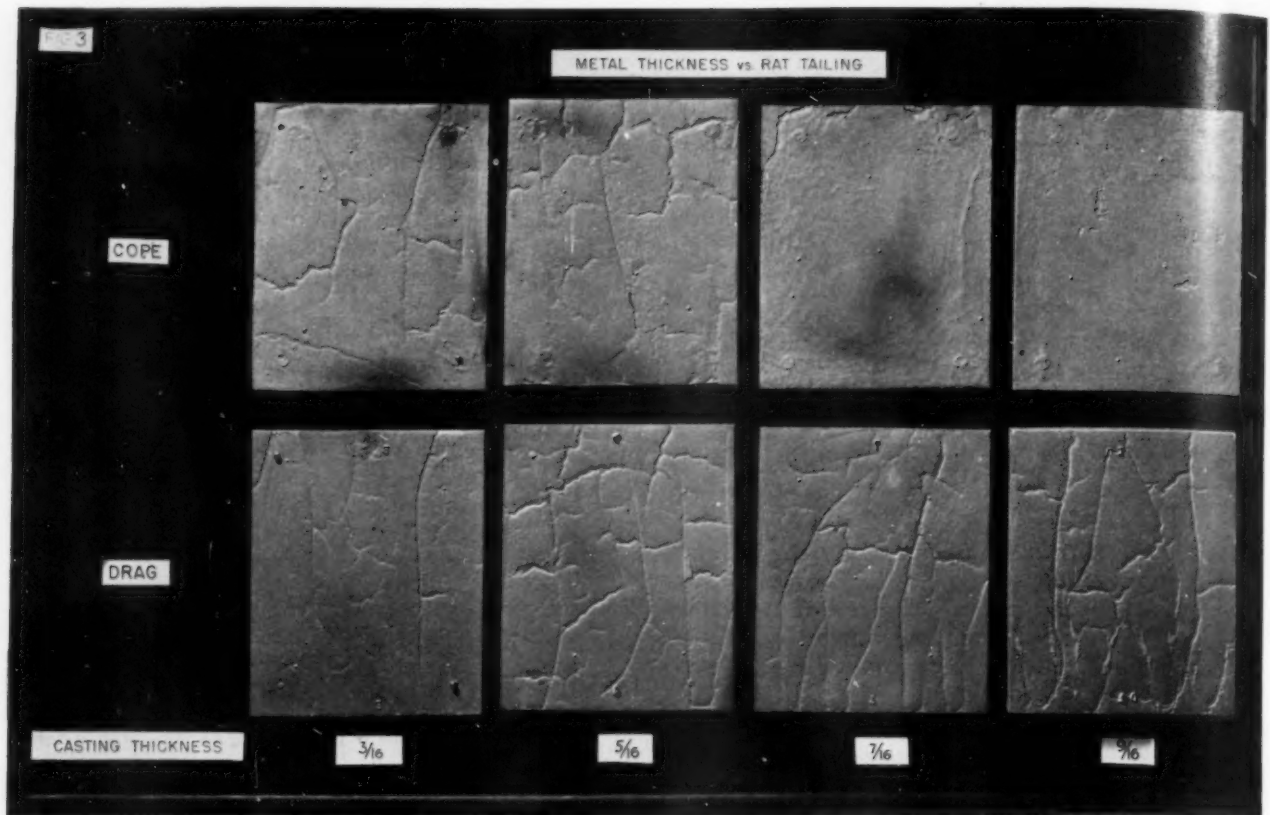


Fig. 3—Cast specimens illustrating effect of metal thickness on rat tailing.

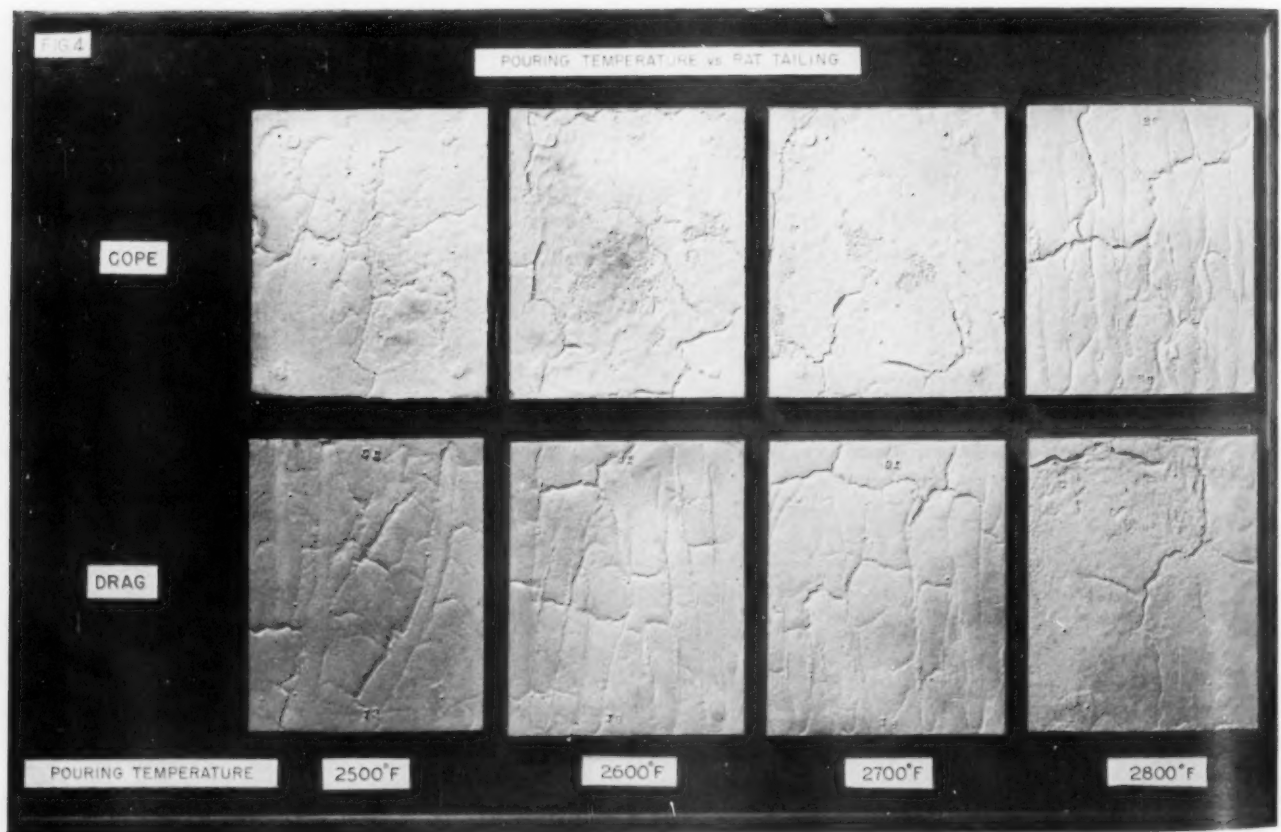


Fig. 4—Cast specimens illustrating effect of pouring temperatures on rat tailing.



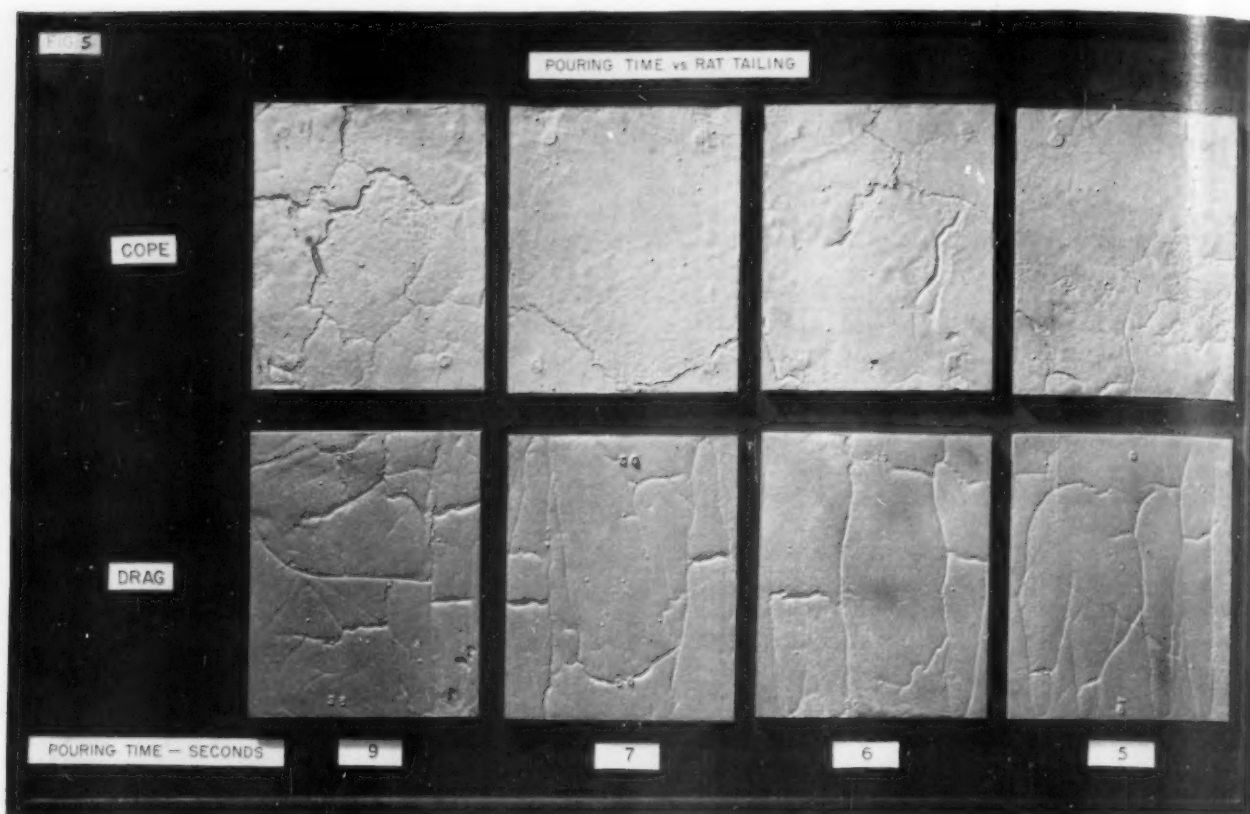


Fig. 5—Cast specimens illustrating effect of pouring time on rat tailing.

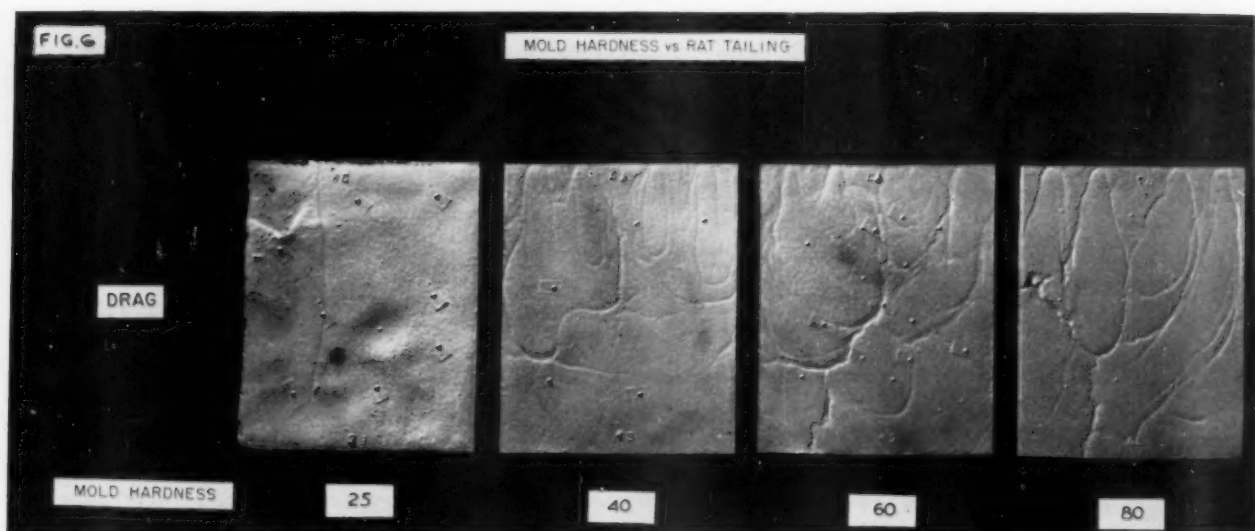


Fig. 6—Cast specimens illustrating effect of mold hardness on rat tailing.

#### Observations:

Rat-tails were observed on all of the castings as is shown in Fig. 4. No appreciable difference in the various test castings was observed.

#### Conclusions:

For the type of sand selected and the conditions under which these tests were made, the pouring temperature need not be held precisely at a chosen

temperature. However, for uniformity, a pouring temperature of 2650 F was selected for all the succeeding tests.

#### Test C—Effect of Pouring Rate

In this test, the pouring rate was changed by using different numbers of ingates to determine whether the severe rat-tailing could be eliminated by pouring rate.

Facing sand mixture No. 1 as described in Table 2 was used in this series of tests.



## Procedure:

Three molds were cast using one, two, three and four ingates  $1 \times \frac{1}{8}$  in. in cross-section.

## Observations:

All castings in this test showed rat-tails (Fig. 5). The flow of metal was usually outlined by rat-tails, particularly when one ingate was used.

## Conclusions:

The rat-tail casting defect was shown to be increased by uneven heating of the mold face; for example, by causing the metal flow to be restricted over a narrow area of the mold surface.

For the succeeding tests the gate arrangement with four ingates was used as representative of good foundry practice. The pattern was gated on the short end and not on the long side. This arrangement of gating made for a more severe test of the sand.

## Test D—Effect of Mold Hardness

The basic facing sand mixture No. 1, as described in Table 2, was used in this test.

## Procedure:

The mold hardnesses indicated below were obtained by varying the number of jolts and squeeze pressure during operations.

Mold Number	Mold Hardness Drag
37	30
38	23
39	25
40	42
41	40
42	40
43	60
44	57
45	57
7	80
8	80
9	80

## Observations:

The castings made in the molds which were rammed to a hardness of less than 60 had few or no rat-tails. Castings made in molds which were rammed to a mold hardness 60 or higher had numerous rat-tails. Figure 6 shows representative casting surfaces obtained with the different mold hardnesses.

## Conclusions:

Rat-tails were largely eliminated by ramming molds to a hardness below 60 which produced low mold strength in terms of psi. The mold hardness for all molds in succeeding tests were held within a mold hardness range of 75 to 80 so that rat-tails would not be eliminated by a mold hardness condition.

## Effect of Mold Hardness on Hot Compressive Strength

In order to determine the effect of mold hardness on properties at elevated temperatures, sand test specimens were rammed to various mold hardnesses and the strength at elevated temperatures measured. The results are shown in Table 3 and Fig. 7.

TABLE 3—EFFECT OF MOLD HARDNESS ON HOT COMPRESSIVE STRENGTH

Mold Hardness Temp., ° F.	Few or No Rat-Tails					Rat-Tails		
	20	30	40	50	60	70	80	85
	Hot Compressive Strength, psi.							
500	35	45	57	67	75	85	105	127
1000	16	22	28	38	55	103	175	215
1500	17	22	30	42	63	115	168	200
2000	10	20	25	43	60	104	220	280
2500	16	22	20	38	47	56	77	135
Density, lb/cu.ft.	76	78	81	86	93.5	100	103.5	

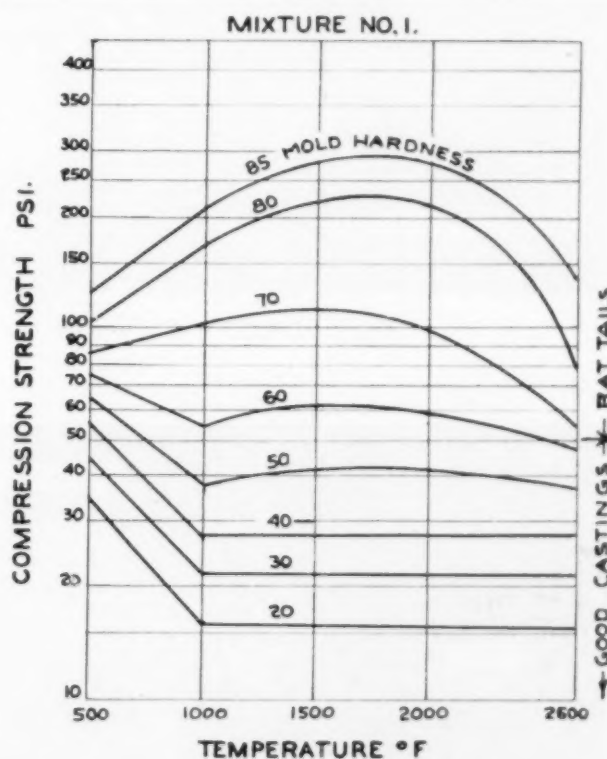


Fig. 7—Effect of Mold Hardness on Hot Strength.

## Test E—Effect of Green Compressive Strength

In order to vary the green compressive strength, three facing sand mixtures were prepared with 6, 10 and 13.3 per cent clay. These mixtures had green strengths of 3.8, 8.1 and 10.15 psi, respectively (Table 4).

## Procedure:

Three molds were made from each type of sand, and poured in the same manner as in previous tests.

## Observations:

In this series of tests, little effect upon the rat-tail phenomena was made by changing green compressive strength in mixtures containing bentonite and clay from 6 to 13 per cent as may be noted in Fig. 8.

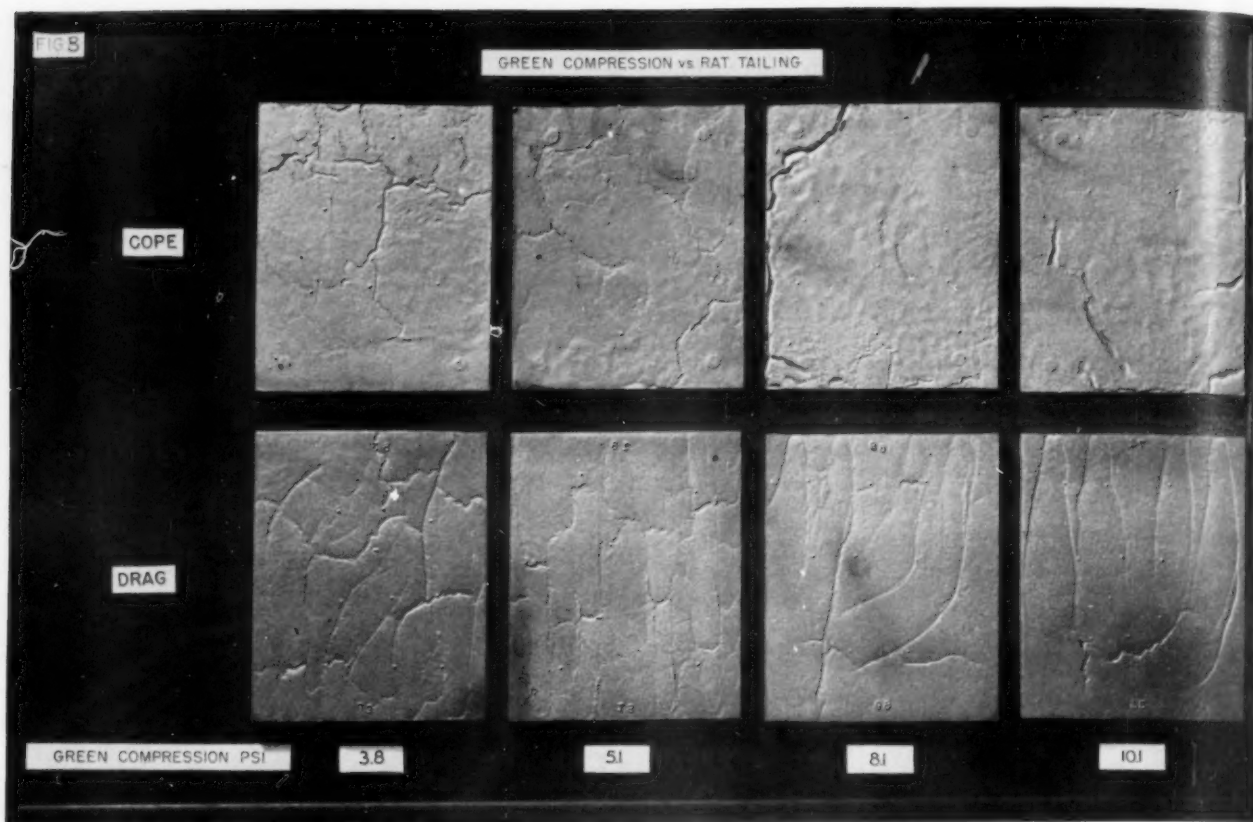


Fig. 8—Cast specimens illustrating green compressive strength on rat tailing.

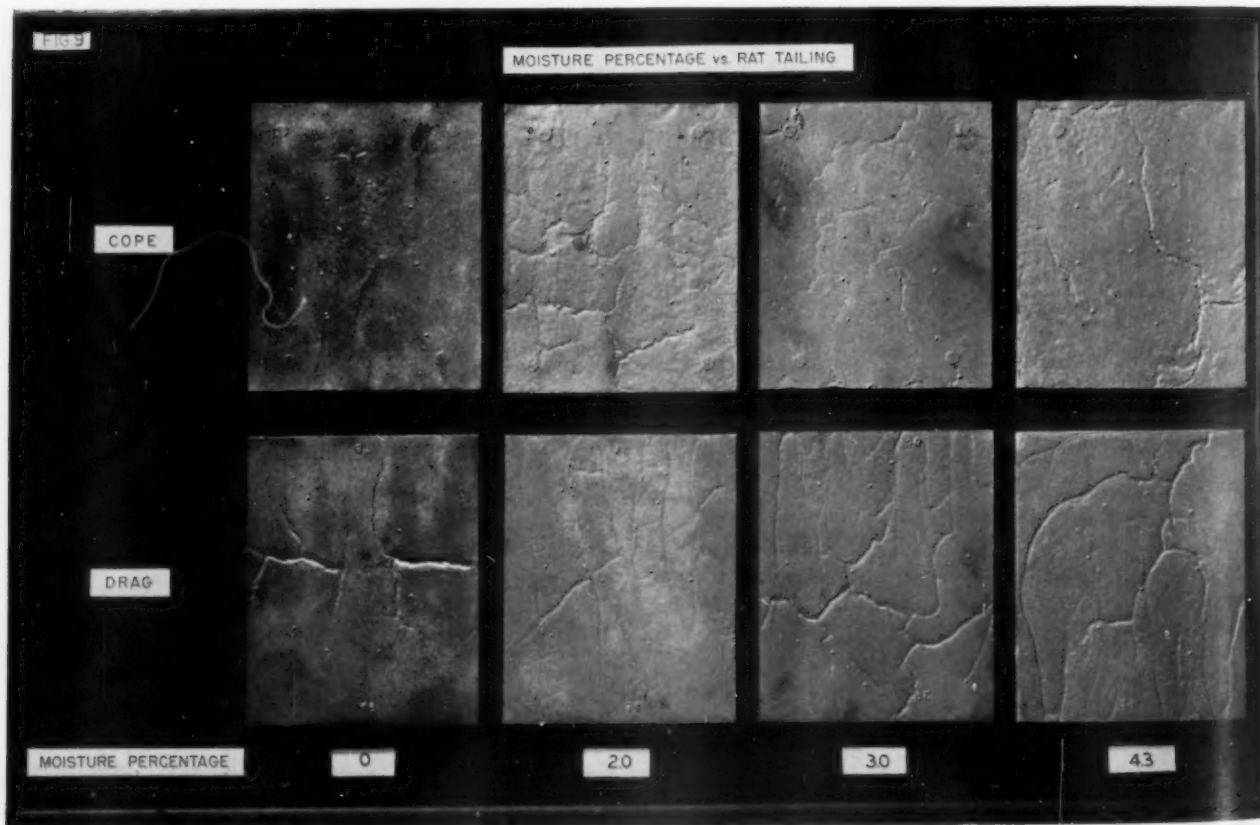


Fig. 9—Cast specimens illustrating effect of mold moisture content on rat tailing.

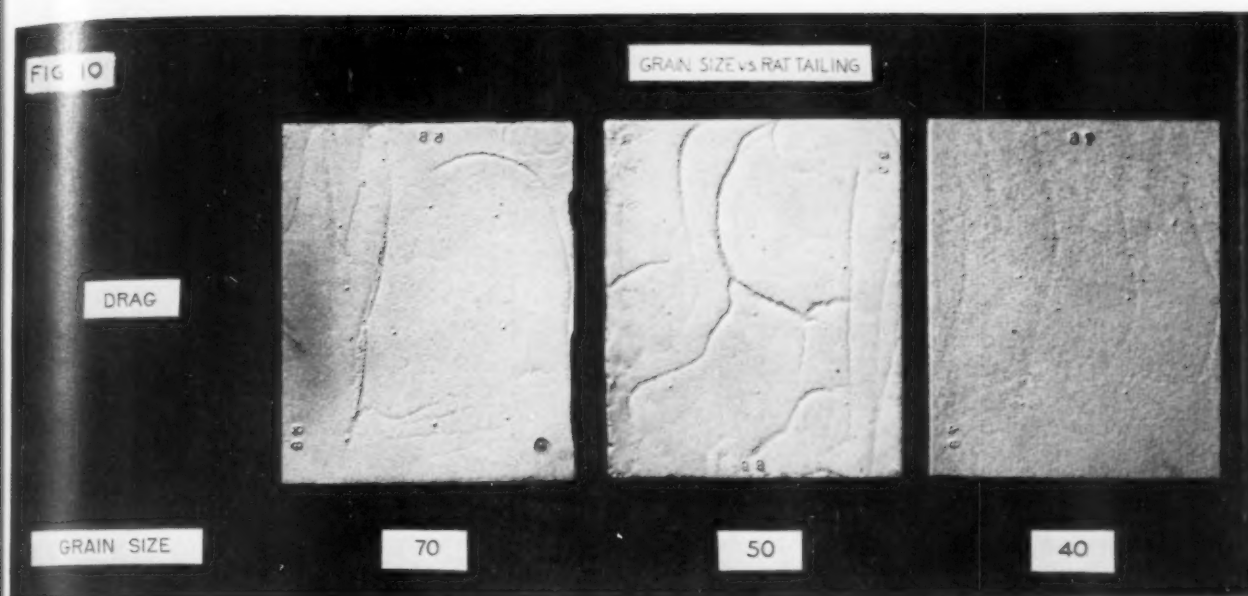


Fig. 10—Cast specimens illustrating effect of sand grain size on rat tailing.

TABLE 4—EFFECT OF GREEN COMPRESSIVE STRENGTH

Facing Sand Mixture	4	5	6
Sand, 100 Mesh, lb	93.8	89.	86.
Clay, lb	5.7	10.	12.6
Western Bentonite, lb	0.5	1.0	1.04
Green Properties:			
Moisture, %	3.4	3.9	4.45
Green Permeability	22.	18.	17.0
Green Compressive strength, psi	3.8	8.1	10.15
Green Deformation, in./in.	0.021	0.018	0.017
Toughness	80.	146.	173.
Flowability	88.	87.	87.
Density, lb/cu ft	100	103	105
Dry Properties:			
Dry Compressive strength, psi	17	415	..
Dry Compressive strength, after 4 Hr Air Set, psi	17	...	11.5
High Temperature Properties:			
Hot Strength at 500 F, psi	81	143	210
1000	93	200	256
1500	106	297	470
2000	275	560	685
2500	53	73	86
Free Expansion at 2500 F	0.0185	..	0.021
Free Hot Shrinkage, 12 min.	0.0085	..	0.075
Confined Expansion at 2500 F	0.0475	0.037	0.031
Confined Hot Shrinkage, 12 min.	0.006	0.006	0.006
Mold Gas Pressure, in. water	20	25	..
Time to Attain Max. Pressure, sec.	21	14	..

Dirt inclusion was observed on castings made in the weakest sand. Lower green, dry and hot strength apparently is conducive to the formation of dirt inclusion.

On castings made in the lower strength sands, the feather edge of the rat-tail defect is easily eroded, resulting in dirt.

#### Conclusions:

Thus, for a severe rat-tailing sand, changing the

TABLE 5—EFFECT OF MOISTURE CONTENT

Facing Sand Mixtures	1	2	3
100 Mesh Sand, lb	90.5	90.5	90.5
Clay, lb	8.75	8.75	8.75
Western bentonite, lb	0.75	0.75	0.75
Green Properties:			
Moisture, %	3.5	2.0	4.1
Green Permeability	20	20	18.
Green Compressive strength, psi	5.4	6.3	5.8
Green Deformation, in./in.	0.020	0.006	0.015
Toughness	108	37.8	87
Flowability	89	90	89
Density, lb/cu ft	102.2	99.8	103
Dry Properties:			
Dry Compressive Strength, psi	105	30	122
Dry Compressive strength after 4-hr Air Set, psi	25	11.5	30
High Temperature Properties:			
Hot Strength, psi at 500 F	97	18	144
1000	143	25.4	193
1500	182	64	308
2000	409	190	489
2500	56	45	78
Free Expansion at 2500 F in./in.	0.0195	0.021	...
Free Hot Shrinkage, 12 min.	0.009	...	...
Confined Expansion at 2500 F	0.041	0.042	...
Confined Hot Shrinkage, 12 min.	0.011	...	...
Mold Gas Pressure at 1200 F in. H <sub>2</sub> O	25	22	22
Time to Attain Max. Pressure, sec.	15	22	14

green strength will not necessarily eliminate rat-tailing.

#### Test F—Effect of Moisture Content

The opinion is often advanced that rat-tails are due to high moisture content. In order to test this hypothesis, molds were made with various moisture contents.

#### Procedure:

Three molds were made with the sand at 3.5 per cent moisture. These molds were then placed in a core



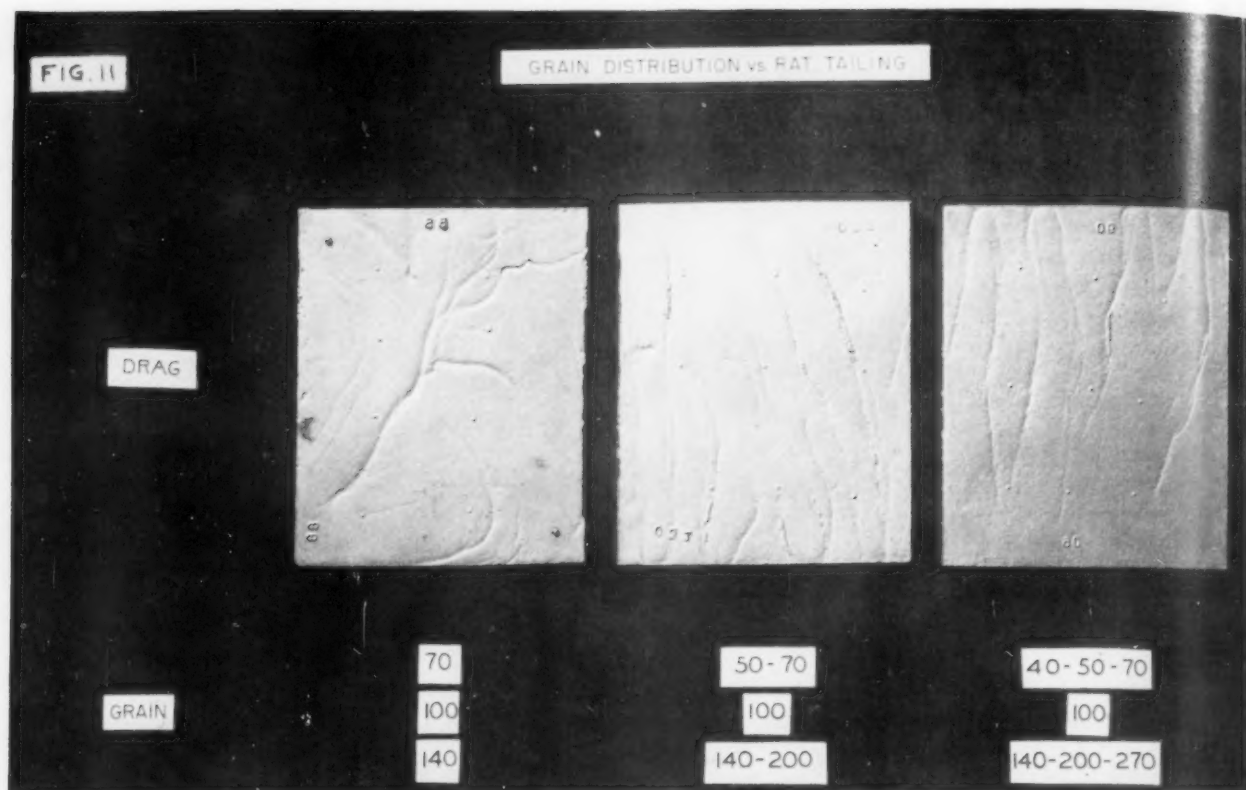


Fig. 11—Cast specimens illustrating effect of sand grain distribution on rat tailing.

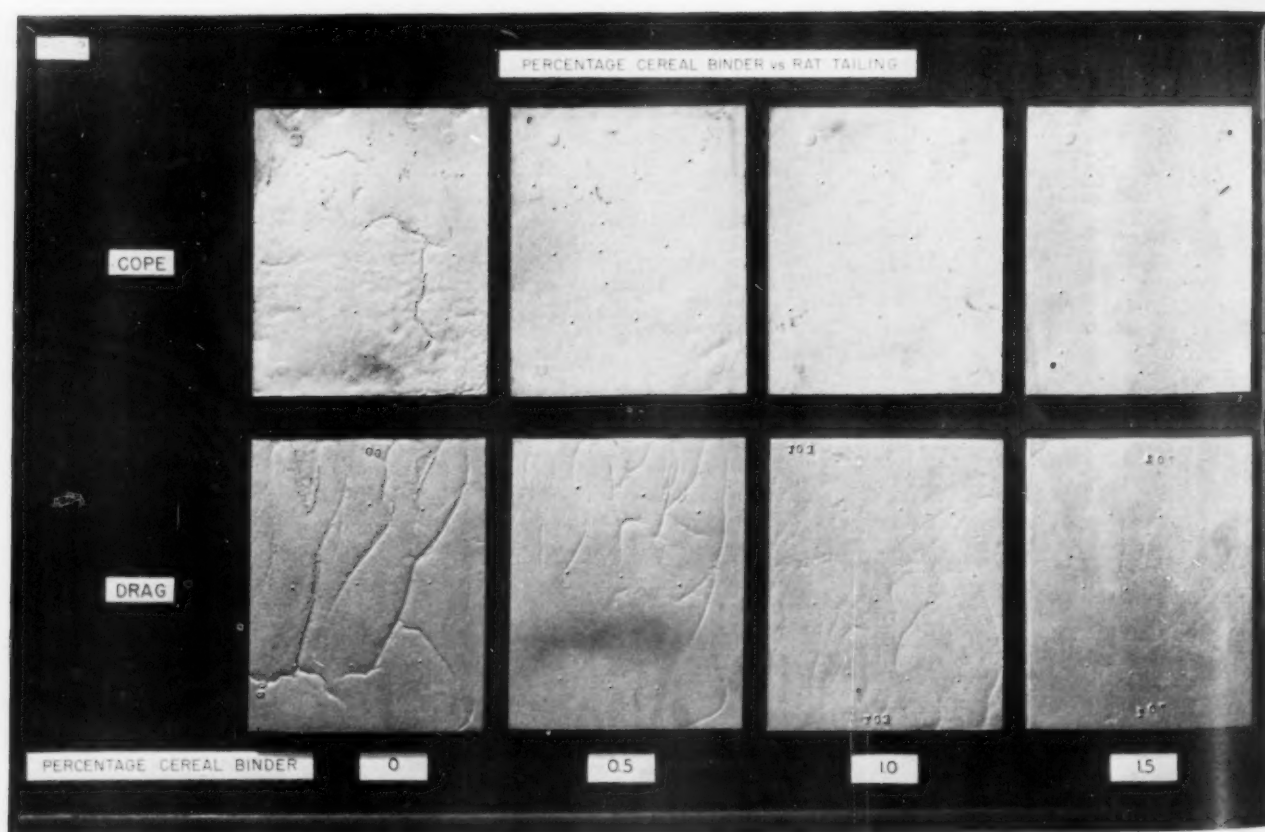


Fig. 12—Cast specimens illustrating effect of cereal binder content on rat tailing.

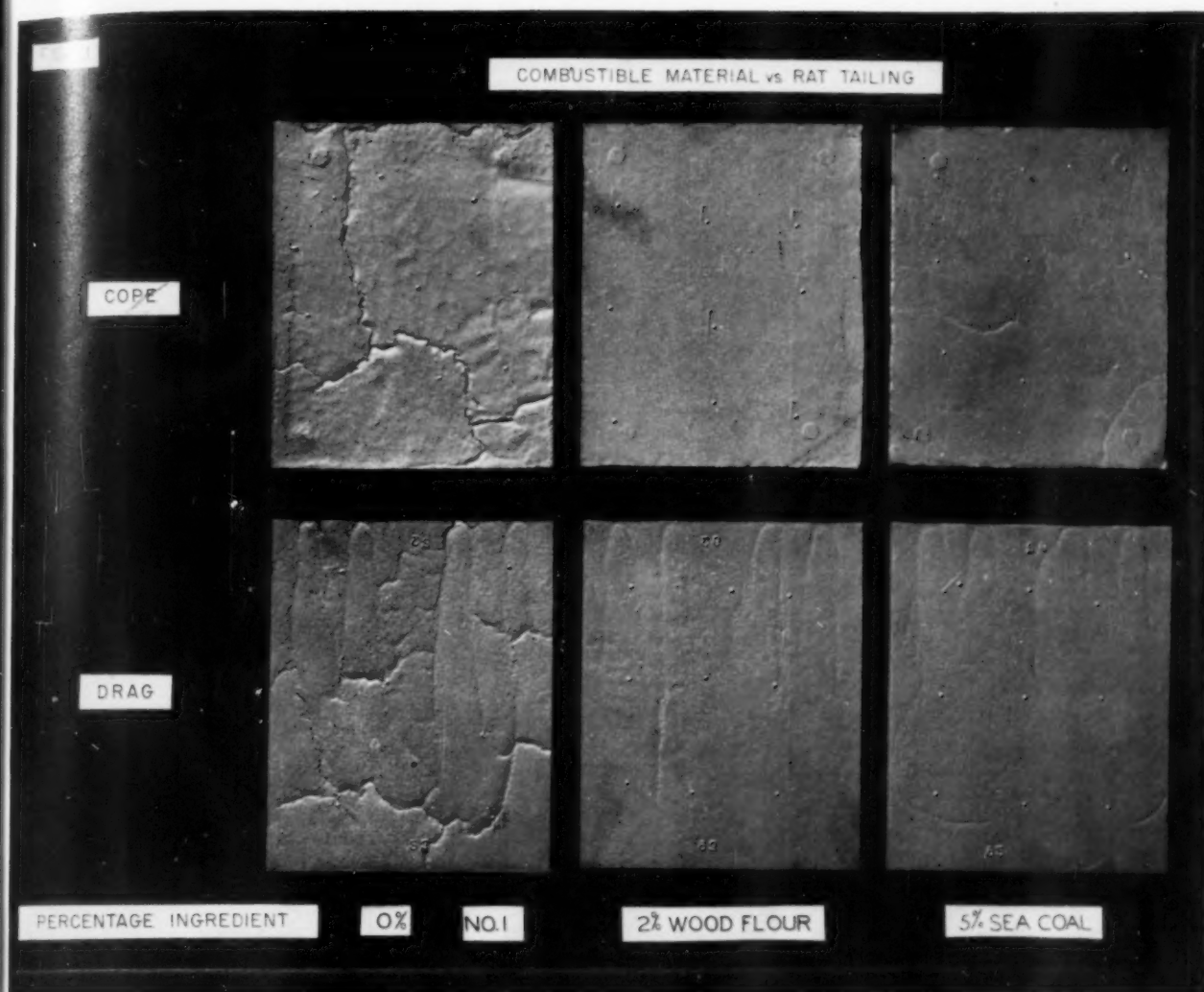


Fig. 13—Cast specimens illustrating effect of combustible material content on rat tailing.

oven at 400 F and dried for six hours. Three molds were made at 2.0 and 4.1 per cent moisture, and were poured as in previous tests. See Table 5.

#### Observations:

The castings made with the above sands are shown in Fig. 9. It is apparent that moisture does not appear to have a marked effect upon the rat-tail defect in this instance when the molds were rammed to the same mold hardness. The magnitude of the rat-tail defect shows an increase with an increase of moisture.

#### Conclusions:

It is logical to conclude that in some cases high moisture might be an indirect cause of rat-tails. Dry strength and hot strength increase with increasing moisture which might cause rat-tails in certain sands.

In this series of tests the moisture was varied and all of the molds rammed to a predetermined hardness value between 78 and 80. In the practice, the high moisture content sand would be heavy and ram hard, thus raising the hot strength and reducing void spaces, both conducive to producing rat-tails. The above test shows that it is not necessarily the steam that blows

off the mold surface, since the mold without moisture produced as many rat-tails as well as the heavily tempered sand. High hot strength in combination with high volume growth of the sand seems to be a major cause of rat-tails.

#### Test G—Effect of Grain Size and Distribution

Sands of each screen size were separated from a washed silica sand. The screened sands were used in preparing facing sand mixtures of various sieve analyses as shown in Table 6.

#### Procedure:

Test castings were made from the sand of the different finenesses. These sands were bonded with 8.75 per cent clay and 0.75 per cent western bentonite and tempered to best workable moisture content as determined by feel.

#### Observations:

No distinct difference in the degree of rat-tail defect was experienced as the grain distribution was changed, except that on the wide grain distribution sands, No. 15, 16 and 17, buckles and rat-tails both appeared.

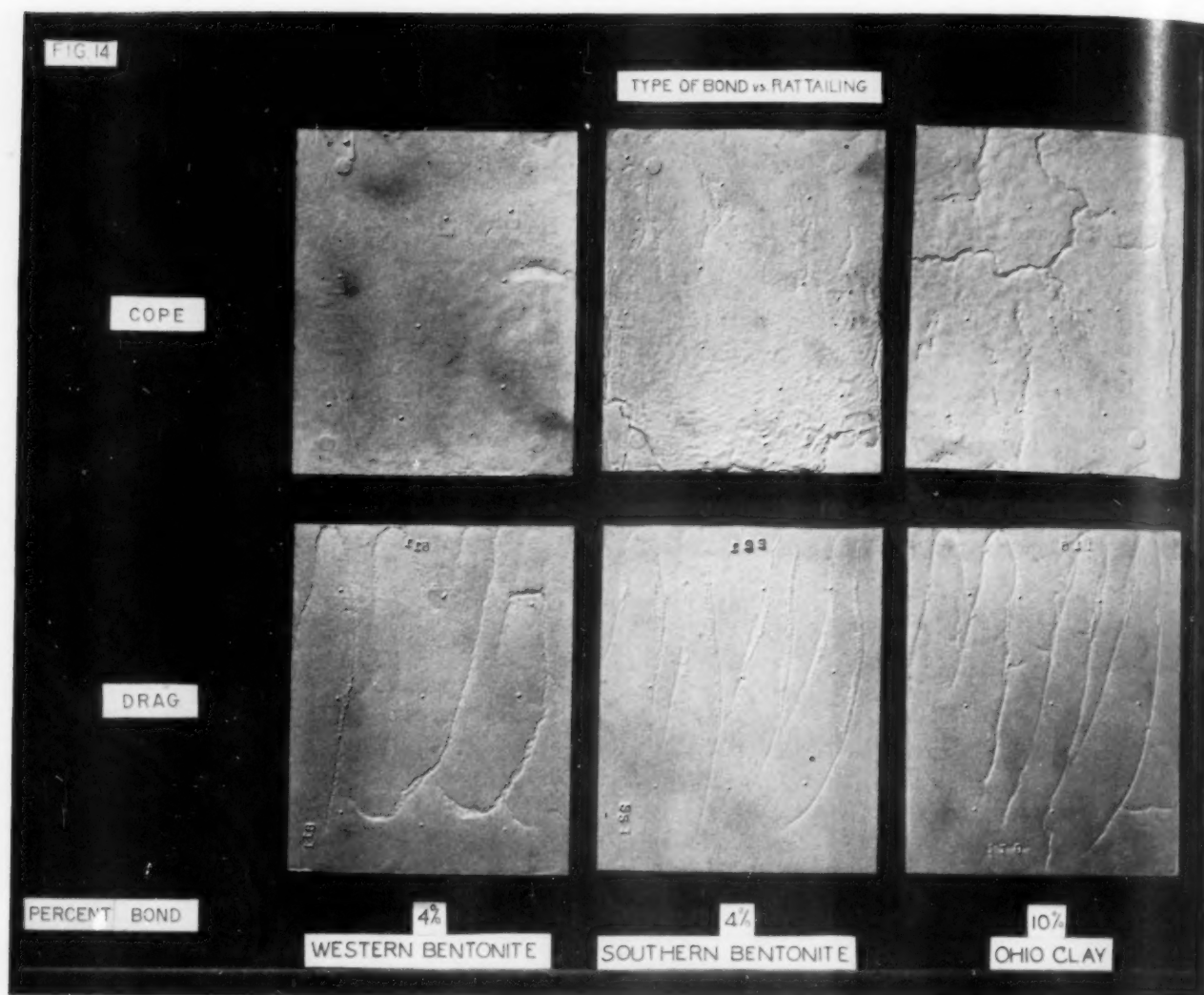


Fig. 14—Cast specimens illustrating effect of type of bond on rat tailing.

The sand mixtures containing 70, 50 and 40 mesh sand (mixture numbers 11, 12 and 13 respectively) may be used to show the effect of grain size on the rat-tail casting defect. Castings from the 70, 50 and 40 mesh sands are shown in Fig. 10. The 40 mesh sand did not produce any rat-tails while the 50 and 70 mesh sands did.

The sand mixtures No. 14 to 17 inclusive give a range in grain distribution. Castings made in these sands are shown in Fig. 11. The sands No. 14 and 16 produced less rat-tails than did the sands No. 15 and 17.

#### Conclusions:

The coarse 40 mesh sand appeared to be less affected by thermal shock.

The shape of the sand grains may be as important as the size of the sand grains.

#### Test H—Effect of Cereal Addition

A 70 mesh washed silica sand was used for these experiments. Mixtures were made containing 0, 1/2, 1 and 1 1/2 per cent cereal by weight.

#### Observations:

Figure 12 shows additions of 1.0 to 1.5 per cent of cereal to the sand to be effective in eliminating the rat-tail defect. See Table 7.

#### Conclusions:

Cereal reduces the flowability and increases the green sand toughness. This tends to reduce expansion and hot strength at 500, 1000 and 1500 F and reducing the flowability produces void spaces which act as a cushion, thereby lessening the tendency to form rat-tails.

#### Test I—Effect of Combustible Materials

Combustible materials commonly added to molding sands are seacoal, wood flour and cereal. Their effect upon the rat-tails was studied by making three molds using facing and sand mixture No. 1 plus the addition of the above materials. The composition of the materials used can be found in Appendix A.

#### Observations:

Figure 13 shows the castings made in the above



TABLE 6—EFFECT OF GRAIN SIZE AND DISTRIBUTION

Mixture No.	Grain Size			Grain Distribution			
	11	12	13	14	15	16	17
Sand, mesh	40	50	70	100	140	200	270
Clay	9.5	9.5	9.5	9.5	9.5	9.5	9.5
Green Properties:							
Moisture, %	2.3	2.2	2.6	3.0	3.0	3.0	3.1
Green Permeability	110	125	340	18.5	22	17	37
Green Compressive strength, psi	5.9	9.1	12.3	11.5	11.3	12.3	8.7
Green Deformation, in./in.	0.014	0.014	0.012	0.013	0.012	0.013	0.014
Toughness	83	127	158	150	135.6	160	122
Flowability	88	88	82	86	...	73	79.5
Density, lb/cu ft	99	101	106	105.5	...	110	...
Dry Properties:							
Dry Compressive strength, psi	34	28.5	87.5	126.7	...	...	...
Dry Compressive strength after 4-hr. air set, psi	14	16	...	...	...	...	26.5
High Temperature Properties:							
Hot Strength, psi at 1500 F	...	...	143.5	...	...	...	245
2000	92	126	342	410	...	186	435
2500	17.5	25.5	13.5	40.3	...	30	38
"A" Sintering Point	2565 F	...	...	...	...	...	2390 F
Free Expansion, 2500 F in./in.	0.0275	...	0.0245	0.026	...	0.025	0.0195
Free Hot Shrinkage, 12 min.	0.012	...	0.010	0.009	...	0.010	0.0165
Confined Expansion	0.043	0.041	0.033	0.035	...	0.032	0.036
Confined Hot Shrinkage	0.060	0.0085	0.0045	0.007	...	0.007	0.009
Mold Gas Pressure, at 1200 F, in water	45	13	3	20	...	21	...
Time to Attain Max. Press., Sec.	15	20	13	17	...	16	...

TABLE 7—EFFECT OF CEREAL ADDITION

Mixture No.	20	21	22	23
70 Mesh Silica Sand, lb	90.5	90	89.5	89
Western Bentonite, lb	0.75	0.75	0.75	0.75
Clay, lb	8.75	8.75	8.75	8.75
Cereal (Corn) lb	...	0.5	1.0	1.5
Green Properties:				
Moisture, %	3.5	3.5	3.6	3.9
Green Permeability	53	59	57	54
Green Compressive strength, psi	5.7	5.4	5.1	5.8
Green Deformation, in./in.	0.015	0.023	0.025	0.037
Toughness	85.5	124	127.5	215
Flowability	89	84	76	74
Density, lb/cu ft	101.5	100	99.75	100.5
Dry Properties:				
Dry Compressive strength, psi	...	...	...	193
Dry Compressing strength, after 4-hr Air Set, psi	...	...	...	23.5
High Temperature Properties:				
Hot Strength, psi at 500 F	81	76	45	53
1000	157	120	63.5	77
1500	302	163	127	82
2000	420	355	387	339
2500	20	32	28	29
Free Expansion				
2500 F in./in.	0.023	0.015	0.015	0.0185
Hot Shrinkage	0.011	0.0155	0.0135	0.0125
Confined Expansion				
2500 F in./in.	0.044	0.034	0.037	0.032
Hot Shrinkage	0.007	0.0085	0.007	0.0069
Sintering "A" deg. F	...	...	...	2600
Gas Pressure at 1200 F, in. water	16	20	19	22
Time to reach max. pressure, sec.	15	17	13	18

TABLE 8—EFFECT OF COMBUSTIBLE MATERIALS

Mixture No.	1	18	19	24
100 Mesh Sand, lb	90.5	85.5	...	88.5
Western Bentonite, lb	0.75	0.75	0.75	0.75
Clay, lb	8.75	...	...	...
Seacoal, lb	...	5.0	...	...
Wood Flour, lb	...	...	2.0	...
Cereal, lb	...	...	...	2.0
Green Properties:				
Moisture, %	3.5	3.0	4.2	5.0
Green Permeability	20	16	12.5	67.5
Green Compressive strength, psi	5.4	7.8	6.1	7.0
Green Deformation, in./in.	0.020	0.013	0.020	0.029
Toughness	108	91	122	203
Flowability	89	78	81	69
Density, lb/cu ft	102	99.5	101	100.5
High Temperature Properties:				
Hot Strength, 500 F	97	62	31	38
1000	143	65	35	86
1500	182	127	80	121
2000	409	210	198	390
2500	56	47	43	46
Free Expansion,				
2500 F in./in.	0.0195	0.020	0.015	0.014
Hot Shrinkage	0.011	0.0025	0.0125	0.009
Confined Expansion,				
2500 F	0.0445	0.034	0.033	0.029
Hot Shrinkage	0.009	0.025	0.0055	0.006
Sintering—"A"	2420 F	2625 F	2450 F	2390 F
Gas Pressure at 1200 F, in. water	25	21	27	25
Time to reach max. pressure, sec.	15	16	14	23

sands. It may be noted that additions of combustible material reduce the tendency to rat-tail. See Table 8.

#### Conclusions:

The combustible materials added lowered the flowability, hot strength and expansion of the sand mixture. Burning of these organic materials also lowers the density of sand and leaves voids which prevent high stresses from being built up in the sand. Without high stresses being built up on the surface of a mold, no mold surface failure results to produce a rat-tail. The estimated change in density due to the additions of combustible material is illustrated in Table 9.

TABLE 9—ESTIMATED CHANGE IN DENSITY DUE TO COMBUSTIBLE MATERIALS

	No Addition	5% Seacoal	2% Flour	2% Cereal
Density, green, lb/cu ft.	102	99.5	101	100.5
Moisture, %	3.5	3.0	4.2	5.0
Volume of 100 lb green sand, cu ft	0.98	1.005	0.9901	0.9524
Volume of moisture, cu ft	0.0564	0.0484	0.06774	0.0806
Volume of combustible material, cu ft	Zero	0.0943	0.1335	0.0514
Density, burnt out sand, lb/cu ft	96	85.4	80.5	85.5

TABLE 10—EFFECT OF BOND TYPE

Mixture No.	27	28	29
100 Mesh Silica Sand, lb	90	96	96
Clay, lb	10	...	...
Western Bentonite, lb.	...	4.0	...
Southern Bentonite, lb	...	...	4.0
Green Properties:			
Moisture, %	3.9	3.2	2.9
Green Permeability	17	32	37
Green Compressive strength, psi	6.5	6.0	7.1
Green Deformation, in./in.	0.016	0.021	0.014
Toughness	102	126	147
Flowability, %	...	78	78
Density, lb/cu ft	100	98	100.5
Dry Properties:			
Dry Compressive strength, psi	66	72	...
Dry Compressive strength, after 4-hr air set, psi	...	...	...
High Temperature Properties:			
Hot Strength, psi at 500 F	76	127	78
1000	134	143	89
1500	218	408	115
2000	56	412.5	42.5
2500	53	...	12
Free Expansion, 2500 F, in./in.	0.040	0.018	0.0223
Hot Shrinkage	0.0107	0.015	0.003
Confined Expansion, 2500 F, in. in.	0.0355	0.035	0.031
Hot Shrinkage	0.009	0.005	0.001
Sintering—A	...	...	2565 F
Gas Pressure—in water	22	22	22
Time to reach max. pressure, sec.	20	17	21

#### Test J—Effect of Bond Type

Three commonly used foundry bonds, western bentonite, southern bentonite and fireclay were used to bond a 100-mesh silica sand. Three castings were made from each sand to determine the effect of each upon the rat-tail. See Table 10.

#### Observations:

Figure 14 shows the castings made in sands bonded with different bonds. Castings made in fireclay and western bentonite bonded sand were observed to have rat-tails.

#### Conclusions:

The southern and western bentonite bonded sand resulted in castings with fewer rat-tails.

The sand mixture bonded with fireclay produced castings showing the rat-tails as shown in Fig. 14. This sand possessed relatively high hot strengths.

#### Test K—Effect of Natural Molding Sands

To determine the effect of natural molding sands on the occurrence of rat-tails, three test castings were made from mixture No. 7, 8, 9 and 10 as shown in Table 11.

TABLE 11—EFFECT OF NATURAL MOLDING SANDS

Mixture No.	7	8	9	10
Green Properties:				
Moisture, %	9.0	10.0	8.2	10.2
Green Permeability	45	57	8.0	14
Green Compressive strength, psi	5.6	12.2	9.3	5.5
Green Deformation, in./in.	0.020	.02024	0.022	0.029
Toughness	110	292	123	269
Flowability	72.5	55.5	75.5	72.5
Density, lb/cu ft	...	108.5	101.7	...
Dry Properties:				
Dry Compressive strength, psi	...	60	35	120
Dry Compressive strength, after 3-hr air set, psi	...	52	11	25
High Temperature Properties:				
Hot Strength, psi at 500 F	85	51.5	27	85
1000	96	112.5	36	129
1500	225	179.7	43	242
2000	43	412.2	110	402
2500	5.5*	16.5	1.0	4
Free Expansion,				
2500 F, in./in.	0.020	0.019	0.008	0.019
Hot Shrinkage	0.031	0.006	0.090	0.085
Confined Expansion,				
2500 F, in./in.	0.018	0.022	0.020	0.020
Hot Shrinkage	0.002	0.022	0.083	0.046
Sintering—"A"	2135 F	2540 F	2215 F	2275 F
Gas Pressure, in. water	...	...	...	...
Time to reach max. pressure, sec.	...	...	...	...

#### Observations:

The castings obtained from these natural molding sands were all free of rat-tail defects except No. 7, which was a used sand. No. 7 sand (Ohio Heap Sand) showed a few rat-tail marks as may be noted in Fig. 15.

#### Conclusions:

The natural sands used in this test were all new, except the No. 7 sand which contained a small percentage of seacoal. Since the natural sands were new,

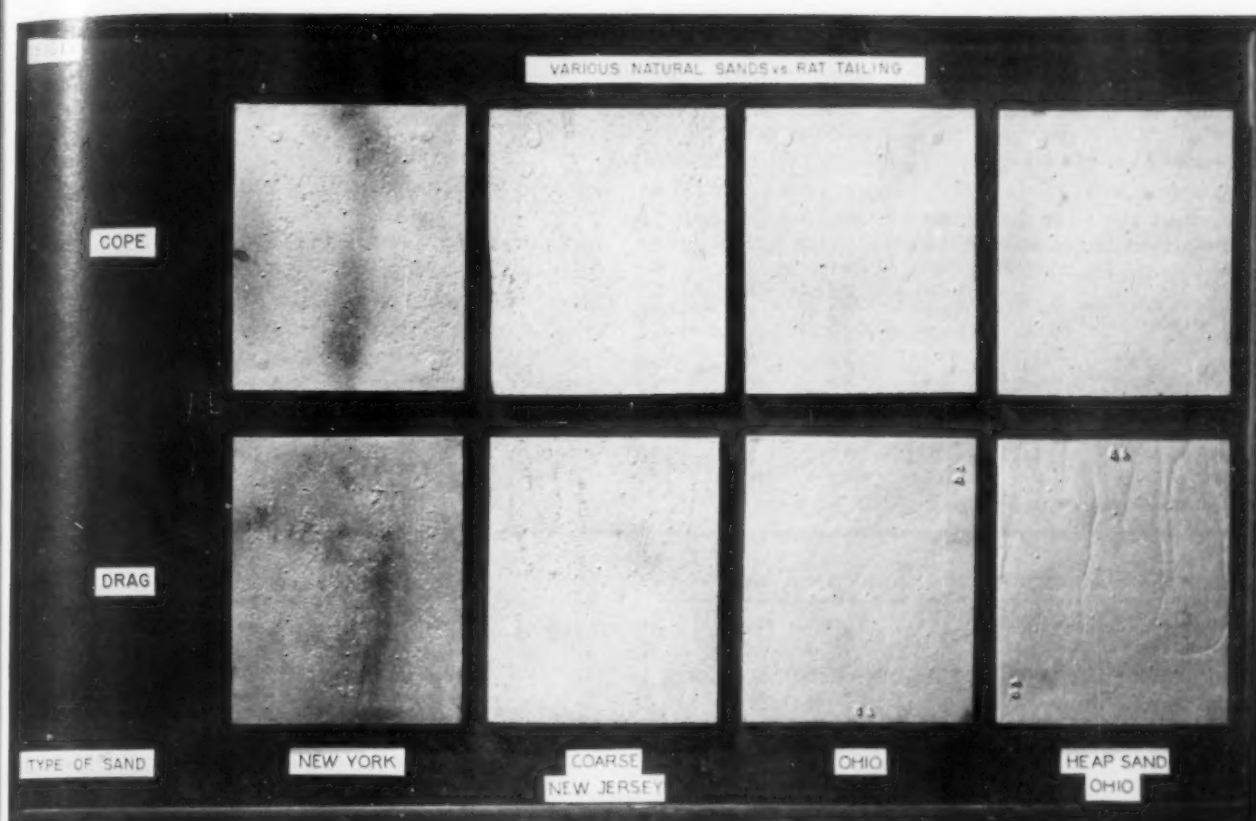


Fig. 15—Cast specimens illustrating effect of various natural sands on rat tailing.

they all contained combustible organic materials and exhibited low tendency to produce rat-tails.

#### Discussion

Although a number of sand mixtures were tested in this investigation, results do not justify broad general statements. The results however, are of sufficient scope to allow for good, practical application in the foundry and should prove helpful in controlling the rat-tail defect. They can also serve to encourage further research on rat-tails and their elimination.

The "100-mesh silica sand" used to make synthetic sand is not representative of the majority of synthetic sands used in foundries. It served as a test sand to illustrate some factors affecting the rat-tail defect.

Since most molding sand is composed of approximately 92 per cent or more  $\text{SiO}_2$ , it is logical to assume that the volume expansion of an individual sand grain is about constant for a wide variety of sands. Practical observation has shown that specimens of rammed sand vary considerably in their expansion characteristics. The difference observed between the thermal expansion of rammed sand specimens is attributed to the mechanical structure of the specimen and to the reaction to heat of the binder on the surface of the grains.

The elimination of the rat-tail defect in this series of investigations was largely effected by making the 100-mesh silica sand mixture less dense at high temperatures, or by the use of low refractory clay; thereby allowing the sand grains to expand more freely at

high temperatures. Two factors which tend to give this effect are use of clay, which shrinks at high temperatures, or the use of materials which result in low density at these temperatures, for example, controlling ramming.

The absence of rat-tails on castings obtained with the natural bonded molding sands is attributed to their low hot-strength at 2000 and 2500 F. The clay in the natural molding sands apparently softens at a lower temperature than clay or western bentonite. The natural sands used had more combustible material, different grain size and distribution, and a much higher clay content than the 100-mesh silica sand mixtures.

The particular base sand chosen for most of the experiments had such a pronounced tendency to produce rat-tails that some variables such as metal thickness, green compressive strength, etc., had no marked effect upon the rat-tail phenomena. It is quite possible that these variables would have some effect of a sand which had a lesser tendency to produce rat-tails. Therefore, further work with a wider variety of sands is considered necessary to clarify the nature of the rat-tail phenomena.

Although the expansion and hot strength tests are useful in studying sand properties at elevated temperatures, a third property of molding sand needs to be measured. The toughness of sand when exposed to high temperatures possibly is a major factor in preventing rat-tails. Presumably, a brittle sand may cause rat-tails whereas one which is plastic may not.



TABLE 12—FINENESS OF SANDS USED IN THE RAT-TAIL CASTING DEFECT STUDY

Sand	100 Mesh*	70 Mesh*	No. 7	No. 8	No. 9	No. 10	No. 11	No. 12	No. 13	No. 14	No. 15	No. 16	No. 17
Sieve Number	Per Cent												
6	...	...	...	...	...	...	...	...	...	...	...	...	...
12	...	...	0.8	...	0.4	...	...	...	...	...	...	...	...
20	...	...	1.4	0.1	2.0	0.2	...	...	...	...	...	...	...
30	...	...	2.4	0.1	2.9	0.1	...	...	...	...	...	...	...
40	...	...	7.2	0.2	4.8	0.7	...	...	90.5	...	...	13	18.1
50	...	...	12.8	0.2	4.1	4.05	...	90.5	...	...	18.1	13	18.1
70	...	17.6	7.8	0.6	15.0	14.8	90.5	...	...	30.2	18.1	13	18.1
100	8.6	41.8	34.4	2.6	8.9	13.0	...	...	...	30.2	18.1	13	18.1
140	43.2	29.6	6.0	8.0	32.3	9.3	...	...	...	30.2	18.1	13	18.1
200	33	8.2	3.6	16.7	3.8	13.2	...	...	...	...	18.1	13	...
270	12.2	1.6	2.4	19.6	1.4	11.15	...	...	...	...	...	13	...
Pan	3.0	0.6	3.6	38.5	1.8	16.0	...	...	...	...	...	...	...
Clay	0.5	0.6	17.6	13.4	22.6	17.5	9.5	9.5	9.5	9.5	9.5	9.5	9.5

\* Washed and Dried Silica Sand.

TABLE 13—COMPOSITION OF COMBUSTIBLE MATERIALS

Retained on U.S. Sieve No.	Per Cent Retained	Chemical Analysis	
SEACOAL			
40	1.8	V.C.M.	34.0%
50	...	Fired Carbon	58.0
70	26.6	Ash	8.0
100	13.4	Sulphur	0.9
140	10.4	Fusion of Ash	2680°F.
200	9.8		
270	6.8		
Pan	31.2		
WOOD FLOUR			
50	Trace	Moisture	.35% max.
100	17.0	Water Soluble	3.67
140	42.5	Naptha Sol.	.545
200	11.5	Ash	0.85 max.
270	2.5	Alcohol Soluble	3.6
Pan	26.5	Balance	Pure-cellulose
CEREAL			
		Moisture	10.0%
		Fat	Trace
		Fiber	None
		Protein	5.7
		Ash	0.35
		P.	4.5

Sands can be cushioned by causing them to have a lower density when heated by molten metal. Materials which are effective in reducing rat-tails are seacoal, wood flour and cereal, since they decompose and provide voids for sand grain growth.

Synthetic sands were largely used in this study so that greater flexibility in sand conditions could be had. One could make a very poor sand or a very good sand. Equally good castings of gray iron were made with either synthetic or natural sands.

Lowering the mold strength by decreasing the mold hardness below 60, eliminated the rat-tail defect for the sands studied.

The use of cushioning materials in facing sands such as wood flour, fine bran, cereal binder and seacoal offer means of eliminating the rat-tail casting defect.

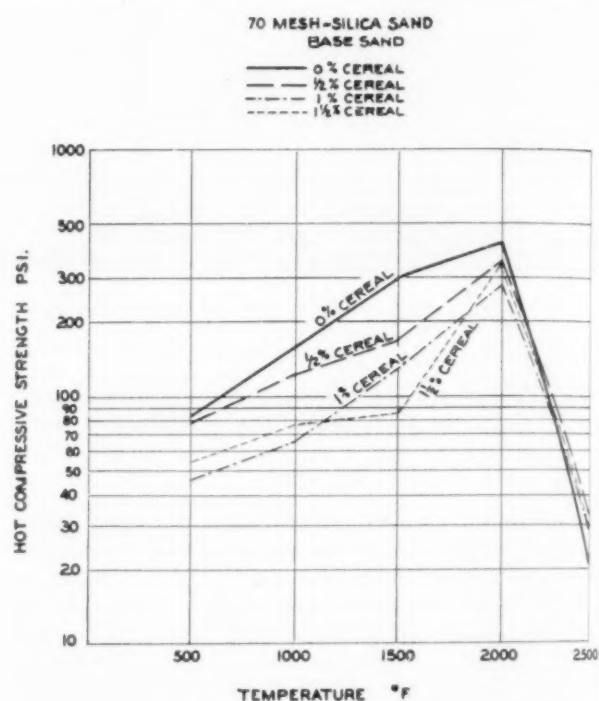


Fig. 16—Effect of Cereal Binder on the Hot Strength of No. 1 Sand Mixture.

## APPENDIX A

### Analysis of Material Used

The fineness data of sands used in the study on the causes of rat-tail casting defects are tabulated in Table 12.

Thirteen different sand grain size structures studied ranged from narrow to wide grain distribution. A study of the data indicates that grain size or grain distribution does not offer a practical and positive means of eliminating the rat-tail casting defect. Using a very coarse grain size does stop rat-tail defects at the expense of casting finish. This is not considered a practical solution inasmuch as rat-tail defects appear mostly on the lighter flat surface castings.

The composition and screen analyses of the mate

rials used in Test I, "The Effect of Combustible Materials" are shown in Table 13.

## APPENDIX B

### Effect of Cereal on Hot Compressive Strength

When cereal binder is added to a sand many of the physical properties of the sand are altered.

Figure 16 shows how cereal addition affected the hot strength of a 70-mesh sand. This sand without cereal, produced rat-tails. The addition of cereal reduced the defect to the extent that 1.5 per cent completely eliminated the rat-tail on this severe rat-tailing sand. The effect of cereal upon hot strength at 2000 and 2500 F was not very large. However, the hot strength at 500 and 1500 F were lowered considerably by cereal additions.

## APPENDIX C

### Effect of Various Ingredients on Hot Strength

Figure 17 shows the effect of southern bentonite, seacoal, cereal and wood flour upon the hot strength of 100-mesh sand mixtures.

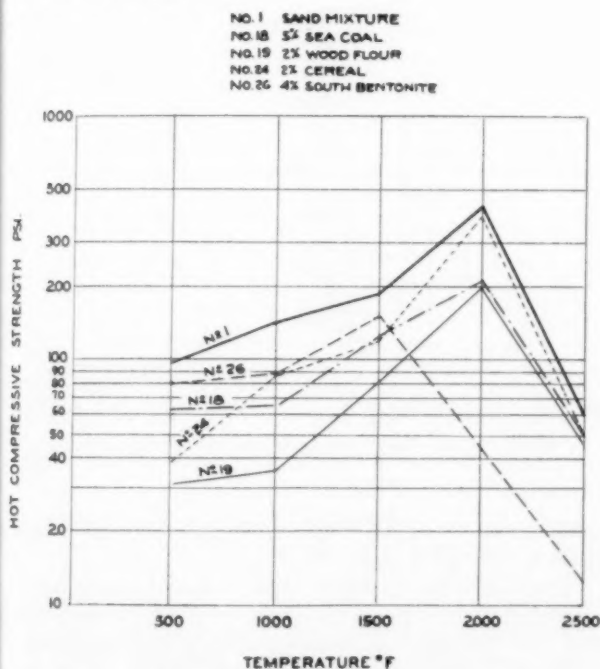


Fig. 17—Effect of Various Ingredients on the Hot Strength.

Southern bentonite lowers the hot strength of the sand at 2500 F. The other materials have little effect upon strength at this temperature. The southern bentonite results in a lower hot strength. The combustible materials have little effect on hot strength at 2500 F. Their function is to provide voids in the sand to accommodate grain growth without high stresses.

## APPENDIX D

### Gas Pressure is not Related to Rat-Tail Defect

The opinion is commonly held that steam or gas pressure is a cause of rat-tail defects. A number of the

sands used in this investigation were tested for their capacity to generate gas.

### Method:

A sand specimen 3 in. long and  $1\frac{1}{8}$  in. dia. was rammed by three drops of a 7-lb. weight falling  $25\frac{1}{8}$  in. A copper tube was rammed in the specimen along the longitudinal axis, the end of the tube was  $\frac{1}{8}$  in. from the end of the specimen. The specimen was immersed to a depth of  $2\frac{3}{4}$  in. in a bath of lead heated to 1200 F, and time and maximum pressure were recorded. This data is given in Table 14.

## APPENDIX E

### Thermal Expansion of Sand is Related to Rat-Tail Defect

Two methods of measuring the expansion of molding sand were used in obtaining the data described in the main body of this report.

### Free Expansion:

A  $1\frac{1}{8}$ -in. x 2-in. specimen was placed on a carbofrax disc. Another carbofrax disc was put on top of the specimen. The specimens were exposed to a temperature of 2500 F. A quartz tube supported the specimen and a quartz rod rested on top of the specimen with a pressure of 4 oz. Readings were taken as soon as the specimen was put in the furnace. A dial indicator was used to measure movement of the specimen.

TABLE 14—GAS PRESSURE AS RELATED TO RAT-TAIL DEFECTS

Sand No.	Description	Maximum Time Required Gas Pressure, to Obtain Max. In. of Water Pres. in Sec.	
		In. of Water	Pres. in Sec.
1	Dried Sand	5	31
	3.5% Moisture	25	15
	2% Moisture	22	22
	4.1% Moisture	22	14
4	3.8 psi Green Compressive Strength	20	21
	8.1 psi Green Compressive strength	25	14
8	New Jersey Molding Sand	18	11
9	New York Molding Sand	24	24
10	Ohio Molding Sand	24	11
11	70 Mesh Sand	4.5	15
12	50 Mesh Sand	13	20
13	40 Mesh Sand	3	13
14	70, 100 and 140 Mesh Sand	20	17
16	40 thru 270 Mesh Sand	21	16
20	No cereal addition	16	15
21	0.5% Cereal	20	17
22	1.0% Cereal	19	13
23	1.5% Cereal	22	18
25	10% Fireclay	14	17
26	4% Southern Bentonite	10	25
18	Seacoal	21	16
19	Wood Flour	27	14
24	Cereal	25	23
28	Western Bentonite	22	17
29	Southern Bentonite	22	21
27	Fireclay	22	20

Comments: For the sands examined, no apparent relationship between gas pressure and the rat-tail defect was found. A dried sand which developed a gas pressure of 5 in. in 31 sec. produced approximately as many rat-tails as the same sand tempered to 4.1 per cent moisture which produced gas pressure of 22 in. in 14 sec. The steam pressure in the molds which were used in this investigation apparently played no part in producing rat-tails.

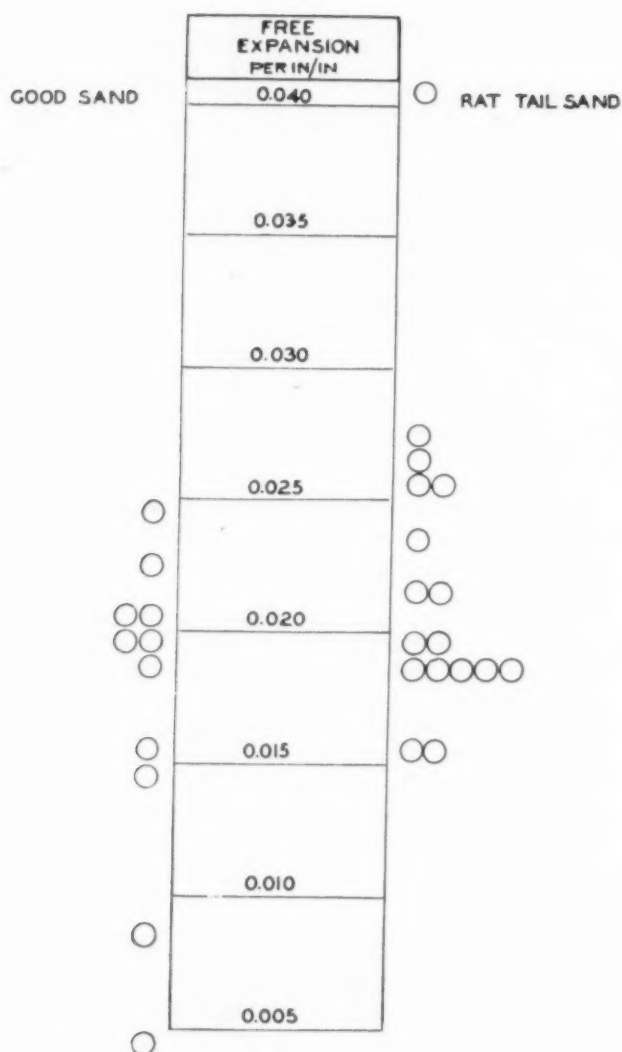


Fig. 18—Relationship between Free Expansion at 2500 F and Rat-Tail Defects. Solid Specimen.

#### Confined Expansion:

This test is similar to free expansion with the exception that the specimen is rammed in a quartz tube. This limits lateral expansion and sagging.

Expansion is recorded as the greatest increase in length observed.

Hot shrinkage is the difference in length between the fully expanded and fully contracted specimen.

A statistical analysis of the test data reported in the main body of this report is given in Fig. 18, 19, 20 and 21.

The free expansion test alone will not distinguish between sands that will or will not produce castings with rat-tails. Note Fig. 18. There is evidence however, that when the free expansion of a sand that produces rat-tails is reduced, then the rat-tail defects will not occur.

With reference to free hot shrinkage, the following is of interest. Sands which soften considerably at 2500 F to yield a high hot shrinkage above 0.018 in./in. or have very low hot shrinkage of 0.006 in./in. did not produce rat-tails. One thus has a range from

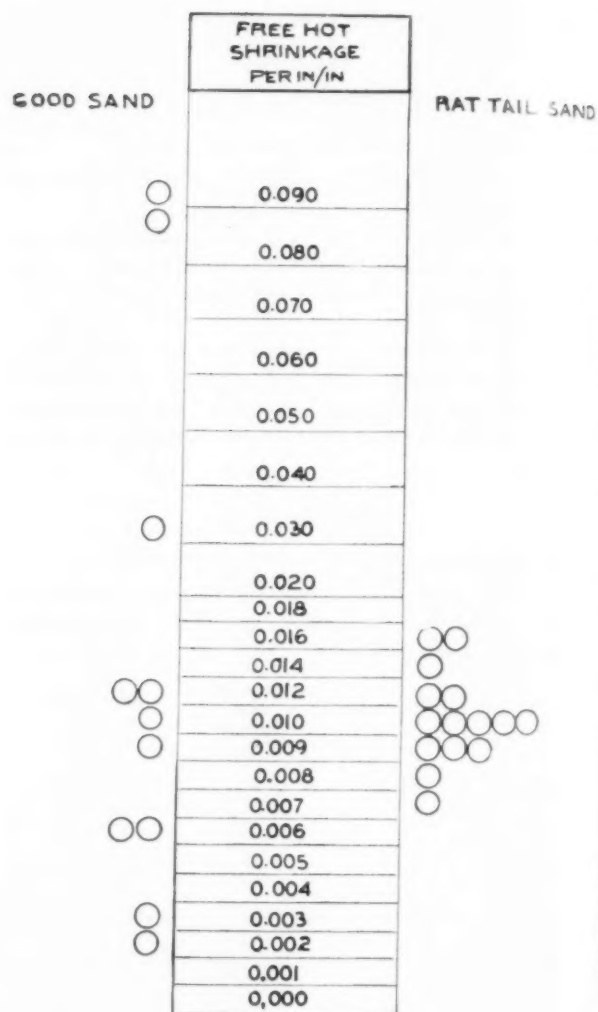


Fig. 19—Relationship between Free Hot Shrinkage at 2500 F and Rat-Tail Defects. Solid Specimen.

0.007 to 0.018 in./in. hot shrinkage where no rat-tail defects were produced. See Fig. 19.

There is some evidence that confined expansion in excess of 0.040 in./in. is conducive to rat-tails. See Fig. 20. The sand producing rat-tails possessed higher confined expansion on the average than sands that did not produce any rat-tails.

In this study, the sands that produced rat-tails possessed a confined hot shrinkage which fell within the range from 0.006 to 0.012 in./in. The sand that did not produce any rat-tails had either less than 0.006 or more than 0.012 in./in. confined hot shrinkage. See Fig. 21.

#### Summary

Figures 18, 19, 20 and 21 summarize the data on expansion tests reported in the main body of this report. Tentative conclusions based on the limited amount of evidence at hand are:

1. Confined expansion test results correlate more closely with casting defects than do free expansion tests.



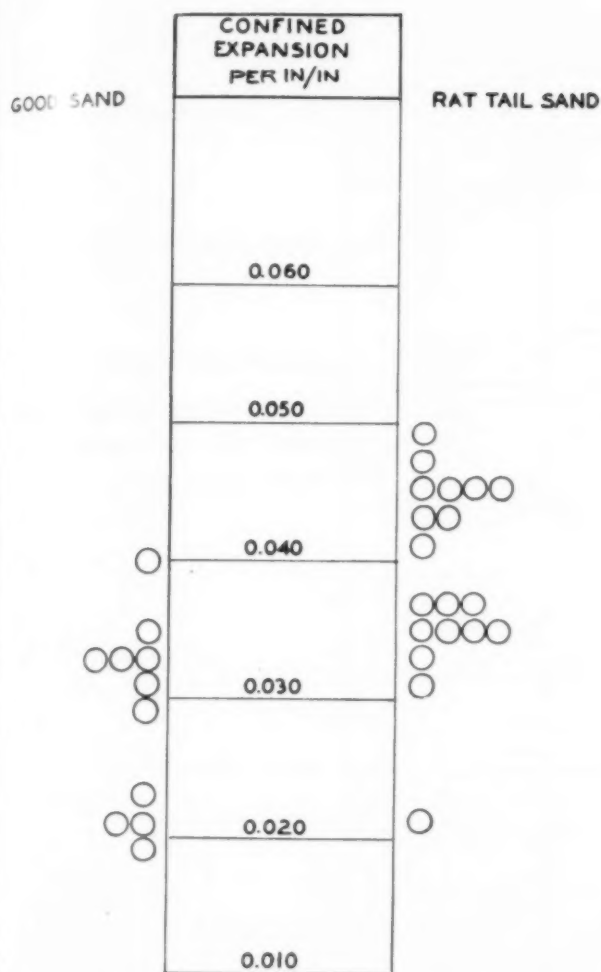


Fig. 20—Relationship between Confined Expansion at 2500 F and Rat-Tail Defects. Solid Specimen.

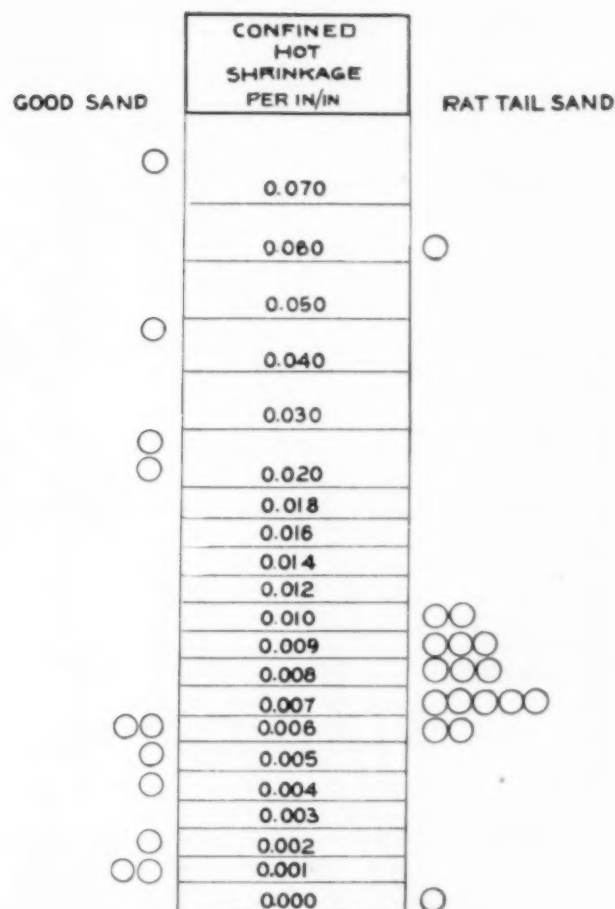


Fig. 21—Relationship between Confined Hot Shrinkage at 2500 F and Rat-Tail Defects. Solid Specimen.

- There is some indication that confined expansion in excess of .040 in./in. is characteristic of the rat-tailing sands examined.
- Both confined hot shrinkage and free hot shrinkage seem to be related to the rat-tail defect.
- The following theory is advanced to account for the results obtained on the hot shrinkage tests: There are three "zones" of hot shrinkage:
  - Those sands which shrink less than 0.006 in./in. generally do not produce rat-tails.
  - Sands which shrink 0.006 to 0.020 in./in. tend to produce rat-tails.
  - Sands which soften and sag more than 0.020 in./in. generally do not produce rat-tails.

## APPENDIX F

### Hollow Expansion Specimens

It is recognized that a 1 1/8-in. diameter or other solid sand specimen does not heat up uniformly, and the expansion when exposed to 2500 F temperature roughly quantitative. During the initial stages of heating the surface is expanding faster than the center, and in the final stages of heating, the center is expanding while the surface is contracting.

In an effort to obtain more uniform heating rates, a hollow 1 1/8-in. O.D. x 1/2-in. I.D. sand specimen was tried. Results are shown in Fig. 22.

Although the amount of work done is insufficient for definite conclusions, this test seems worthy of further investigation. The expansion curve obtained by this method indicates that the hollow specimen changes length more quickly and smoothly when exposed to heat than does the solid 1 1/8-in. dia. sand specimen.

## APPENDIX G

### Hot Strength and Rat-Tails

The hot strength determinations in this report were made with a table rise of 1 in. per min. so that the sands were crushed under constant deformation. The higher the hot strength of a sand at the face of a mold, the greater was the magnitude of the stresses caused by the expansion of the sand at the mold face.

The hot strength at 2000 F of the various sands is plotted in Fig. 23 in two columns, one for good sands and one for rat-tail sands. None of the good sand possessed very high hot strengths. The majority of the rat-tail sands possessed high hot strengths. Thus,

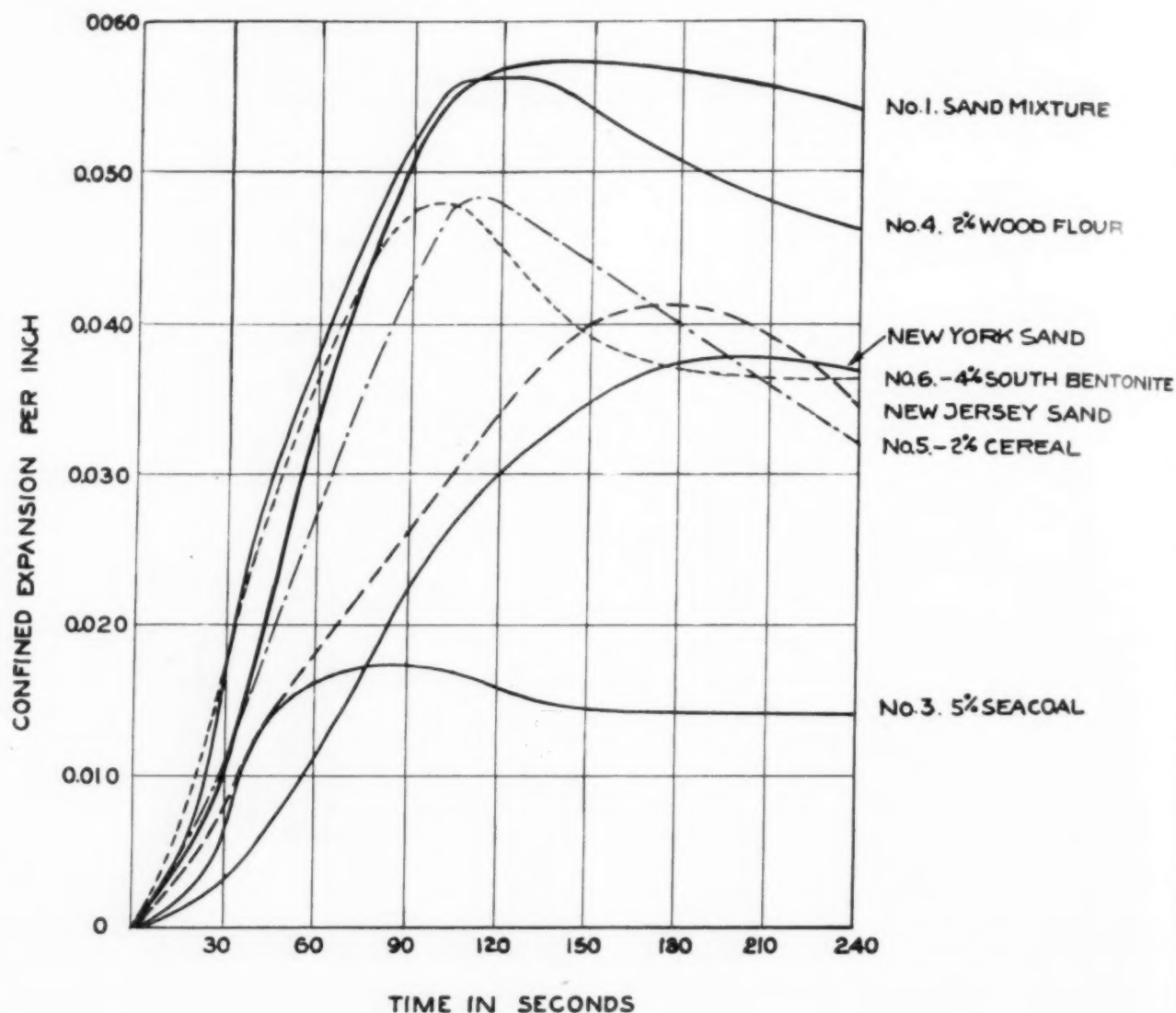


Fig. 22—Confined Expansion at 2500 F. Hollow Test Specimen.

using sands with hot strengths as low as possible without obtaining cuts and wash defects, is good practice in a rat-tail elimination program.

#### APPENDIX H

##### *Expansion, Hot Strength and Rat-Tails*

It was characteristic of the sands which caused rat-tails in the investigation that they had a confined expansion greater than .040 in./in. and also a hot strength in excess of 100 psi at a temperature of 1000 F.

Figure 24 shows these data graphically. The rat-tail producing sands can be readily distinguished from the good foundry sands on the basis of expansion and hot strength.

The prevention of casting defects due to cracking or spalling of the mold can be prevented by reducing hot strength and expansion of sand mixtures.

Hot strength (at 1000 F) can be reduced by the

addition of combustible materials such as seacoal, wood flour, cereal, etc. Hot strength can also be reduced by softer ramming.

Expansion characteristics can be modified by the use of different clays. Combustible materials will lower expansion of sand. Binders that increase the toughness of sands thereby reducing flowability which provides greater amount of void spaces between the sand grains.

#### APPENDIX I

##### *Influence of Flowability on Rat-Tails*

The property of flowability as measured in this study is that property of a sand which measures the ability of the sand grains to flow together so that adjacent grains touch each other. A molding sand which flows together so that adjacent sand grains touch each other would possess 100 flowability. All sands with a flowability above 82 produced castings with rat-tails. Sand with a flowability below 82 produced castings with and without rat-tails. Sands with high flowability thus require greater precaution than with low flowability. This relationship is shown in Fig. 25.

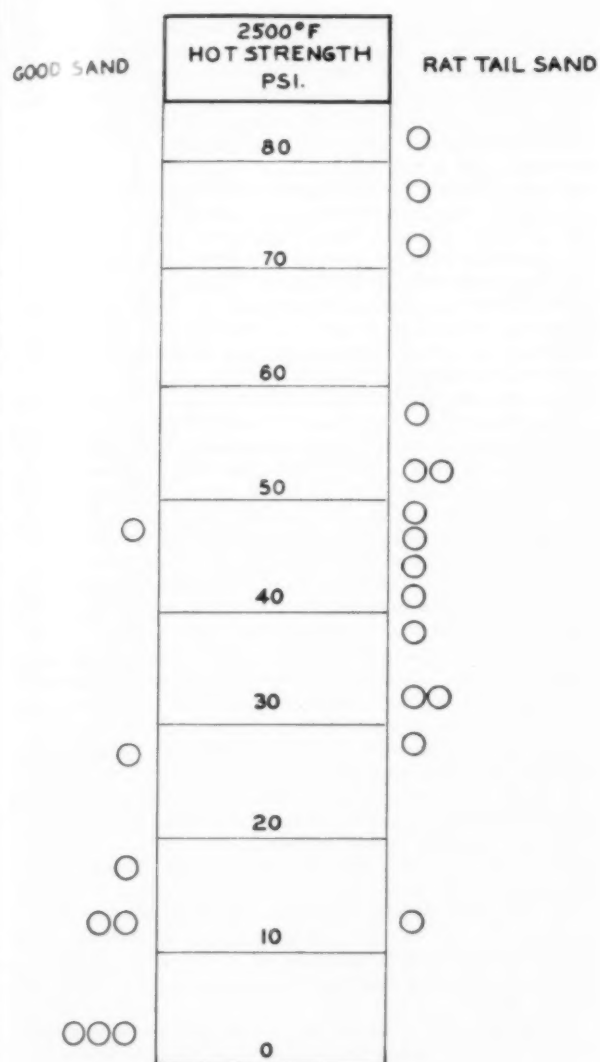


Fig. 23—Relationship between Hot Strength and Rat-Tail Defects.

A combination of high flowability, high hot strength and high expansion would produce a condition that would severely tax the face of the mold and tend to produce rat-tails.

#### Conclusions

- Two conditions that may cause rat-tail defects to appear on castings are:
  - Insufficient void spaces between contact faces of adjacent sand grains.
  - High mold stresses.
- The insufficiency of void spaces between the contact faces of adjacent sand grains may be caused by:
  - High mold hardness.
  - Fine sand.
  - High flowability.
- The development of high stresses at the face of the mold may result from:
  - Large flat mold surfaces which expand appreciably.
  - High expansion.

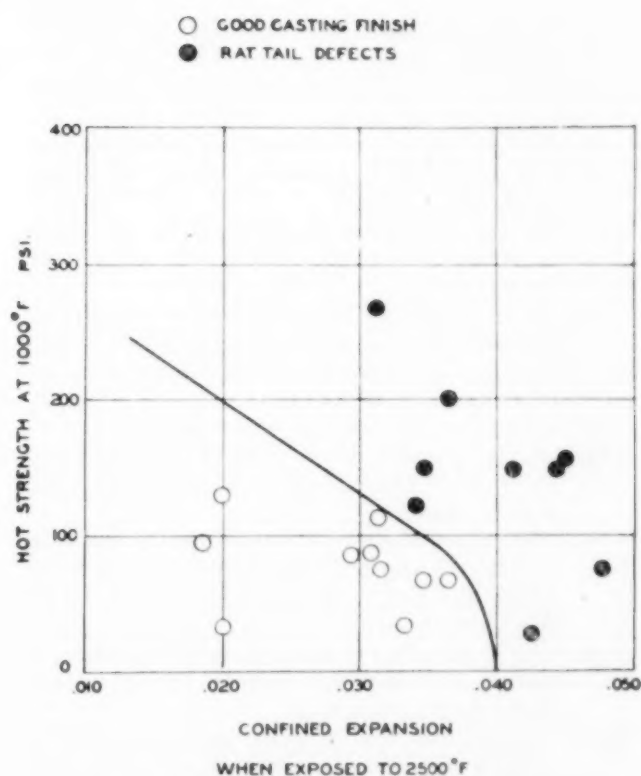


Fig. 24—Hot Strength and Expansion as Related to the Formation of Rat-Tails.

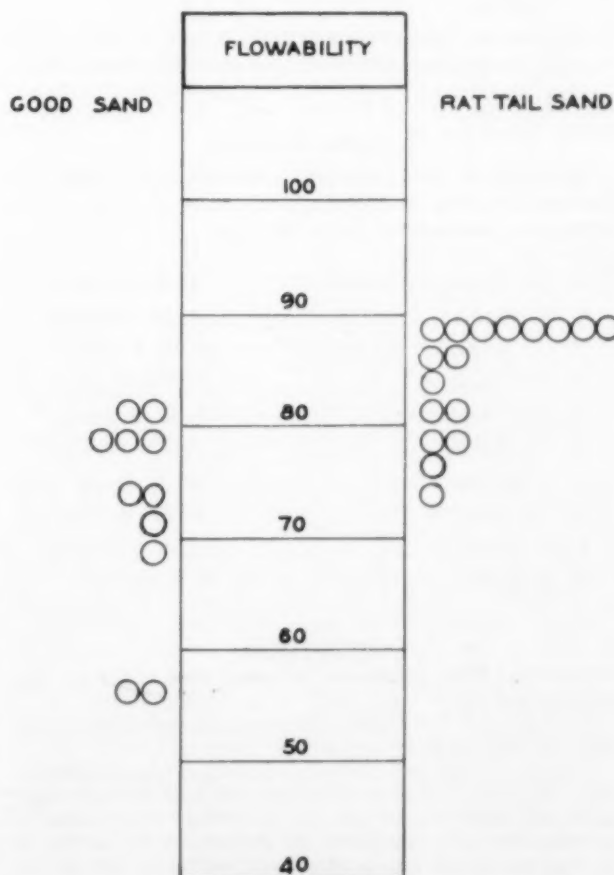


Fig. 25—Relationship between Flowability and Rat-Tail Defect.



- c. Insufficient combustible materials.
- d. High hot strength.
4. Methods which may be used to reduce or eliminate rat-tail defects on castings are:
  - a. Reduce mold hardness as far as practicable.
  - b. Use coarse sand providing finish of casting permits.
  - c. Reduce flowability of the sand by increasing toughness of the sand mixture, for example, by adding cereal.
  - d. Break continuity of flat areas by introducing grooves, circular depressions or other discontinuities, as reported by G. R. Gardner\* of the Aluminum Co. of America.
  - e. Reduce the hot strength of the sand.
  - f. Reduce the expansion of the sand as measured by a  $1\frac{1}{8} \times 2$ -in. specimen with a  $\frac{1}{2}$ -in. dia. hollow center.
  - g. The cause of a rat-tail defect may be stated as a break in the mold face due to compressive forces set up by high volume growth of a high hot strength sand.
  - h. Use in the sand mixtures controlled amount of combustible material, such as cereal, seacoal, wood flour or fine bran.
  - i. Control the hot strength and expansion by using a selected new sand or bonding material so that they fall within safe limits.
  - j. Select natural molding sand or correctly compounded synthetic sands which have low expansion.
5. The control of rat-tail defects as well as other casting defects can be maintained through sand control and selection.

#### Committee Personnel

Members of the Committee on Physical Properties of Iron Foundry Molding Materials at Elevated Temperatures consisted of the following:

H. W. Dietert, <i>Chairman</i>	Henri Louette
G. F. Watson, <i>Vice-Chairman</i>	R. W. Mason
W. A. Spindler, <i>Secretary</i>	C. S. Parsons
R. W. Bennett	E. Passman
Robert Doelman	F. B. Rote
H. H. Fairfield	Victor Rowell
J. A. Gitzen	W. N. Seese
Harry Gravlin	R. D. Walter
John Grennan	D. C. Williams
H. J. Jamieson	E. C. Zirzow

#### DISCUSSION

*Chairman:* J. A. RASSENFOSS, American Steel Foundries, East Chicago, Ind.

*Co-Chairman:* E. C. ZIRZOW, National Malleable & Steel Castings Co., Cleveland.

*CHAIRMAN RASSENFOSS:* My question pertains to grain distribution. The general conclusion was reached from this study that grain distribution has no effect on rat-tailing. That conclusion is rather odd, because generally the amount of fine materials in the sand would have a tremendous effect, on hot strength and on

confined expansion, or any expansion. Let us discuss expansion itself and for the moment not say how we measure it. Theoretically, we know that when silica is heated, it expands considerably and it is thus rather odd to find that grain distribution has no effect on the occurrence of rat-tailing. Do you think that this conclusion would be valid if grain distribution were varied over a wider range?

*MR. DIETERT:* Our work on grain distribution was of a limited nature. We were seeking the major reason for the rat-tail. I believe that if we had continued our work on grain distribution vs. rat-tail, we would have found that grain distribution does to a limited extent affect rat-tailing. It is not a major contributing cause. When you speak of fine material you are speaking of fineness. The Committee found that fineness does affect rat-tails. I believe the major contributing cause of rat-tailing has been found by the Committee's work.

*CHAIRMAN RASSENFOSS:* As I understand it, one of the main conclusions of the work is that rat-tailing can be correlated with hot strength as measured at the temperatures indicated and with confined expansion. As I recall, there were three consecutive double illustrations shown in which you had rat-tailing on the left picture and no rat-tailing on the right. In these instances the hot strength had gone down, the confined expansion had gone up. It seems rather odd that if the confined expansion test were the real measure of what is going on, an increase in the expansion of the sand will result in fewer rat-tails. What is the explanation for that?

*MR. DIETERT:* The amount of reduction in expansion was not sufficient to offset the increase in hot strength. A few thousandths, for example 0.005 in., is not a major change. One must reduce the expansion in the order of 33 per cent.

*CHAIRMAN RASSENFOSS:* The hot strength you showed in those instances was increasing, yet the confined expansion was decreasing. It does not add up very well. In some of those cases you added wood flour or some other combustible material. These additions certainly should not contribute to an increase in expansion, yet the expansion, as measured by this test, actually increased.

*MR. DIETERT:* The expansion tests referred to are confined expansion. In this test the specimen is practically in its own atmosphere with limited oxidation. Under these conditions, wood flour and cereal do not necessarily reduce expansion. They reduce hot strength.

*CHAIRMAN RASSENFOSS:* Maybe we are not measuring accurately what we want to know. It would appear from theorizing on the problem that, for instance, a sand mixture made up of sand grains and clay additions which would be employed to bond it would have a certain expansion value. Why, when he put in wood flour or cereal binders which would burn out during heating of the sand, should the expansion of the sand go up? It would seem logical that the expansion would decrease or remain constant. Certainly, it should not increase. It should show a decrease, particularly if there is a load on it, as there is in most of these tests.

*MR. DIETERT:* No, certain types of cereal bond do not decrease the confined expansion of sands, even though you would like to have them do so. There are one or two types that do show a slight increase. These confined expansion tests were made very carefully. The sand specimen is placed within a quartz tube which prevents rapid oxidation. Our final report will show the detailed method that was employed, and in no case did we have any conflicts. By plotting the hot compressive strength and confined expansion vs rat-tail, one finds that the castings with rat-tail will fall in a definite zone separate from those without rat-tails. One might devise a formula where the sum of hot compressive expansion plus confined expansion must be less than a certain number to avoid rat-tails. In practice it is easy to reduce the hot compressive strength of a sand to a reasonable value and not worry about changing the confined expansion which is a difficult property to reduce to one-half original value.

*CO-CHAIRMAN ZIRZOW:* There was a question in reference to a graph which showed an increase in confined expansion. I think the method or the accuracy of the test was questioned. As I recall the graph showed about 0.040 to 0.043 variation. That could be a permissible error.

*R. C. WALKER<sup>1</sup>:* We make a flat cover approximately 7 in. wide,

\* "Physical Properties of Molding Sands," TRANSACTIONS, American Foundrymen's Association, vol. 55, p. 332 (1947).

<sup>1</sup> Whitin Machine Works, Whitinsville, Mass.

10 to 12 in. long and about  $\frac{3}{16}$  in. thick. It is molded in a synthetic sand, bonded with about 5 per cent southern bentonite. We get rat-tails in the cope. If we face the mold with a strong synthetic sand bonded with 10 per cent southern and 6 per cent western bentonite, with low moisture content but high green compressive strength, we find that we do not get rat-tails. We have no means of measuring hot strength or expansion. What is the explanation for this? Why do we not get rat-tails with the increased amount of bond in the mold? Would we not have a higher hot strength?

MR. DIETERT: When you increase the bond of your sand, you also make the sand tougher and it will not ram down as firmly into the mold. You are obtaining larger void spaces into which the sand grains can slip as they increase in volume. If you would measure the hot compressive strength and the confined expansion at the temperatures that we recommended, you would find that the sum total of those two on this facing sand is less than the original sand.

MEMBER: In your diagrams showing effect of distribution of grain size, would it not have shown more if you had carried out further experiments with combination of Albany sand and synthetic sand to show effect of the scattering of the grains by adding a different kind of grain to the synthetic sand you already had? In other words, the combination of natural sand with synthetic sand would probably have shown an entirely different relationship than the wider distribution of strictly synthetic sand.

MR. DIETERT: Yes, it is possible that we could have used a combination of synthetic sand and natural sand and have altered our sands so that we would have had a lower hot strength and possibly lower confined expansion, or materially changed one and left the other one alone. Yet our work showed that grain distribution was not the main cause for rat-tailing. We were trying to determine the main cause of rat-tailing not all the slight variations in molding composition that would slightly alter rat-tailing. We were looking for major causes, and I believe we found them.

MEMBER: That is true, but foundrymen who make stove plate and sanitary ware know that there are a number of sands which are known to rat-tail and a few sands which are known not to rat-tail. There must be a difference in the inherent nature of certain sands to perform as such.

MR. DIETERT: That is true. It is common knowledge that many of our natural sands have a lower expansion than many of the compounded synthetic sands. When we add a natural sand to a synthetic sand, we probably derive some of the benefits of both, particularly in the lighter castings field where the grain is obtained from the natural sands. They are getting a bond from some synthetic bonds, such as bentonites or clays. Some natural sands cause rat-tail defects due to their hot strength. Other natural sands have low hot strengths and do not cause rat-tail to show on the castings.

J. B. CAINE: Did any of your test sands glaze?

MR. DIETERT: Yes, we had one sand that did glaze. It was an Albany sand. It definitely glazed, and it took quite a lot of sand-blasting to remove it.

MR. CAINE: Did it rat-tail?

MR. DIETERT: There were no rat-tails on that particular job.

MEMBER: Have you tried the sintering test on rat-tail?

MR. DIETERT: We tried the sintering test and every test we could think of but the only two that correlated with rat-tail were hot strength and confined expansion. The sintering test would pick out a few, but it would not pick out every one of them. You must consider both tests together if you want to be certain.

H. H. FAIRFIELD: In regard to expansion as affected by wood flours, extensive testing has shown that most wood flours and cereals do not decrease the expansion of sand. In fact, some of them give a slight increase, probably because they form a more rigid refractory structure and it can expand to its full extent. The test is made under a very light load. Without the wood flour, the sand may be a refractory that cracks and has actually a lower expansion under test. The main effect of wood flours and cereals is to lower the hot strength.

MR. DIETERT: We found the free expansion tests, where the specimen in the furnace is not supported, to give unreliable results. If the sand specimen spalls exceedingly or has low dry

strength, you will not measure true expansion. If the sand specimen is rammed within a quartz tube then more consistent results are obtained and the test data checks with casting results.

CHAIRMAN RASSENFOSS: Referring to Mr. Fairfield's remarks, it seems rather odd that you should coat the sand grain, the main expansion material in the mixture, with a material like cereal binder which burns up at the temperatures we are talking about and not obtain a decrease in the expansion of the sand. When you make a confined expansion test the material is enclosed in a tube. Why should increasing the thermal stability of the sand in a test like that tend to increase the expansion? It would seem that with a confined expansion test, you would always get the benefit of every bit of the expansion and the effect of thermal stability would tend to be erased. Mr. Dietert has pointed out that in a free expansion test, thermal stability is important and has a tremendous effect on the result, but it would not seem possible that it would be in a confined expansion test where, even if the sand specimen does crack, it cannot get out of the tube; it has no room to move; it can only expand outward as the tube expands and that would not be a great deal, because a low expansion fused quartz tube was employed.

MR. FAIRFIELD: The test we were making was under a very light load. If you were to make a test under a load of 5 lb or so, you would find that with the cereals and wood flours, your expansion would probably be less. The statement that the sand grains are coated with cereal or wood flour is not correct. The average mixing procedure does not distribute the material that uniformly, so you actually usually have grain to grain contact just the same. You simply create more void space so that the sand has lower strength.

CHAIRMAN RASSENFOSS: But does not creating more void space allow more room for expansion of individual grains and thus lower overall expansion of the sand?

MR. FAIRFIELD: The sand is expanding relatively freely, so that the effect of the wood flour cannot be observed. If you were to put a load on it and force the sand grains to readjust their positions under load while heating, you would find a lower expansion with the cereal.

D. C. WILLIAMS: I would like to have Mr. Fairfield explain more fully his statement that wood flour or cereal makes a more refractory structure.

MR. FAIRFIELD: By that I meant that it has less tendency to break up or form cracks when heated. It has better spalling resistance. We tested about 15 varieties of wood flour and wood flour substitutes carefully. For most of them there was no appreciable change in amount of expansion when they were added to molding sands.

H. A. ROTH: Your remark about the expansion is rather interesting. What type of grain did you use? Was it round or angular?

MR. FAIRFIELD: It was A.F.S. test sand.

MR. ROTH: That is a silica sand which would just expand. What difference in specific gravity did you find among the wood flours? Someone without testing would assume that a wood flour in a sand mold would be very fluffy. If wood flour is finely ground and can be immersed in water you will very likely find that it is heavier than the water.

My impression has been that wood flour would shrink in a mold and allow room for silica sand to expand. In a confined space, however, as when wood is destructively distilled, a lot of gas is given off along with tar, acetic acid and wood alcohol. Charcoal remains. It would be different in an oxidizing atmosphere. The sand becomes black does it not because the wood is not burned up?

Results from destructive distillation of hardwood show that 4000 lb of wood give up 1000 lb of charcoal, 70 lb of wood alcohol, 150 lb of acetic acid, 200 lb of tar and 7000 cu ft of carbon gases (*Organic Chemistry* by Fiesel and Fiesel). All dry wood substance has 1.56 sp gr while conifers have 50 to 80 per cent cavities (*Chandlers Encyclopedia*, vol. on Wood, Lumber and Timber).

MR. FAIRFIELD: These tests were made in a confined atmosphere, or an atmosphere reduced by the wood flour. The effectiveness of the wood flour was mainly due to its particle size. We did not measure specific gravity.

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# EFFECT OF MANGANESE-SULPHUR RATIO ON THE RATE OF ANNEAL OF BLACK-HEART MALLEABLE IRON\*

By

J. E. Rehder\*\*

## ABSTRACT

*In order to find the ratio of manganese to sulphur, in white cast iron suitable for making black-heart malleable iron, which will give minimum annealing time, two series of test bars were cast in which the manganese content was varied. One series contained 0.09 per cent sulphur, and the other 0.21 per cent sulphur, other constituents being in normal amounts used in commercial malleable foundries. Minimum first and second-stage isothermal annealing times were determined for each bar, and the results plotted versus manganese-sulphur ratio. It is shown that pronounced minima are produced in second-stage (sub-critical) annealing times by a suitable manganese-sulphur ratio, with less effect on first-stage annealing times. With higher sulphur content the minimum is in a narrower range of manganese than with low sulphur content, indicating that better control of manganese content is important commercially in high-sulphur white irons. The optimum manganese content is given by the formula  $\% \text{ Mn} = (1.7 \times \% \text{ S}) + 0.15$ .*

## Introduction

IT IS WELL KNOWN that in white cast iron for conversion into black-heart malleable iron, the sulphur and manganese contents must be suitably balanced in order to obtain minimum annealing cycles, since an excess of either manganese or sulphur decreases the rate of decomposition of cementite and thereby increases the time necessary for complete annealing. The amount of manganese necessary to compensate for or neutralize a given amount of sulphur is determined in practice by the use of one of various formulas, the particular equation used depending on the locality and experience of the user. There is appreciable variation in the results given by the formulas, and it would be of value to know which is correct. It is the object of this investigation, therefore, to determine under controlled conditions so that other factors will be constant, the ratio of manganese to sulphur in white cast iron that will produce the minimum annealing time. The importance of variation in this ratio is also sought.

It should be pointed out that such ratio, when

determined, is equally of value to gray cast iron foundries, so that the excess of manganese usually present in gray cast iron can be properly evaluated as to effect and importance.

## Previous Work

Documentation of the fact that cementite is stabilized thermally by an excess of either manganese or sulphur is not considered necessary in view of its wide acceptance. Manganese, when present in excess, is believed to combine with substantially all of the sulphur present to form manganese sulphide,  $\text{MnS}$ , which may contain dissolved iron sulphide,  $\text{FeS}$ , when the excess of manganese is low. Excess manganese over that present as sulphide is contained in the cementite as dissolved manganese carbide,  $\text{Mn}_3\text{C}$ , forming a double carbide of iron and manganese. As the excess of manganese is increased, the manganese content of the cementite is increased and the cementite becomes more stable towards decomposition.

The mechanism by which manganese stabilizes cementite is not known definitely, but manganese carbide has a higher heat of formation than iron carbide (cementite). It should be noted that with a given excess of manganese, decreasing the carbon content of the white cast iron will decrease the amount of primary cementite present, thereby increasing the manganese content of the primary cementite and increasing its thermal stability. Lower carbon white cast irons should therefore be more sensitive to manganese excess than will higher carbon white cast irons. This will be true of first stage annealing only, as the amount of cementite in the pearlite formed on cooling through the critical temperature will not be affected, and its manganese content should be constant with constant excess manganese present.

Sulphur, when in excess, combines with any manganese present to form manganese sulphide and the remainder forms an iron sulphide which may contain some manganese sulphide in solution. The mechanism by which the presence of excess sulphur stabilizes cementite thermally is not known.

Manganese sulphide is easily identified metallographically as a dove-gray non-metallic inclusion.

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Occasionally, well-formed cubic crystals may be seen. Iron sulphide is identified as a yellowish-khaki colored non-metallic inclusion, usually rounded in outline. According to Boyles<sup>1</sup> inclusions consisting of FeS-MnS solid solutions may be some intermediate color.

The ratio of manganese to sulphur that should theoretically produce pure manganese sulphide is 1.72 to 1, but it is well known that an excess of manganese over this ratio is necessary to produce maximum annealability in white cast iron, and mass law considerations show that when the manganese and sulphur are present as a minor constituent in iron, an excess of manganese is required.<sup>2</sup> The allowance made in practice for the amount of excess manganese is the point in which the above-mentioned formulas differ, as will be seen from the following commonly used equations.

$$\text{Mn} = (1.7 \times \text{S}) + 0.15$$

$$\text{Mn} = (2.0 \times \text{S}) + 0.10$$

$$\text{Mn} = 3 \times \text{S}$$

In gray cast iron practice the equation

$$\text{Mn} = (1.7 \times \text{S}) + 0.35$$

is recommended<sup>3</sup> as giving maximum softening effect.

The effect of sulphur and of manganese contents on the time necessary for annealing white cast iron containing 2.50 per cent carbon and 1.05 per cent silicon was determined quantitatively by Kikuta,<sup>4</sup> but no attempt was made to determine the ratio of manganese to sulphur at which minimum annealing times were obtained. With white cast irons containing 0.026 per cent sulphur, increasing manganese content was shown to increase first stage annealing time linearly from 5 hours at 0.22 per cent manganese to 7 hours at 1.01 per cent manganese, and thence very rapidly. The effect of manganese content on second-stage annealing time was not clearly shown, as both manganese content and temperature of anneal were varied simultaneously. With white cast iron containing 0.22 per cent manganese, increasing sulphur content was shown to increase first stage-annealing time linearly from 5 hours at 0.026 per cent sulphur to 11¼ hours at 0.152 per cent sulphur, and then very rapidly. In second-stage annealing, the time necessary was 10 hours at 0.026 per cent sulphur, 23 hours at 0.066 per cent sulphur, and no sign of any progress in second stage annealing after 22 hours, at 0.072 per cent sulphur. At a manganese-sulphur ratio of 3.06 then, second stage annealing was stopped entirely.

The equilibrium relationships in the iron-manganese-sulphur ternary system have been reported in some detail in the literature, and reviewed by Benedicks.<sup>5</sup> However, since they do not refer to quantitative effects on rate of graphitization, they will not be reviewed here.

#### Experimental Method

Briefly, the experimental technique was to make up two series of test-bars, one containing approximately twice as much sulphur as the other, in which the manganese content was varied; then to determine the minimum times in which first and second stage annealing could be carried out on these bars under uniform conditions. The primary object was to maintain all conditions as uniform as possible, from raw

materials through melting procedures to annealing, varying only the manganese-sulphur ratio. Deviations caused the melts or bars to be discarded.

Two lots of base iron were melted in a 50-lb. basic-lined high-frequency furnace and poured into one-inch diameter bars in green sand molds. The heats were made up from wash metal, ingot iron, and 50 per cent ferrosilicon, with sulphur added as elemental sulphur after melt-down to give approximately twice as much sulphur in one melt as in the other. The chemical analyses of the two lots of bars is given in Table 1. The sulphur contents given in this paper were by the combustion method.

TABLE 1—CHEMICAL ANALYSIS OF BASE IRONS

Element	LS	HS
Carbon, per cent	2.45	2.45
Silicon "	1.26	1.21
Manganese "	0.06	0.07
Sulphur "	0.098	0.205
Phosphorus "	0.06	0.06
Chromium "	less than 0.01	less than 0.01

Each lot of bars was cut into several 2-lb. 4-oz. portions, and these were melted in a small high-frequency furnace with suitable additions of high-carbon ferromanganese to give two series of melts with increasing manganese contents. Each melt was poured into four 6 x 1 x ¾-in. bars in open-top oil-sand cores. The furnace was lined with a sillimanite-grog mixture, and a blank heat was melted and poured to preheat and set the lining.

To make a melt, one third of the total base stock (about 12 oz.) was melted and the ferromanganese addition made. Another third of the base stock was immediately added and when the whole melted, a small addition (uniform in quantity throughout the heats) of 50 per cent ferrosilicon was made to increase the silicon content slightly. The remaining third of the base stock was then immediately added and the whole melted. The temperature of the melt was raised to 166 C (2912 F) as measured by a silica-sheathed platinum, platinum-rhodium thermocouple and the melt poured immediately into the baked sand molds.

Power input was uninterrupted throughout a melt, and was constant for each heat so that rate of melting and the time above the melting point was constant. The ferromanganese addition was made as described both to give maximum time for solution and mixing, and to avoid if possible the inoculating effect of high carbon ferromanganese additions, to molten white cast iron which the author has observed in commercial practice. Similar considerations apply to the ferrosilicon addition.

The test bars from each heat were broken out of the baked sand cores when black and placed in marked envelopes.

The following annealing procedure was used for each set of test bars. From each set of bars from a heat, one bar was sectioned with an abrasive cut-off wheel, using ample coolant, into several ¾-in. wide sections, discarding the top and bottom half inch.

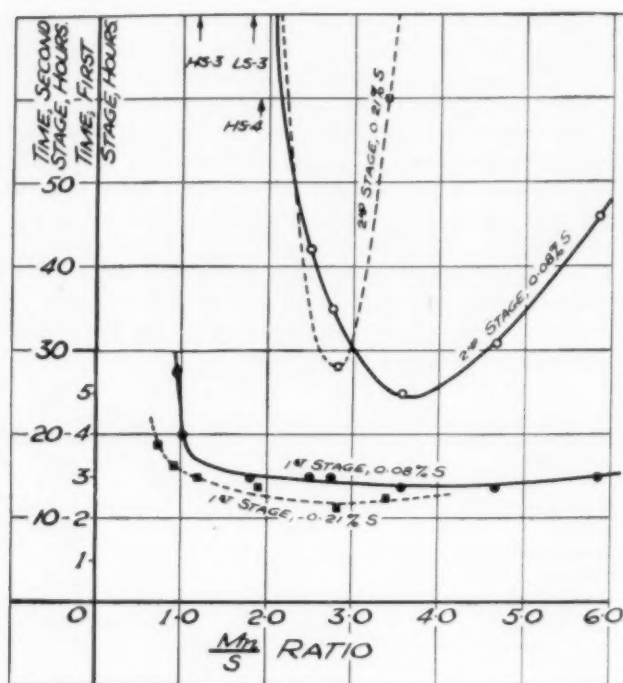


Fig. 1—Effect of manganese-sulphur ratio on the time necessary to decompose primary and pearlitic cementite in low and high sulphur content white cast irons.

Another bar was drilled for a sample for chemical analysis. The  $\frac{3}{8} \times \frac{3}{8} \times 1$ -in. sections and a full bar from the heat were placed in an electrically heated muffle furnace standing at 954 C (1750 F) and the chrome-alumel controlling thermocouple was placed in contact with the specimens. The amount of scaling was decreased by keeping a scorifier filled with coke in the back of the furnace chamber. The specimens

Fig. 3—Bar LS-1, after 48 hr at 1325 F. X100, 2% nital etch.

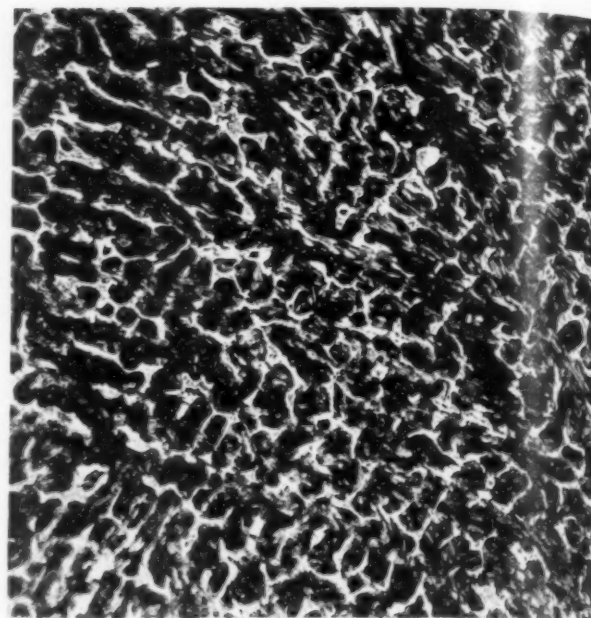
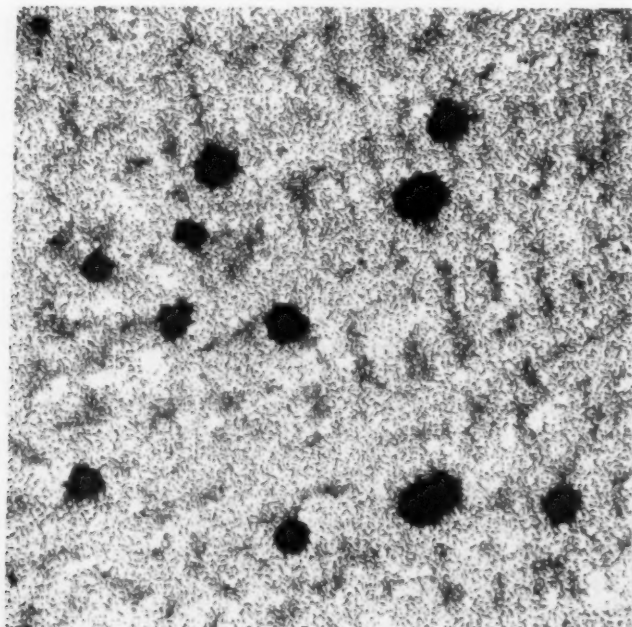
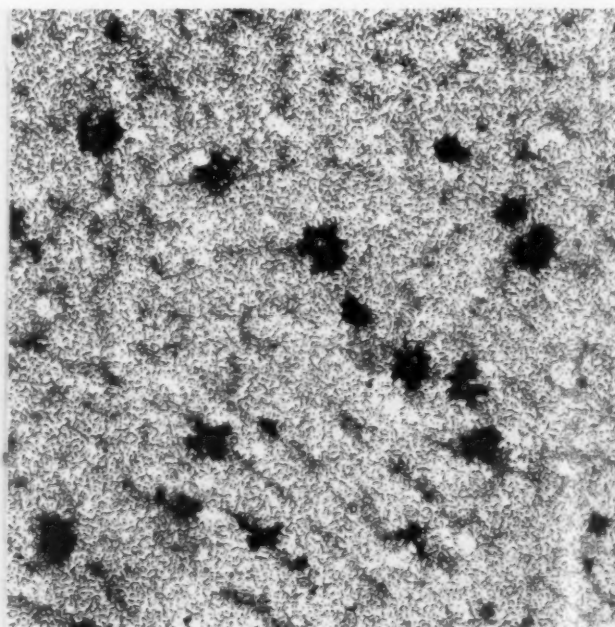


Fig. 2—Bar LS-5, as cast. X100, 2% nital etch.

TABLE 2—LOW SULPHUR SERIES: CHEMICAL ANALYSES AND ANNEALING TIMES

	LS-1	LS-2	LS-3	LS-4	LS-5	LS-6	LS-7	LS-8
Carbon, per cent	2.40	2.37	2.39	2.45	2.37	2.38	2.43	2.42
Silicon "	1.32	1.29	1.29	1.32	1.29	1.27	1.28	1.28
Manganese "	0.08	0.10	0.15	0.20	0.23	0.29	0.37	0.48
Sulphur "	0.083	0.090	0.083	0.080	0.084	0.081	0.079	0.082
Mn: S ratio	0.96	1.03	1.81	2.50	2.73	3.58	4.68	5.85
1st stage, hr.	5½	4	3	3	3	2¾	2¾	3
2nd stage, hr.	48+	48+	48+	42	35	25	31	46
Nodules/sq. mm.	26	20	21	30	41	39	38	41

Fig. 4—Bar LS2, after 48 hr at 1325 F. X100, 2% nital etch.



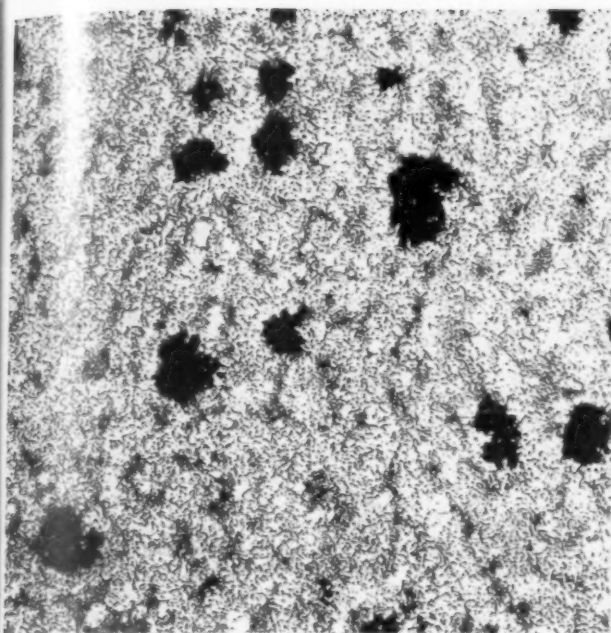


Fig. 5—Bar LS-3, after 48 hr at 1325 F. X100, 2% nital etch.

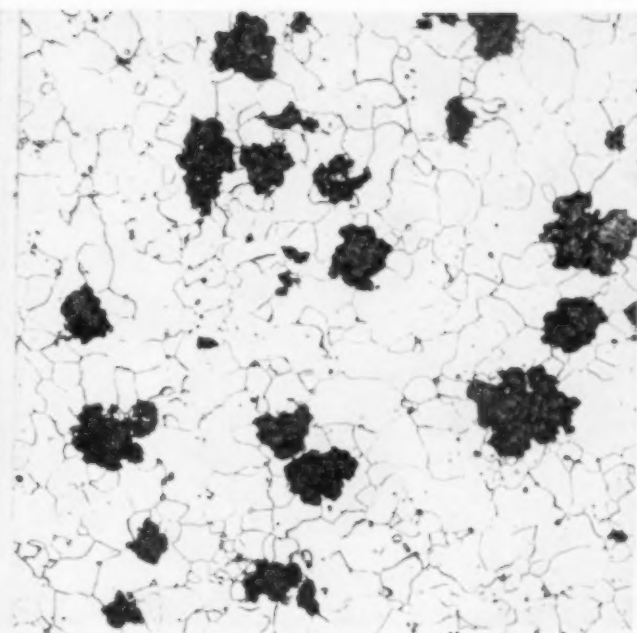


Fig. 6—Bar LS-4, fully annealed. X100, 2% nital etch.

TABLE 3—HIGH SULPHUR SERIES: CHEMICAL ANALYSES AND ANNEALING TIMES

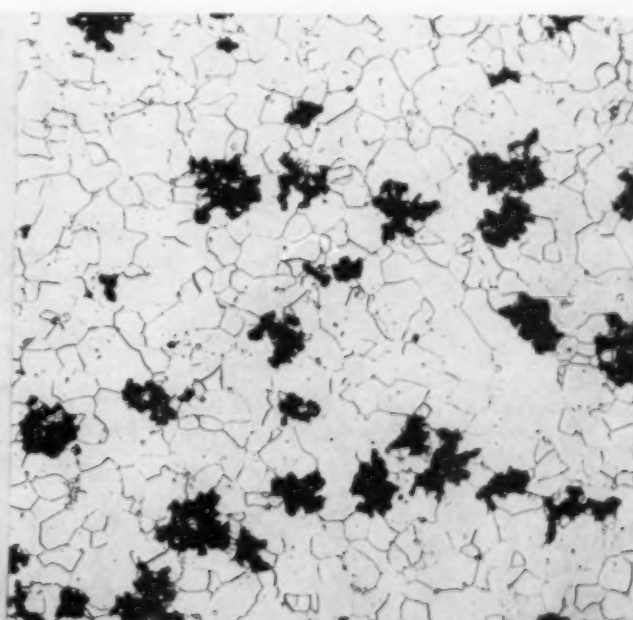
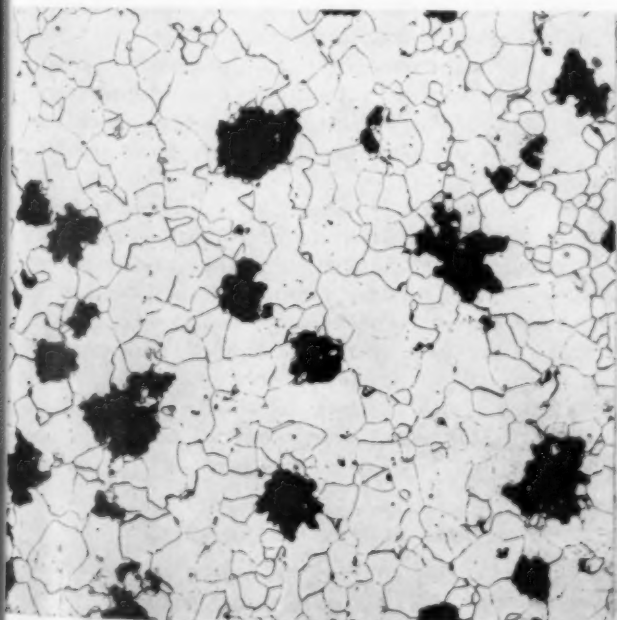
	HS-1	HS-2	HS-3	HS-4	HS-5	HS-6
Carbon, per cent	2.43	2.44	2.37	2.42	2.43	2.47
Silicon "	1.20	1.26	1.22	1.32	1.28	1.30
Manganese "	0.15	0.20	0.25	0.40	0.59	0.70
Sulphur "	0.205	0.215	0.210	0.211	0.209	0.206
Mn: S ratio	0.73	0.93	1.19	1.90	2.82	3.40
1st stage, hr.	3 3/4	3 1/4	3	2 3/4	2 1/4	2 1/2
2nd stage, hr.	48+	48+	48+	60+	28	60
Nodules/sq. mm.	36	24	34	38	41	52

TABLE 4—PERCENTAGE OF FeS OF TOTAL FeS AND MNS INCLUSIONS

Low Sulphur Series			High Sulphur Series		
Bar No.	Mn: S ratio	% FeS	Bar No.	Mn: S ratio	% FeS
LS-1	0.096	90-95	HS-1	0.73	30-40
LS-2	1.02	60-70	HS-2	0.93	15-25
LS-3	1.81	30-40	HS-3	1.19	10-20
LS-4	2.50	5-10	HS-4	1.90	Trace
LS-5	2.74	Trace	HS-5	2.82	None
LS-6	3.58	None	HS-6	3.40	None
LS-7	4.68	None			
LS-8	5.85	None			

Fig. 7—Bar LS-5, fully annealed. X100, 2% nital etch.

Fig. 8—Bar LS-6, fully annealed. X100, 2% nital etch.





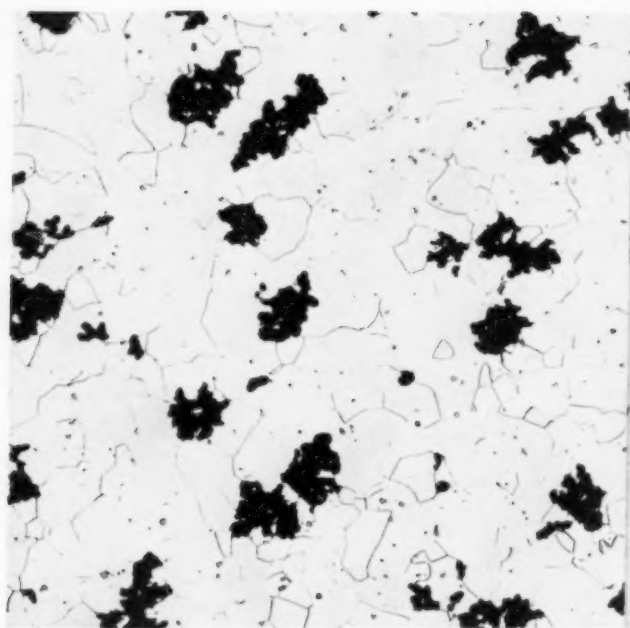


Fig. 9—Bar LS-7, fully annealed. X100, 2% nital etch.

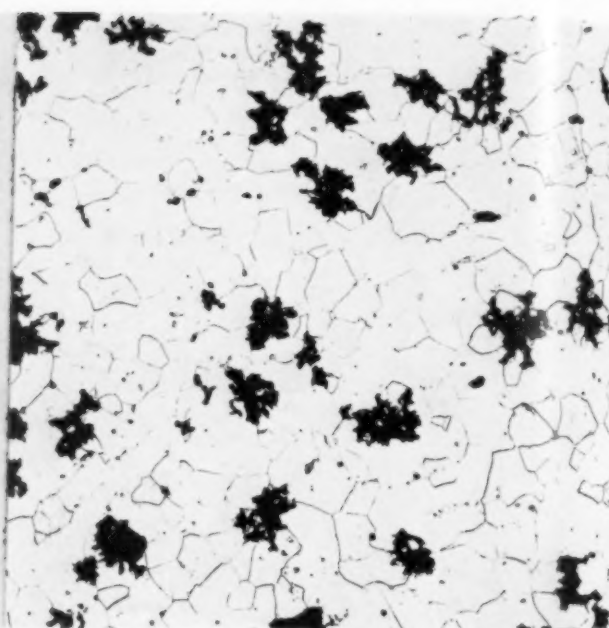


Fig. 10—Bar LS-8, fully annealed. X100, 2% nital etch.

came up to the control temperature of 954 C (1750 F) in about 15 minutes, and temperature control by an L and N recording and controlling potentiometer was  $\pm 2$  C (3.6 F). The same furnace, controller, and procedure were used for all samples.

Measuring time from when the samples reached control temperature, a  $\frac{3}{8} \times \frac{3}{8} \times 1$ -in. sample was removed every 30 minutes (and every 15 minutes as the end-point was approached), quenched in water, broken across the center, and the fractured face pol-

ished, etched, and examined under the microscope for primary cementite. The first stage of anneal was considered complete when no trace of primary cementite was visible at 500 diameters in five fields selected at random. The remaining samples, including the full bar, were then removed from the furnace and allowed to cool in air to room temperature.

The air-cooled full bar was cut into  $\frac{3}{8}$ -in. wide sections (discarding the top and bottom half-inch) on an abrasive cut-off wheel using ample coolant, and after bringing the muffle furnace to equilibrium at 718 C (1325 F), the sections were placed in the furnace and the control thermocouple arranged to touch the specimens. The specimens reached the control temperature of 718 C (1325 F) in about 10 minutes, and a sample was withdrawn every five hours (and every hour as the end-point was approached), air-cooled, sawn in two across the center, and the cut face polished, etched, and examined under the microscope for pearlitic cementite. The second stage of anneal was considered complete when no cementite was visible at 500 diameters in five fields selected at random.

The white cast irons were all examined at 100 and 1500 diameters in the etched condition, to note the arrangement of primary cementite and to discover the types and relative amounts of inclusions present. Estimates, based on thorough visual examination of each white iron specimen at 1500 diameters, were made of the relative proportions of iron sulphide and manganese sulphide type inclusions present. No primary graphite was visible in any as-cast bar.

## Results

### Chemical Analyses and Annealing Times

The chemical analyses and times for completion of first and second-stage annealing for the low sulphur

Fig. 11—Bar HS-1, after  $3\frac{3}{4}$  hr at 1750 F and quenching in water. Martensite, graphite, duplex MnS-FeS inclusions. X1500, 2% nital etch.





Fig. 12—Bar HS-6, as cast. Pearlite, cementite and MnS inclusions. X1500, 2% nital etch.



Fig. 13—Bar HS-1, as cast. FeS and duplex MnS-FeS inclusions in primary cementite. X1500, 2% nital etch.

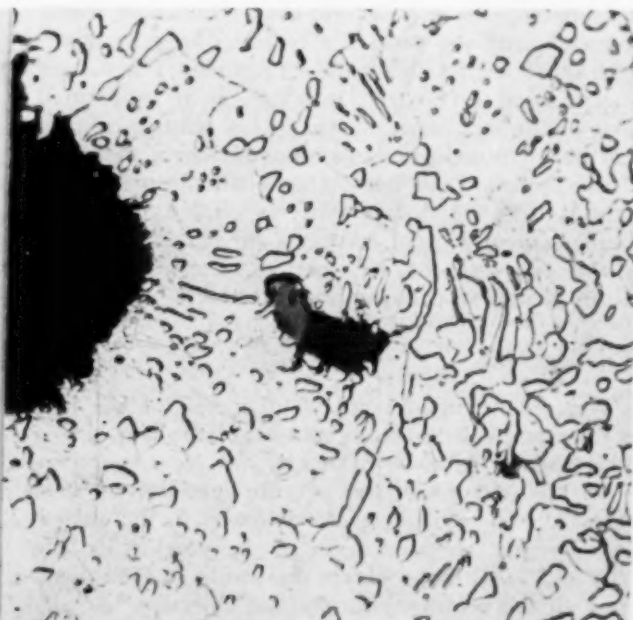
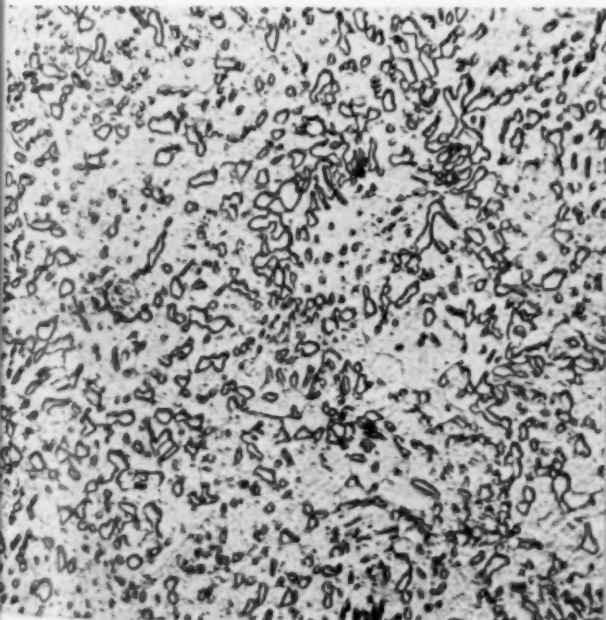
series of bars are given in Table 2, and the data for the high sulphur series in Table 3. Phosphorus and chromium contents were not reported since there was little chance for variation from the amounts present in the base irons. The notation "48+" in the times for second-stage annealing indicates that after 48 hours at 718 C (1325 F) no decomposition whatever of the pearlite could be detected even adjacent to temper carbon nodules. Complete annealing would probably require in these cases times of the order of hundreds

of hours. The pearlite had spheroidized, but not decomposed.

In the case of sample HS-4, most of the pearlite had decomposed in 22 hours, but a very fine spheroidized residue persisted and was still present after 60 hours at 718 C (1325 F). This residue is shown in Fig. 16. In the case of sample HS-6 small amounts of spheroidized pearlite persisted until 60 hours at temperature, but these spheroids were of normal size and eventually decomposed.

Fig. 14—Bar LS-3, after 48 hr at 1325 F. Spheroidized pearlite. X750, 2% nital etch.

Fig. 15—Bar LS-1, 48 hr at 1325 F. Graphite, pearlite, duplex MnS-FeS inclusions. X1500, 2% nital etch.



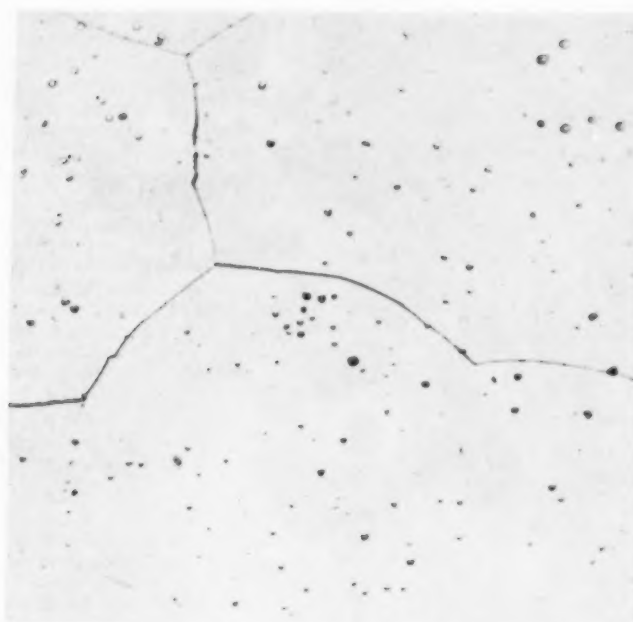


Fig. 16—Bar HS-4, annealed 60 hr at 1325 F. Fine spheroidized pearlite. X1500, 2% nital etch.



Fig. 17—Bar HS-6, annealed 48 hr at 1325 F. Graphite, MnS, spheroidized pearlite. X750, 2% nital etch.

Included in these Tables are temper carbon nodule counts, expressed as the number per square millimeter of actual surface. The counts were made at 100 diameters, taking an average of at least five fields in each case.

The chemical composition, other than manganese and sulphur, is seen to be relatively constant. The times necessary for first and second stage annealing are plotted versus the manganese sulphur ratio for both series of bars in Fig. 1.

In Table 4 are given the estimates made of the relative amounts of iron sulphide and manganese sulphide inclusions present in the white cast irons of the two series. Only two types of sulphide inclusions were noted, one a yellowish-khaki colored rounded inclusion, and the other the typical dove-gray, somewhat more angular manganese sulphide. The former was assumed to be iron sulphide, as it constituted most of the inclusions present in low manganese content bars. As manganese content increased the color of this inclusion did not change but more manganese sulphide was noted, frequently in duplex inclusions. As manganese content increased further, the amount of the iron sulphide decreased to zero, although no visible change in coloring was detected.

#### Metallography

The microstructures at 100 diameters of the bars as cast varied little throughout all of the bars, and a typical structure is shown in Fig. 2.

At 1500 diameters the pearlite laminations were observed to become more closely spaced as the manganese content increased, most of the pearlite in bars LS-8 and HS-6 being barely resolvable at 1500 diameters with an oil immersion fluorite objective.

In Fig. 3 to 10 inclusive are shown the microstructures of the LS series after completion of anneal. Bars LS-1 to LS-3 are shown after 48 hours at 1325 F.

In Fig. 11 is shown the microstructure of sample HS-1 after annealing at 954 C (1750 F) for 3¾ hours and water quenching. The background is lightly etched martensite and two duplex MnS-FeS inclusions are shown. It is interesting to note that graphite nucleation has apparently taken place preferentially at the MnS surface.

In Fig. 12 is shown two manganese sulphide inclusions enclosed in primary cementite in bar HS-6 as cast. One of the inclusions shows interference fringes sometimes seen in manganese sulphide. In Fig. 13 iron sulphide and duplex MnS-FeS inclusions enclosed in cementite are shown in bar HS-1.

It was mentioned above that in samples with low manganese-sulphur ratio, no detectable decomposition of pearlitic cementite took place in 48 hours at 718 C (1325 F). In Fig. 14 is shown the matrix, between temper carbon nodules, of bar LS-3 after this heat treatment. The pearlite is seen to have spheroidized. In bar LS-1, shown in Fig. 15, it is seen that pearlitic cementite extends up to the graphite nodule surface. A duplex MnS-FeS inclusion is visible in this latter figure, and spheroidization is less complete in this sample with lower Mn:S ratio.

The persistent spheroidized constituent in annealed sample HS-4 mentioned above is shown in Fig. 16. Although too fine to identify by etching tests, it is apparently spheroidized cementite. The spheroidized pearlite that was slow to disappear in second stage annealing of sample HS-6 is shown in Fig. 17, and is a normal spheroidized pearlite. It occurred mostly along ferrite grain boundaries, and eventually disappeared after 60 hours at temperature.



### Discussion of Results

It is evident in Fig. 1 that variation of the ratio of manganese to sulphur in these irons affects chiefly the second stage of annealing, or the decomposition of pearlitic cementite. Times for first-stage annealing are apparently not sensitive to this ratio, except for very low ratios, and a shallow minimum occurs approximately at the ratio of manganese to sulphur that gives minimum second-stage annealing time. It is interesting to note that the higher sulphur content irons complete the first stage slightly more rapidly than the lower sulphur content irons.

In both series of irons, second-stage annealing was apparently stopped completely when the manganese sulphur ratio was below about 1.9. This fact may be of value in the manufacture of pearlitic malleable iron, as first stage times are little affected. However, no determinations of mechanical properties were made in the present study, and these would have to be determined.

The fact that first stage annealing times are little affected below about 1.2 manganese-sulphur ratio suggests that the tendency to mottle as cast would be decreased only when this ratio is less than about 1.0.

As the optimum manganese-sulphur ratio is approached from below, second-stage annealing times decrease rapidly. When this ratio is exceeded, annealing times increase, but less rapidly. Apparently at higher sulphur content the curve is steeper, indicating that at higher sulphur content, control of manganese content must be more critical. The optimum manganese content for the LS series is given by

$$\% \text{ Mn} = (1.7 \times \% \text{ S}) + 0.15$$

$$\% \text{ Mn} = (1.7 \times \% \text{ S}) + 0.15$$

and this is probably true for the higher sulphur HS series, although a sample with this particular manganese content (0.51 per cent for the HS series) was not obtained. In the LS series, in order to obtain second stage annealing time within 10 per cent of the minimum possible, the manganese content would have to be in the range 0.26 to 0.34 per cent with 0.08 per cent sulphur.

Comparison of Tables 2, 3 and 4 shows that minimum annealing times (maximum rate of decomposition of cementite) approximately coincide with the disappearance of iron sulphide from the as-cast microstructure, as would be expected. The 0.15 per cent excess of manganese over that representing 1.7 times the sulphur content given in the above equation must therefore represent the excess necessary for mass law considerations.

Comparison of Tables 2 and 3 with Fig. 1 shows that minimum annealing times are not coincident with maximum nodule counts, and suggests that explanation of faster annealing rates as resulting from more numerous and closely spaced temper carbon nodules may be in error.

The microstructures of the annealed samples in Fig. 3 to 10 inclusive show that the temper carbon nodules depart farther from true spherical form as the manganese-sulphur ratio increases. The same trend was evidenced in the HS series. The grain size of

the ferrite in the fully annealed samples is approximately the same throughout.

### Conclusions

The following conclusions may be drawn from the data presented:

1. In white cast irons with 0.08 and 0.21 per cent sulphur, variation in the manganese-sulphur ratio between 1.8 and 5.0 causes little change in the times necessary for first stage annealing.

2. The time necessary for second stage annealing of white cast iron is sensitive to the ratio of manganese to sulphur, the sensitivity increasing with sulphur content.

3. The manganese content at which minimum annealing time is obtained in white cast iron is given by the equation

$$\% \text{ Mn} = (1.7 \times \% \text{ S}) + 0.15$$

4. For white cast iron containing 2.40 per cent carbon, 1.30 per cent silicon, and 0.08 per cent sulphur, manganese content should be controlled within the limits of 0.26 to 0.34 per cent to obtain annealing times within 10 per cent of the minimum possible.

5. The only change detected in the microstructure of fully annealed samples as the manganese-sulphur ratio increases is a trend towards more irregular shape of temper carbon nodules.

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### DISCUSSION

Chairman: G. VERNERHOLM, Ford Motor Co., Detroit, Mich.

Co-Chairman: W. B. McFERRIN, Electro Metallurgical Co., Detroit.

H. A. SCHWARTZ (*written discussion*): The paper constitutes an excellent systematic attack upon a subject of considerable practical interest. It represents a well organized detailed contribution to a field about which there exists much general knowledge and perhaps some prejudice, but in which there is little record of definitive experimentation.

It seems to this commentator that the assumption that all the manganese over that present as sulphide is contained in the cementite as manganese carbide, requires some further proof. There seems good reason to believe that some manganese is in the ferrite, and also it is extremely doubtful whether manganese dissolved in cementite does not merely represent the substitution of manganese for iron atoms in the cementite lattice.

Such spectrograms as the commentator has seen do not indicate the presence of a stranger lattice in white cast iron but only a combination of the patterns of alpha iron and cementite.

The author's statement that the mechanism by which the presence of excess sulphur stabilizes cementite thermally is not known, is perhaps literally correct. He has, however, omitted consideration of the film theory propounded by Levy. The commentator does not regard Levy's work as final proof of the existence of these microscopic or sub-microscopic films. However,

<sup>1</sup> Manager of Research, National Malleable and Steel Castings Co., Cleveland.

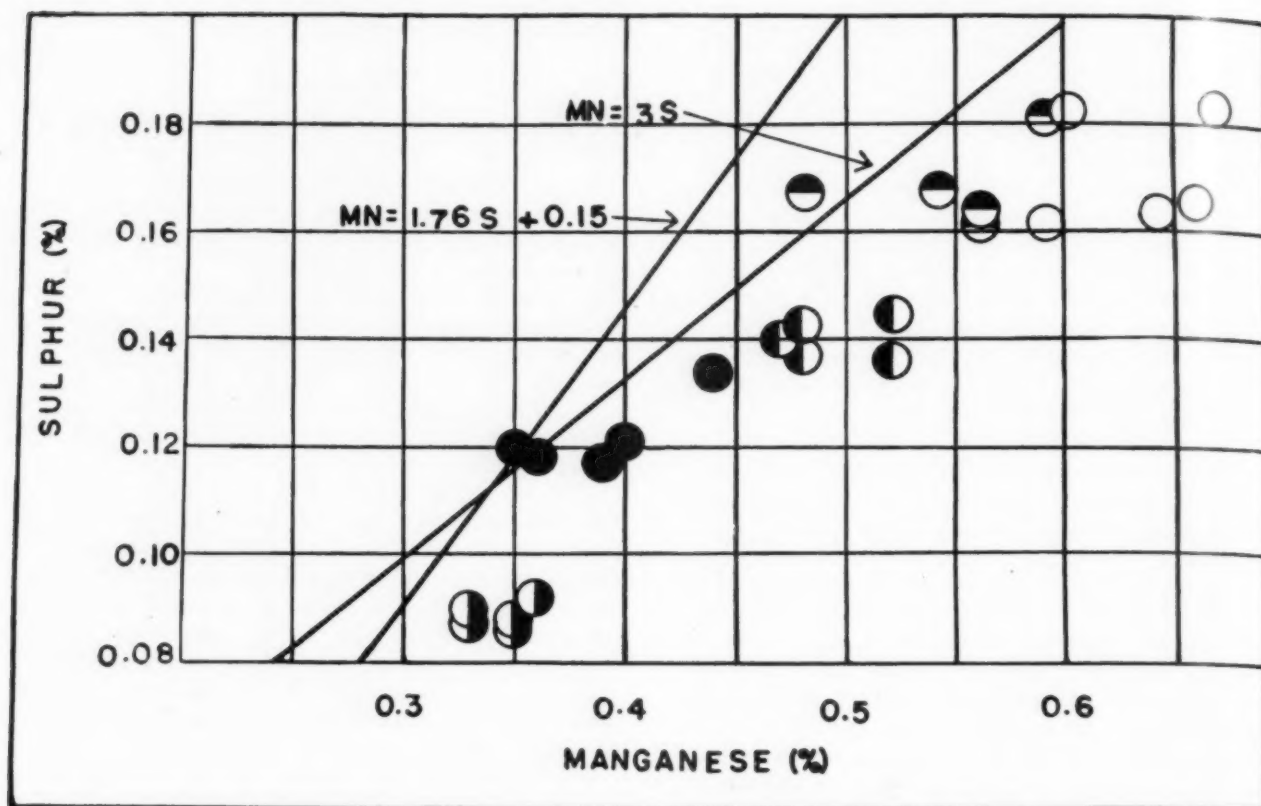


Fig. A

the behavior of elements which combine with sulphur in their effect on annealability bears a logical relation to their relative stabilities and the question whether their melting points are likely to be below that of white cast iron or above. The author does not appear to have confirmed Levy's observations of a somewhat laminar sulphide inclusion within cementite. Thus his work weakens any assumption based on Levy's observations. It may, however, be a little too soon to cast the older work completely into the discard since it might be that however excellent Rehder's metallographic technique seems to have been, some other technique might have brought out vestiges of films.

The author's experimental procedure seems to have been extremely well planned for the purpose intended. It should, however, be pointed out to the reader who may wish to use the annealing times determined by Rehder that the melting process adopted in the laboratory was one which, very probably, would be conducive to a fairly slow rate of annealing since the melting stock contains no graphite and the furnace lining was basic.

Rehder's precautions to get comparative results are so complete that there should be no hesitancy in accepting his relative conclusions.

The general form of Fig. 1 seems entirely consistent with previous observations. It is rather well known qualitatively that the effect of an excessive manganese in first-stage annealing is relatively slight. That there should be a minimum of heat treating time in second-stage annealing is also reasonable since in the beginning manganese neutralizes the effect of sulphur and beyond the minimum exercises its own retarding effect.

This commentator was originally one of the defenders of the idea that the best manganese content was 1.76 times the sulphur plus 0.015 per cent. This proportion, however, did not seem to give good results as commercial practice rose in sulphur content due to the advent of duplexing. It would see that the Mass Law referred to by the author would require that for a given small maximum permissible iron sulphide the manganese sulphur ratio would have to be constant if both elements are present in relatively small amounts. This does not appear to be

corroborated by the author's Fig. A.

Does the author feel that his work sufficiently establishes the facts so that the plausible results of the Mass Law should be disregarded? In other words, should we seek a better mechanism than the simple substitution of manganese for iron in the sulphide?

There is some evidence of practical experience that the control of manganese is more critical in high sulphur irons than in low sulphur irons, although the commentator does not remember ever to have seen this matter summarized in quantitative form.

The only criticism which can be made of so excellent a paper as the present is that perhaps two sets of alloys are insufficient to furnish complete confirmation of the perhaps unexpected departure from Mass Law considerations.

It has nothing directly to do with Mr. Rehder's investigation but a survey of commercial practice as to manganese and sulphur may perhaps be of some interest to the reader.

The writer has lately had occasion to compare the analyses of malleable iron made by five of the leading producers in the United States for one of the important automobile interests. That concern furnished to this laboratory, five castings from each of the five producers. (Only one of these was a National Plant.) The manganese-sulphur content of the malleable iron, the sulphur by combustion, was found to be as plotted in Fig. A. It will be noted that the manganese is somewhat higher than required by the conventional idea of a 0.15 per cent excess above that to form manganese-sulphide, and that the slope of these figures is more nearly that corresponding to a ratio of 3 to 1 for manganese and sulphur. It might be possible that a parabolic relation would have been preferable.

The only connection which these data have with the subject investigation is that commercial operations have a way of arriving, by trial and error, at desirable practices.

R. SCHNEIDEWIND (written discussion)<sup>2</sup>: Mr. Rehder has made a long needed contribution to the knowledge of malleable iron and has made quantitative a subject which before was known but empirically. Although well discussed in the text, the writer has not stressed in his conclusions the influence of the presence of iron sulphide. It is known that sulphur (or iron sulphide)

<sup>2</sup>University of Michigan, Ann Arbor, Mich.

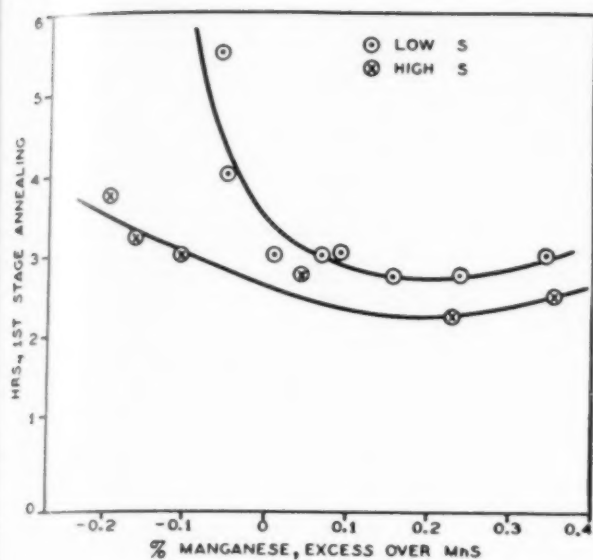


Fig. B

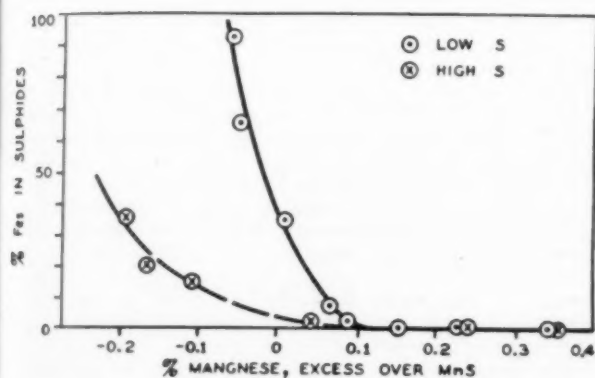


Fig. C

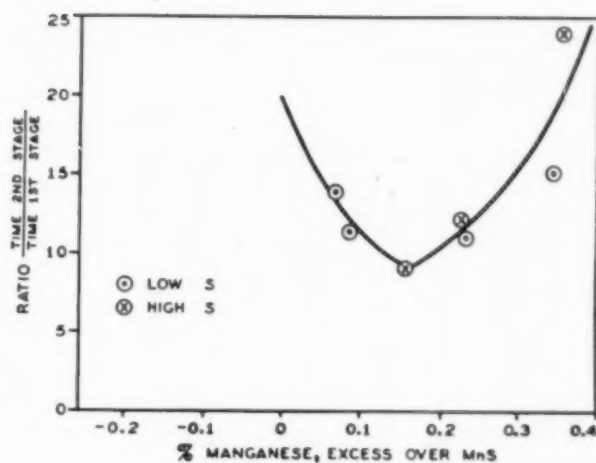


Fig. E

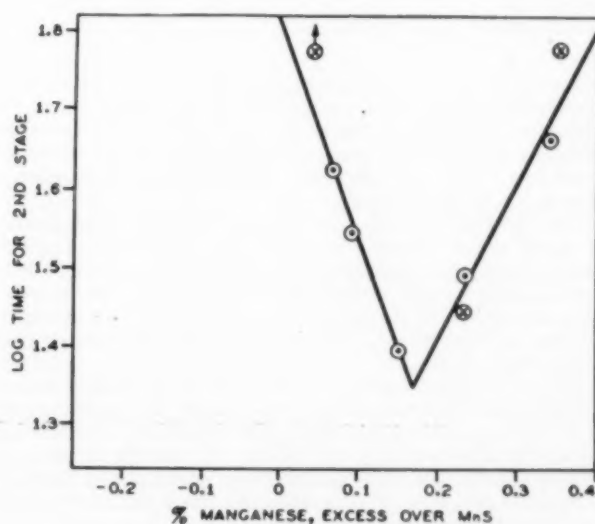


Fig. F

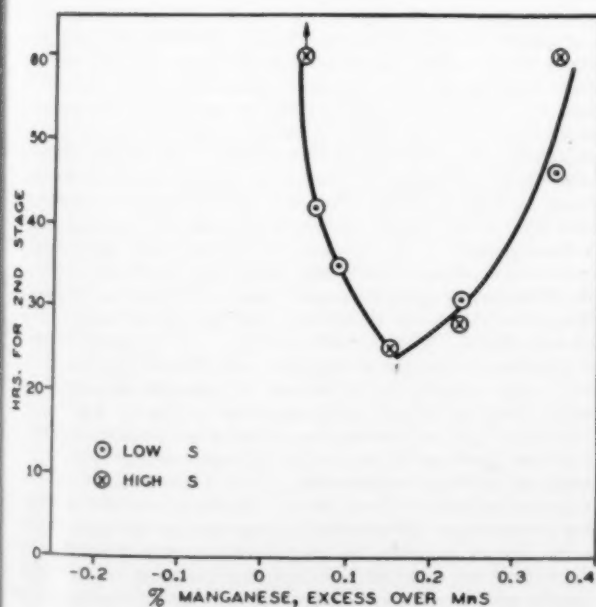


Fig. D

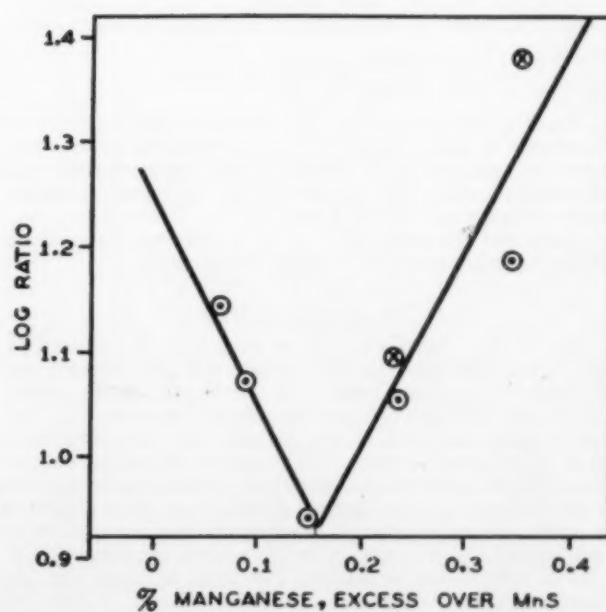


Fig. G



TABLE A

Iron	% S	Theoretical	Actual	Excess	1st	2nd	2nd	2nd	Log	
		Mn.					Stage			Stage
		(1.7)					Stage			1st
	% S	(% S)	Mn.	%	hours	hours	Stage	hours	Ratio	
LS-1	.083	.141	.08	— .061	5.50	48+	.....	.....	.....	
LS-2	.090	.153	.10	— .053	4.00	48+	.....	.....	.....	
LS-3	.083	.141	.15	+ .009	3.00	48+	.....	.....	.....	
LS-4	.080	.136	.20	.064	3.00	42	14.0	1.624	1.146	
LS-5	.084	.143	.23	.087	3.00	35	11.7	1.544	1.068	
LS-6	.081	.138	.29	.152	2.75	25	9.1	1.398	0.959	
LS-7	.079	.134	.37	.236	2.75	31	11.3	1.492	1.054	
LS-8	.082	.139	.48	.341	3.00	46	15.3	1.664	1.185	
HS-1	.205	.349	.15	— .199	3.75	48+	.....	.....	.....	
HS-2	.215	.366	.20	— .166	3.25	48+	.....	.....	.....	
HS-3	.210	.357	.25	— .107	3.00	48+	.....	.....	.....	
HS-4	.211	.359	.40	+ .041	2.75	60+	.....	.....	.....	
HS-5	.209	.356	.59	.234	2.25	28	12.4	1.477	1.094	
HS-6	.206	.350	.70	.350	2.50	60	24.0	1.778	1.380	

and manganese each have a retarding influence on second stage graphitization and a less marked one on first stage. Hence we can reason that the optimum amount of manganese is that necessary to form MnS. Less manganese than this permits the formation of FeS; an excess brings about stabilization of pearlite due to the manganese itself.

Mr. Rehder shows that the optimum quantity of manganese to insure the absence of FeS is somewhat in excess of the theoretical figure. The suggestion is made that the results of these experiments may be more clearly brought out if the deviation in concentration of manganese from the proportions expressed as MnS are computed as has been done in the accompanying table (Table A).

By plotting the excess (or deficiency) of manganese against time for first stage annealing it may be seen in Fig. B that the best results are obtained, not at the theoretical concentration of manganese but with a slight excess. By plotting the per cent FeS in the sulphide phase against excess manganese as in Fig. C it may be seen that the amount of FeS drops to zero when the per cent excess manganese is between 0.1 and 0.2 per cent.

In order to locate the point of excess more accurately, Fig. D can be plotted with per cent excess manganese versus hours for second-stage annealing. A definite minimum point in the curve shows at 0.16 per cent excess manganese. The same minimum points is obtained by plotting excess manganese versus the ratio

2nd Stage annealing time  
1st Stage annealing time as shown in Fig. E.

Another way to illustrate the consistency of Mr. Rehder's fine data is to replot Fig. D and E using logarithm of annealing time and logarithm of the annealing time ratios. Straight lines are obtained which very sharply indicate the optimum manganese concentration. See Fig. F and G.

I feel that this paper is the result of much careful work and should be of great value to the malleable industry.

#### MR. REHDER REPLIES

MR. REHDER (*Author's Written Reply to Messrs. Schwartz and Schneidewind*): The kind comments of Dr. Schwartz and Dr. Schneidewind are noted and appreciated by the author. Their discussion contributes materially to the paper.

Concerning the note of Dr. Schwartz that "the assumption that all the manganese over that present in the sulphide is contained in the cementite as manganese carbide, requires further proof," it was not the author's intention to imply that such was the case, though the text should have been more specific in this respect. In the opinion of the author, manganese in excess of that present as manganese sulphide in white cast iron is present both in carbide and in ferrite, with the relative proportions probably given by a partition coefficient. Recent work of Dr. Schwartz indicates this to be the case.

The author agrees with Dr. Schwartz that small amounts of manganese dissolved in iron carbide are probably present as manganese atoms substituting for iron in the cementite lattice. The points at which the lattice would become a double manganese-iron carbide or a manganese carbide with some substituted iron atoms (as manganese content increases) are not known to the author, but an increase in thermodynamic stability with manganese content would seem probable.

Concerning the mechanism by which the presence of excess sulphur stabilizes cementite thermally, the interpretation is correct that the author does not believe that the mechanism is established or agreed upon. The film theory of Levy is a reasonable one, and as Dr. Schwartz said, is lent strength by analogy. However, film theories suffer the defect that in most cases the thickness of film theoretically necessary to be effective is much less than can be seen microscopically. In no case, even with most careful technique, has such a film been detected by the author even in very high sulphur content irons. This does not prove their absence however. The most noticeable effect adjacent to the carbide in high sulphur content irons with low manganese content, is a broadening of a band of ferrite separating pearlite from carbide masses.

Dr. Schwartz mentions that the melting process used may be conducive to a fairly slow rate of anneal partly due to the use of a basic furnace lining. It should be pointed out that only the base stocks were basic melted, and that the remelts on which annealing cycles were determined were carried out in an all acid-lined furnace. It is not felt that the annealing times found, especially first stage, are unduly long.

Concerning Mass Law effects, it was noted in the paper that minimum annealing times coincided roughly with the disappearance of iron sulphide from the matrix, and this could represent the point of given small permissible excess iron sulphide. The author definitely does not consider that the Mass Law is proven inapplicable to this case by the present data or by any other he has seen, but it is noted that the best-fitting line drawn through the points in Dr. Schwartz's Fig. A would be parallel to and to the right of the line representing manganese equal to three times the sulphur content. This indicates a formula approximately

$$\% \text{ Mn} = (3 \times \% \text{ S}) + 0.06$$

and the presence of the constant apparently does not corroborate Mass Law effects.

The chart shown by Dr. Schwartz in Fig. A, representing practices of five producers of malleable iron, is interesting and revealing. The author agrees that commercial operations have a way of arriving by trial and error, at desirable practices, but where sulphur content is concerned, it is probable that some history is concerned. As we all know, some years ago sulphur was considered to be the root of most evils in both gray iron and malleable foundries, and it was believed that the lower the sulphur content the better was the iron. Those ideas have since been revised, but many people still like to be on the 'safe side' by using a slight excess of manganese. By the same process of trial and error mentioned above, it was found that in general an excess of sulphur caused more trouble than the same excess of manganese, and the results given in the present paper corroborate this finding, but this simply strengthened the idea that it was better to be on the safe side in manganese.

The difficulty here is that there was some difference of opinion as to how much manganese in excess of theoretical was necessary to give minimum annealing times, as shown by the various common formulae given in the first part of the present paper. Each foundry added its own safety factor to the minimum amount of manganese it believed necessary, and the result is shown in the considerable spread of points, especially at higher sulphur contents, shown in Fig. A. For example at 0.16 to 0.17 per cent sulphur, manganese contents of 0.48 to 0.66 per cent are being used, a variation of 0.18 per cent of manganese. This variation represents 31 per cent of the total manganese present. Each plant represented on the graph in Fig. A obviously believes that it is using the optimum amount of manganese necessary to neutralize the effect of sulphur, but if any one of these plants showed a variation in its melting control of 18 points of manganese, the matter would be considered serious. The graph shown in Fig. A is, therefore, interesting, but is believed to be of little value in establishing a

relationship between manganese and sulphur contents that will give minimum annealing times.

With respect to Dr. Schneidewind's remarks, the author is grateful for the work done by Dr. Schneidewind in preparing the tabulations and graphs in Table A and Fig. B to G inclusive. These graphs show more clearly than does the one given in Fig. 1 of the paper, the fact that in this work, minimum annealing times were found when the excess of manganese over the amount theoretically necessary to form manganese sulphide was close to 0.15 per cent.

The amounts of iron sulphide present in the matrix of each sample were estimated visually, and it is encouraging to see in Fig. C that the results are consistent. Where iron sulphide is specified in Table 4 of the paper as none, there was none visible during lengthy search of the microstructure with a viewing camera, and this occurred only when the manganese excess was 0.15 per cent or higher. It is probable that some small decreasing amount of iron sulphide is dissolved in the manganese sulphide as manganese exceeds an excess of 0.15 per cent, but it is no longer visible as a separate phase.

### MR. JOSEPH'S DISCUSSION

C. F. JOSEPH (*Written Discussion*):<sup>2</sup> The paper by Mr. Rehder on sulphur-manganese ratio for malleable iron should be of interest to the foundry industry.

The proper manganese-sulphur ratio for commercial malleable iron is a debatable question and has been the subject of much discussion at technical meetings for many years. To definitely state certain limits for manganese for a given sulphur analysis, is bound to lead to much argumentation.

Additional factors influencing the sulphur-manganese ratio should be of interest.

The method of manufacture of the malleable iron casting, such as the melting process used and the annealing temperature employed, have a bearing on the manganese-sulphur ratio. In turn, the melting process used, together with the raw materials, such as malleable pig iron, silvery pig iron, nature and quality of steel scrap, fuel and many other variables, will have a pronounced effect on the chemical composition of the casting produced.

Reference is made particularly to the carbon and silicon analysis. Malleable iron is produced with a total carbon plus silicon as low as 3.00 per cent and as high as 4.30 per cent. This allows the silicon in the casting to vary from 0.80 to 1.75 per cent. This wide variation in analysis calls for a different manganese-sulphur ratio, even though sulphur is high or low.

Manganese and sulphur have a strong affinity for one another; so much so, that in the presence of moderate excess of manganese, practically all the sulphur present occurs as manganese sulphide,  $MnS$ . This sulphide has a rather high melting point, 2930 F, and is practically insoluble in molten iron and separates out as small globules. Due to their low specific gravity, these globules tend to float to the surface of the molten metal, which affords an opportunity for sulphur removal by suitable slags. Manganese sulphide occurs as small particles of dove-gray color. Any excess sulphur present over manganese-sulphide forms an iron sulphide,  $FeS$ —with a melting point of 2125 F. Even after a drastic anneal with high sulphur, up to 0.80 per cent of combined carbon is often retained in the iron.

Sulphur uncombined with manganese has a powerful stabilizing action on the iron carbide. Manganese sulphide formation, therefore, prevents the sulphur from exerting its normal hardening influence. Manganese-sulphide is composed of 55 parts manganese and 32 parts sulphur, or about 1.7 parts manganese to 1 part sulphur. However, if the reaction is to be complete, an excess of manganese is required.

Just what this excess should be is a debatable question. Some investigators claim the manganese should be twice the sulphur, plus 0.15, plus or minus 0.03 per cent. Others say the manganese should be not less than three times the sulphur. Using these two rules for an 0.06 per cent sulphur, would give 0.24-0.30 per cent by the first method and a minimum of 0.18 per cent manganese by the second method. General experience in most plants indicates a sulphur of 0.06 per cent calls for a manganese of 0.26 to 0.31 per cent, which agrees very closely to the first method.

Taking a higher sulphur of 0.12 per cent, the first method of twice the sulphur, plus 0.15, plus or minus 0.03 per cent would give 0.36-0.42 per cent. The second method of a minimum of three times the sulphur is 0.36 per cent manganese. General experience calls for a 0.12 per cent sulphur to carry 0.36-0.42 per cent manganese, and hence, both methods are accurate.

Taking a still higher sulphur of 0.20 per cent, the first method of twice sulphur plus 0.15 would bring the limits 0.52-0.58. The second method, a minimum of three times the sulphur would be 0.60 per cent. General experience for this sulphur is rather conflicting. One company operating a cupola air furnace melting unit, producing iron with a sulphur content of 0.20-0.25 per cent, is operating with 0.50-0.55 per cent sulphur, 0.60 per cent is regarded as being too high. This company produces a high carbon iron, around 3.00 per cent with a silicon of 0.85 per cent and pours light castings.

Another company operating with a 0.20-0.25 per cent sulphur using a cupola air furnace melting unit, produces automotive and railway equipment castings. The carbon is held around 2.50 per cent, silicon 1.15. This company obtains the best results with a manganese of 0.60-0.65 per cent.

Other exceptions have been noted by numerous other foundries. One foundry running 0.06-0.08 per cent sulphur operates on a 0.30-0.35 per cent manganese. Another foundry running a duplexed iron with 0.10-0.12 per cent sulphur prefers to run the manganese 0.42-0.46 per cent and 0.48-0.50 does not give any trouble. It can be generally stated that most foundries prefer to keep the manganese on the high side of the accepted rules, rather than the low side. There is by far a lesser tendency to produce pearlitic rims and improve annealability if manganese is kept on the high side.

Regarding the maintaining of manganese on the high side, the company the writer is connected with produces a cupola-electric duplexed iron running around 0.11-0.13 per cent sulphur. About twenty years ago we operated on a 2.80 carbon, 1.00 silicon iron with 0.12-0.15 sulphur, and a manganese of 0.35-0.38 gave us good results. Forty per cent seemed too high. Later we lowered the carbon to 2.70 per cent and increased the silicon to 1.20-1.25 and we still maintained the sulphur at 0.12-0.15 and found a slightly higher manganese 0.36-0.42 was preferable. About seven years ago, we lowered our carbon to 2.60 per cent and increased the silicon to 1.35-1.45 per cent, sulphur running lower, 0.11-0.13 per cent, and we are now operating with a manganese of 0.42-0.47 per cent. This higher manganese has helped the annealability of the iron and we have noted a slight increase in the physical properties. In cases of break-downs or other unavoidable delays, if lower silicon irons are produced, the higher manganese irons have a better chance of producing a satisfactory annealed iron. In our case, 0.50 manganese is much preferred to 0.37 per cent with either a 1.25 silicon or a 1.45 silicon iron.

The purpose of this discussion is to emphasize the importance of different analysis of carbon and silicon, melting processes, raw materials and possibly metal thickness or casting weight in relation to the proper manganese to use with a given sulphur. In reality, there is no fixed ratio which will hold good for all conditions. Commercial operations will determine what combination of manganese and sulphur is necessary to obtain the most desirable casting properties and structure.

### MR. REHDER REPLIES TO MR. JOSEPH

J. E. REHDER (*Author's Written Reply to Mr. Joseph*): The comments and discussion of Mr. Joseph are very welcome, since they are based on extensive commercial experience.

The author agrees with Mr. Joseph that variables such as the source of charge materials, the materials themselves, the type of melting unit used, and the method of operating the melting equipment, all may have an effect on the manganese-sulphur ratio. However, it is difficult to see how these variables would affect the optimum ratio of manganese content to sulphur content to give minimum annealing time, although it would be unwise to state that they do not have an effect on this optimum ratio.

As Mr. Joseph has stated, and as shown in the paper, an excess of manganese over the amount theoretically necessary to form manganese sulphide, is necessary to ensure absence of

<sup>2</sup> Central Foundry Div., General Motors Corp., Saginaw, Mich.

iron sulphide from the structure and to provide minimum annealing times. It is noted that in examples of 0.06 per cent sulphur iron and 0.12 per cent sulphur irons given by Mr. Joseph, that the rule for manganese-sulphur ratio given in the paper does seem to correspond with general experience for minimum annealing times.

In the example given for sulphur content of 0.20 to 0.25 per cent, it is agreed that there is considerable difference of opinion in practice as to the optimum amount of manganese necessary, but it is suggested that there are two reasons for apparently conflicting opinions.

These are that in the higher sulphur content irons, variations in manganese or sulphur content away from balance have sharper effects than in lower sulphur content irons, as shown in the paper; and errors or differences in analytical techniques and results become correspondingly more important. The author does not consider the evolution method for sulphur reliable in white cast irons with sulphur contents of 0.20 per cent and over, and errors in this determination will create wide apparent differences in the optimum amount of manganese necessary to neutralize the effect of sulphur on annealing rate. For these reasons, comparisons between different foundries may be valueless unless chemical analyses are checked independently by the same methods. Another reason for apparent discrepancies in manganese-sulphur formulas between foundries is that some foundries have better control over their annealing cycles than have others, so that poor annealability in the latter may be due to factors other than manganese-sulphur ratio.

In the author's own experience in operating a cupola-air furnace duplexing unit on Grade 35018 malleable iron, during a period of several weeks of high-sulphur coke due to a fuel shortage, sulphur content of the iron ran 0.20 to 0.23 per cent. Manganese content was kept in the range of 0.50 to 0.55 per cent, and no trouble was encountered in annealing. In fact, the average of the tensile test results improved slightly.

The author agrees with Mr. Joseph that most foundries prefer to keep the manganese on the high side of the accepted range, rather than the low side. This point was brought out in the reply to Dr. Schwartz's discussion above.

Concerning Mr. Joseph's interesting description of changes in sulphur and manganese contents over the years as carbon was decreased and silicon increased in the iron at Saginaw Malleable, it is noted that on the 2.80 carbon 1.00 silicon iron, with sulphur at 0.12 to 0.15 per cent, the manganese content found most suitable coincided almost exactly with the formula

$$\% \text{ Mn} = (1.7 \times \% \text{ S}) + 0.15.$$

With 2.70 carbon and 1.20 to 1.25 silicon iron, the manganese-sulphur ratio found suitable again coincided with the above formula within 0.02 per cent manganese. Considering differences in manganese content determinations in different laboratories, this is believed to be a good check.

With 2.60 carbon and 1.35 to 1.45 silicon iron, the manganese content being used runs 0.08 to 0.10 per cent higher than the value given by the above formula. No explanation is evident for this.

With respect to effect of carbon and silicon contents, some data from gray iron practice may be of interest. It was shown in the paper that minimum annealing times were obtained when free iron-rich sulphides were just removed from the microstructure, this requiring an excess of manganese over theoretical MnS of 0.15 per cent. When free iron-rich sulphides were visible in the microstructure, rate of anneal (rate of graphitization) was much retarded. Several samples of relatively high sulphur content cast irons have come to the author's attention, from different foundries, and with sulphur contents of from 0.18 per cent to 0.26 per cent. Carbon and silicon contents were in the range of 3.00 to 3.40 per cent and 2.00 to 2.50 per cent respectively. The manganese contents in each case were such as given by the formula

$$\% \text{ Mn} = (1.7 \times \% \text{ S}) + 0.19 \text{ to } 0.22.$$

The castings were of light section ( $\frac{1}{8}$  to  $\frac{3}{8}$  in.) and were all of completely normal gray iron structure, with no chilled edges and normal machinability. All were examined for free iron-rich sulphides in the microstructure, and none were found. Hence it would seem that for higher carbon and silicon contents, freedom from iron-rich sulphides and accompanying in-

complete graphitization and poor machinability, may be obtained with a manganese content of as little as 0.20 per cent above theoretical MnS. It should be emphasized that it is definitely not recommended that manganese content of gray cast iron should be run this low intentionally. The data are given from the viewpoint of rate of graphitization only.

#### MR. CARTWRIGHT'S DISCUSSION

A. E. CARTWRIGHT (*Written Discussion*):<sup>4</sup> Mr. Rehder's contribution to exact information regarding this phase of the metallurgy of malleable iron is most valuable. His findings are the more impressive because of his meticulous attention to all details of experimental procedure.

It may be of interest and in some relation to this study for the writer to recall serious difficulties he encountered some 12 or 13 years ago in obtaining satisfactory second stage annealing in the production of short cycle malleable. It was about the time that interest had focused on the proposition that superheating of a base hard iron allowed of retaining carbon in the combined state in the presence of higher silicon content than un-superheated iron would permit,—the increased silicon markedly reducing the necessary annealing time. The indirect arc electric furnace was considered ideal for this purpose and a steel scrap charge carburized and with added silicon was used.

When it came to annealing the product, however, very erratic results were obtained until it was realized that, though the residual manganese content was generally quite low (0.35-0.45 per cent) yet the sulphur was so low in the scrap used that the manganese-sulphur ratio was frequently of the order of 10 or 12-1. Addition of sulphur in amount to bring this ratio to 3-1 eliminated annealing difficulties due to this cause. The main point in recalling this lies in the fact that the writer used iron sulphide for the addition of sulphur to the bath and suggests that use of this might result in a more reliable quantitative method of addition than would the very volatile elemental sulphur when translated into foundry-scale operation.

On one point particularly the writer would appreciate some further information or, at least, an opinion:

The analysis of the base iron used conforms with that of black-heart malleable most commonly produced and Mr. Rehder properly limits his conclusions to iron of that composition. Where straight cupola malleable is concerned, with a white iron carbon range of 2.80-3.10 per cent and silicon of 0.70-1.10 per cent, would the same Mn-S ratio be likely to give optimum second stage annealing results as for the lower carbon product?

#### MR. REHDER'S REPLY

J. E. REHDER (*Author's Reply to Mr. Cartwright*): The comments of Mr. Cartwright with respect to correction of manganese-sulphur ratio in malleable iron melted from carburized steel charges in the indirect arc furnace are appreciated and to the point. The author has had experience melting malleable iron commercially in this type of furnace, and found, as has Mr. Cartwright, that due to the nature of the furnace charge (5 to 10 per cent silvery pig, 40 per cent steel scrap, 50 to 55 per cent sprue, and graphite and ferro-silicon to suit), sulphur must be added because of the inherently high manganese-sulphur ratio of the charge. It is agreed that pyrite is the most suitable form of sulphur addition, but recoveries of elemental sulphur are surprisingly high and uniform, and since pure iron sulphide did not happen to be available at the time of the work described in the paper, possible contamination or inoculation of the experimental melts by use of impure mineral pyrite was avoided by the use of elemental sulphur. In commercial operation pyrite or iron sulphide was found satisfactory as a sulphur addition.

Another point encountered in the author's experience with commercial melting of malleable iron in the indirect arc furnace, mentioned above, may be of interest. Annealing in that plant was done in electrically heated ovens, and the annealing cycle (40 hr total) was set up to anneal the white iron with the high manganese-sulphur ratio found as melted. In the course of experimental work, two sets of white iron tensile and bend test bars were made up, poured from the same ladle, but with a sulphur addition being made to the iron for the second set of bars. The manganese content was about 0.38 per cent and the sulphur content was 0.04 per cent in one set of bars and 0.09 per cent in the other set. The bars were tied together with wire

<sup>4</sup> Crane Limited, Montreal, Quebec, Canada



and annealed, so that all received the same anneal. It was found on testing the annealed bars that the higher sulphur content bars were about 2,000 psi higher in yield strength, 4 per cent (16 to 20 per cent) higher in elongation, and over 45 degrees better in bend test (135 to 180 degrees) than the lower sulphur content bars. The microstructures of both sets of bars were almost identical, both being completely ferritic, the only difference being in the amount of manganese sulphide present and a relatively small difference in shape of graphite nodule.

Concerning Mr. Cartwright's question on cupola malleable iron, it was hoped originally to extend the work described in the paper to other ranges of carbon and silicon, but time limitations did not permit. However, based on experience and on some work on gray iron recently done, the author is presently of the opinion that the formula for manganese-sulphur relationship given in the paper should hold true for higher carbon content cupola melted iron.

W. R. BEAN:<sup>5</sup> Perhaps I can throw a bit of light on the origin of the use by the industry of the manganese-sulphur ratio. Two men besides myself in 1905 assembled a large number of the then used wedge tests from consecutive heats from a number of malleable foundries, all of which were given the same annealing treatment. These two were the late B. J. Walker of Erie Malleable Iron Co. and the late Dr. Enrique Touceda of Albany, N. Y., both of whom contributed immeasurably to the advancement of the art of malleable iron production. Our purpose in undertaking that study was to determine, if possible, where quality in malleable iron was determined, whether it came with the melting treatment or whether it was largely affected by the annealing treatment. Chemical analyses were made from all of this series of heats and, to the surprise of all of us, there was a very wide difference in manganese and sulphur contents.

Perhaps I should mention that many malleable foundries at that time considered charcoal iron to be an essential element in the production of quality malleable iron, and mixing was done principally by grading, by the fractures of the different pig irons which were used in the melts.

What we found was that in most of the foundries, all on the direct melting process, the usual sulphur content was of the order of 0.03 to 0.04 per cent and manganese was in the range of 0.17 to 0.24 approximately. The product of one foundry, however, showed consistently a sulphur content of between 0.10 and 0.13 per cent. That, to the three of us who were doing this work, was unheard of in an air furnace malleable iron, but with that composition, the quality was there, because the manganese happened to be in those heats approximately  $2\frac{1}{2}$  times the sulphur content. As a result of those findings, the three of us just arbitrarily set a manganese content of  $2\frac{1}{2}$  times the sulphur content, and that was the ratio which was used by a good many malleable foundries for a number of years, up to 1915 or 17.

The ratio of twice the manganese plus 0.15 was established as a result of a discussion which took place between Dr. Schwartz and myself, going back to the records to which I have referred as the basis. It was he who pointed out to me the fact that manganese sulphide contained manganese and sulphur in the ratio of 1.76 to 1. We then decided that a ratio of twice the sulphur plus 0.15 for the manganese was easier for the ordinary foundryman; that it was easier for him to multiply by 2 than it was to multiply by 1.76; that twice the sulphur plus 0.15 was perhaps better than  $2\frac{1}{2}$  times the sulphur content and that, I believe, the story of the beginning of the manganese-sulphur ratio.

MR. REHDER: I appreciate Mr. Bean's comments. It is interesting to learn that twice the sulphur plus 0.15, which is one of the formulas for manganese-sulphur balance commonly used, was moved up from 1.7 to 2 times the sulphur because of the simpler arithmetic involved. Actually there is little objection, since metallurgical control in many foundries is not close enough to make the small difference important. However, at higher sulphur contents such as prevailing in many duplexing plants today, this difference may become of interest, since at higher sulphur contents the manganese content should be controlled within narrower limits, as indicated in the paper.

It may be well to mention again the risk that a foundry may run if a definite ratio of manganese to sulphur is used. In a given plant, the sulphur content of the iron may have run close to, say, 0.08 per cent for years. In such a case, use of a definite ratio of, say,  $3\frac{1}{2}$  to 1 may indicate the correct manganese content; but this is also true if the formula  $(1.7 \times \% S) + 0.15$  is used. However, if sulphur content of the melt increases for some reason, and the same definite ratio of  $3\frac{1}{2}$  to 1 is used, more manganese will be indicated than is really necessary, and if used, annealability will suffer. The formula from the paper apparently gives accurate results over a range of sulphur contents.

MR. BEAN: I am not arguing that as correct, but just citing the case as a matter of interest. What you have done is a remarkable piece of work, in that it provides a scientific answer to questions that we did not have the answers to in the early years of this century.

W. R. JÄSCHKE:<sup>6</sup> How important is the accuracy of routine analysis on sulphur when applying these rules?

MR. REHDER: I am glad you brought that point up, since it is an important one. We do all our sulphur determinations by the combustion method, and consider it the only reliable method of obtaining accurate sulphur contents of white cast irons, especially of high sulphur white cast irons. We have found that the evolution method for sulphur will sometimes consistently give results as much as 0.06 to 0.10 per cent lower than the combustion method on the same sample, when the true sulphur contents are over about 0.15 per cent. Since many foundry laboratories use the evolution method, I think that the large amounts of manganese sometimes apparently necessary to neutralize the effect of sulphur on annealing are more an indication of inaccurate (low) sulphur analyses than of anything else. When sulphur content of the iron is running about 0.20 per cent, the question of accuracy of sulphur determinations may be more important than any considerations of suitable neutralizing manganese contents.

CHAIRMAN VENNERTHOLM: Figure 1 is of particular interest in that it demonstrates not only that the second stage annealing time can be reduced by using low sulphur irons, but also that the latitude of the manganese sulphur ratio is much broader as compared with the higher sulphur irons. For instance, for a 30-hr, second stage cycle, an 0.08 sulphur iron has a manganese to sulphur ratio of from 3 to approximately 4.5 as against 2.6 to 3 for an 0.21 sulphur iron. This is of considerable importance to the foundry in that it simplifies the metallurgical control when low sulphur is used.

MR. REHDER: In my experience, I have seen some excellent malleable irons containing 0.22 per cent sulphur, properly balanced with manganese and run through the regular anneal. The only disadvantage that I have seen to these high sulphur irons is the fact that manganese control has to be rapid and accurate, since as the sulphur content increases, the safe plus and minus tolerance on manganese content decreases. This effect is separate from neutralization point, as the correct amount of manganese for most rapid annealing is still given by the formula arrived at in the paper. However, if in plant operation there were any doubt about the accuracy of sulphur analyses in particular, I would be inclined to play a little safe on the manganese side until the analytical results were made more reliable.

R. N. SCHAPER:<sup>6</sup> Does balancing the sulphur with the manganese eliminate the hot tearing which is present in malleable irons where the sulphur is around 0.20 and the manganese is on the low side? In other words, no attempt has been made to balance the sulphur.

MR. REHDER: I have no data on the relationship between manganese-sulphur balance and tendency towards hot tearing. However, my opinion is that hot tearing will start to show as soon as you get free iron sulphide in the matrix, and you will get free iron sulphide as soon as you go much below the neutralization point given by the formula. It seems probable that severity of hot tearing may increase as the amount of manganese decreases below  $(1.7 \times \% S) + 0.15$ .

<sup>5</sup>Whiting Corp., Harvey, Ill.

<sup>6</sup>Marion Malleable Iron Works, Marion, Ind.

# EFFECT OF FOUNDRY PRACTICE ON PROPERTIES OF SOME BINARY COPPER-SILICON ALLOYS

By

A. I. Krynitsky,\* W. P. Saunders \* and H. Stern \*

## ABSTRACT

A study of the tensile properties of four different types of bars of copper-silicon alloys containing from 2.0 to 4.9 per cent silicon was carried out. The tensile strength increases with an increase in silicon content and a decrease in pouring temperature. An increase in the amount of precipitated constituents in the microstructure of these alloys construed to be the kappa and alpha plus kappa phases respectively, seems to be associated with an increase in tensile strength and a decrease in elongation. The optimum tensile properties were obtained from the Navy 10B type of tensile test bar.

In a study of the effects of different ambient atmospheres on the physical properties of the 4.9 per cent silicon alloy the castings made under hydrogen and carbon monoxide possessed lower tensile properties, density, and indentation hardness than those made under helium, air, carbon-dioxide, and nitrogen. The chief cause of unsoundness as determined by hydraulic pressure tests was attributed to the presence of interconnected shrinkage cavities caused by inadequate feeding.

## 1. Introduction

THE WARTIME SHORTAGE of tin increased the importance of the copper-silicon alloys since they were resorted to as replacements for tin bronzes. These alloys, usually containing one or more additional elements, are employed extensively among other uses for castings as housings, brackets, bushings, etc., requiring high tensile properties and good corrosion resistance.

Smith,<sup>1</sup> Andersen,<sup>2</sup> and others have investigated the constitutions and microstructures of the copper-rich alloys. The constitution diagrams presented by Smith (Fig. 1) and Andersen are in essential agreement, but neither these nor other available diagrams show the structural phases of the alloys at room temperatures. Gillett<sup>3</sup> in his general discussion of silicon alloys stated that the alpha alloys (Fig. 1) increase in strength and retain reasonably good toughness as silicon is added up to its limit of solubility. Substantiating evidence of this is revealed in the results of Voce<sup>4</sup> who showed that cast binary copper-silicon alloys containing 3 per cent and 4.5 to 4.8 per cent silicon had tensile properties which compared favorably with those of admiralty gun metal (10.2 per cent tin, 3.9 per cent zinc, balance copper) and phosphor bronze (12.55 per cent tin,

1.01 per cent zinc, 0.24 per cent phosphorus, balance copper). However, he reported an accompanying sharp decrease in tensile properties for an increase in silicon content to 6.5 per cent which is appreciably beyond the alpha solubility limit. Despite the availability of this and similar important data, on silicon bronzes, there is a dearth of information on the influence of factors such as pouring temperature and ambient atmospheres during melting and casting on the physical properties of castings. This prompted a study embodying these factors of cast copper-silicon alloys.

Although most of the commercial alloys of this type contain one or more elements in addition to copper and silicon, this work was limited to a study of the binary alloys in order to determine the influence on copper of silicon *per se*. The maximum silicon content of the alloys studied was maintained below 5 per cent since the indications are that the tensile properties are adversely affected if the alloys contain silicon beyond the limit of alpha solubility (Fig. 1).

The initial phase of the study was devoted to the

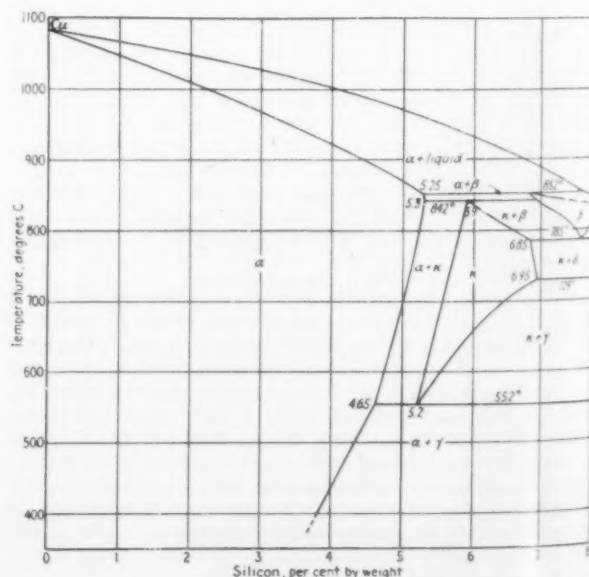


Fig. 1—Copper-silicon equilibrium diagram (Smith).

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determination of the effect of silicon content (2 to 4.9 per cent) on the tensile properties, electrical resistivity and pressure tightness of the binary alloys cast under ordinary atmospheric conditions. Since the method of casting tensile bars is known to have a profound influence on the results, four well known types of bars were used for this phase of the work to determine which gave the optimum results. The effect of ceramic hot tops (riser sleeves) on the pressure tightness of the different alloys also was studied. The second phase of the study was devoted to the determination of the effect of melting and casting under various gases on the physical properties of the 4.9 per cent silicon alloy. This alloy was chosen because, in general, it was found to have the highest tensile properties and pressure tightness when cast in air. This work was supplemented with a metallographic study of the alloys.

## II. Materials and Procedure

### A. Preparation of Stock Metal

The copper was melted under dried charcoal, heated to a temperature of about 2246 F (1230 C) and then the silicon was added. Under these conditions oxidation of the silicon was retarded and the silicon dissolved rapidly. The alloys were prepared in high frequency furnaces as melts ranging in weight from 200 to 430 lb. each. The metal was cast into pigs which were used as stock metal for subsequent remelting and casting into specimens. Commercially pure copper and silicon were used in preparing the alloys which contained 2.0, 2.5, 3.4, 3.5, 4.7 and 4.9 per cent silicon, respectively.

An analysis of the stock metal revealed that the 2.5 per cent and one of the 4.9 per cent silicon alloy melts contained 0.20 per cent iron. In all other cases the iron content did not exceed 0.06 per cent. The presence of 0.20 per cent iron resulted from iron skimmers inadvertently used instead of graphite skimmers in preparing these melts. The work of Andersen and Kingsbury<sup>5</sup> indicates that an iron content of 0.20 per cent does not significantly influence the alpha solubility limit of the silicon. Furthermore, no appreciable differences in tensile properties were obtained by the authors for specimens of the 4.9 per cent silicon alloy containing 0.20 per cent and 0.05 per cent iron, respectively, melted in air. Consequently it is assumed that 0.20 per cent iron had no effect on gas solubility of this alloy. The high (0.20 per cent) iron melt containing 4.9 per cent silicon was used exclusively for the studies of the specimens cast under different ambient atmospheres.

### B. Preparation of test specimens.

The specimens for the first part of the investigation were cast from melts of approximately 90 lb. each prepared from the stock metal ingots. The scrap metal remaining after casting a series of specimens was used in preparing subsequent melts of the same composition. The melts were made in a high frequency furnace using clay-graphite crucibles. No charcoal or flux was used. Although complete recovery of the silicon in preparing the stock metal was not obtained there

was practically no loss of this element during remelting. The compositions of these alloys, as determined by chemical analysis are shown in Table 1. Temperature measurements of all melts were made with a chromel-alumel thermocouple encased in a closed-end glazed porcelain tube which, in turn, was protected by a closed-end graphite tube. The maximum heating temperature throughout the work was 2246 F (1230 C). The temperature in pouring these melts ranged from 306 F (170C) to 36 F (20C) above the liquidus temperature of the alloys, indicated in Smith's diagram (Fig. 1). The molds for the test specimens were prepared from modified Albany heap sand, with properties as follows: water content, 6 to 7 per cent; permeability number, 14 to 20 (A.F.A. units); green compressive strength, 7 to 11 psi.

TABLE 1—TYPICAL ANALYSES OF COPPER-SILICON ALLOYS, MELTED AND POURED IN AIR

Cu, per cent	Si, per cent	Fe, per cent
97.9	2.0	0.03
97.2	2.5	0.20
96.4	3.4	0.03
96.0	3.9	0.02
95.3	4.7	0.03
95.0	4.9	0.06

The specimens cast for the first phase of the investigation were tensile test bars and hydraulic pressure cylinders. In the initial stages of this part of the work the tensile test bars, cast in groups of four molds, consisted of end-gate, fin-gate, keel and Navy 10B types (Fig. 2 to 5, inclusive). Each type was poured two in a mold, except for the fin-gate bar, Fig. 3, which was cast singly. However, the results of tests at this stage, confined to the 3.4 per cent silicon alloy, showed that the end-gate bar had coarser grain structure, considerably more segregation and cavitation, and accompanying lower tensile properties than the other types of tensile test bars studied. Therefore, the use of this type of bar for subsequent test work was discontinued. The variation in temperature during the pouring of any set of these molds did not exceed 27 F (15 C). These bars were machined to conform

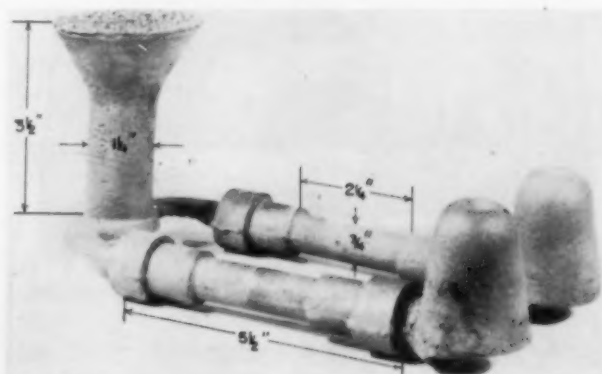


Fig. 2—End-gate tensile bar.



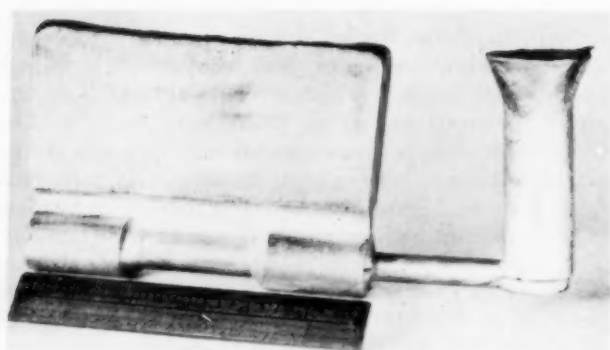


Fig. 3— $\frac{1}{4}$ -in. web fin-gate tensile bar.

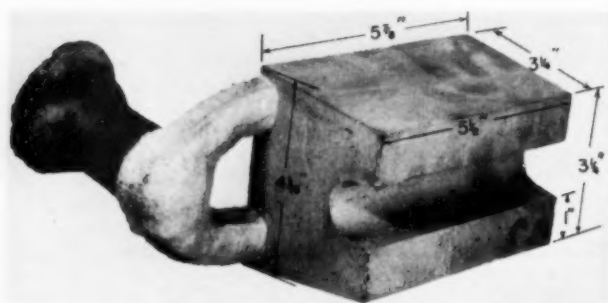


Fig. 4—Keel tensile bar.



Fig. 5—Navy non-ferrous tensile test bar type 10B.

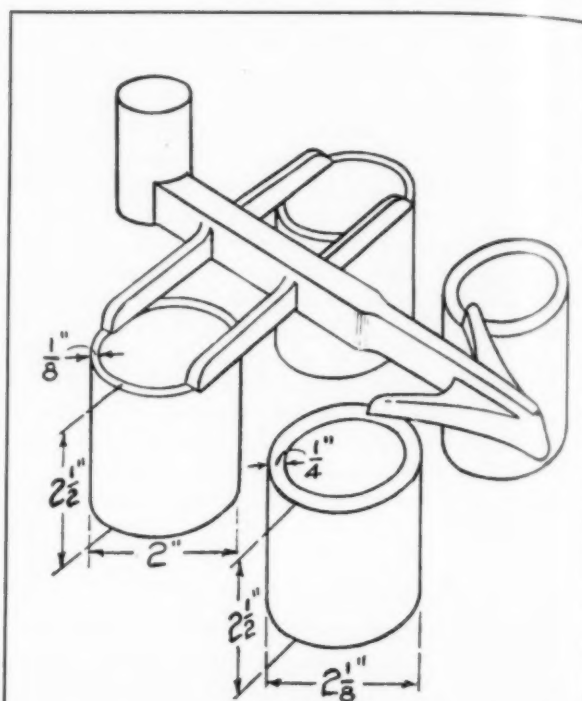


Fig. 6—Schematic representation of a casting of specimens for hydraulic tests.

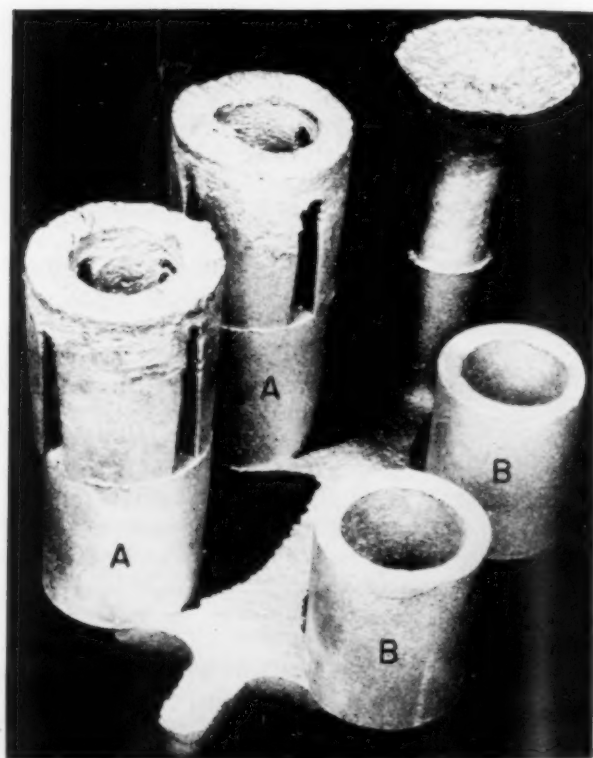


Fig. 7—Assembly of cylinders as cast, A with and B without insulated risers used in pressure test, each with  $\frac{1}{4}$ -in. wall thicknesses. All specimens before testing were machined to  $\frac{1}{8}$ -in. wall thickness.

to federal specification QQ-M-151a for 0.505-in. diameter, 2-in. gage length tensile test bars.

The hydraulic pressure test cylinders at first were cast four cylinders to a mold as indicated in Fig. 6. The two thin-walled castings in each group were trimmed to a length of 2 in., most of the material being removed from the gate-end. These specimens were tested hydraulically with the inner and outer surfaces in the as-cast condition. The other two cylinders in each group, likewise trimmed to a length of 2 in., were machined both inside and outside to a wall thickness of  $\frac{1}{8}$  in. before testing. It was found that all these cylinders invariably leaked at pressures of 300 psi or less. On the basis of microscopic examination, it appeared that shrinkage cavities caused by insufficient feeding were chiefly responsible for the unsoundness of the specimens. In an attempt to improve feeding during casting of these cylinders, a method described by Taylor and Wick<sup>6</sup> was used. Four pressure-test cylinders  $2\frac{1}{8}$  in. outside diameter by  $2\frac{1}{2}$  in. high with  $\frac{1}{4}$ -in. walls were cast in a mold as shown in Fig. 7. Two of the cylinders were supplied with risers which were insulated with gypsum sleeves (Fig. 8) during casting. All these cylinders were machined inside and outside to a wall thickness of  $\frac{1}{8}$  in. and to a length of 2 in. before testing.

In the second part of the investigation the 430-lb. batch of the 4.9 per cent silicon alloy was used. Separate 70-lb. portions of this virgin stock were used for each atmospheric melt and were melted in a clay-graphite crucible lined with aluminum oxide cement, and provided with a cover designed to permit the maintenance of a controlled atmosphere through the introduction of different gases (Fig. 9). A double-walled crucible was employed to prevent an excessive drop of temperature during pouring.

During the melting operation the gas was conducted into the covered crucible at the rate of 25 cu ft per hr. The rate was increased to 60 cu ft per hr when the lid was removed during additions to the melt and while pouring. The gas was directed over the surface of the melt and not bubbled through it. Advantage was taken of the characteristic stirring action on molten metal by the high frequency current to facilitate intimate contact between the melt and the ambient atmosphere. The ambient atmospheres employed were hydrogen, carbon monoxide, helium, air, carbon dioxide, and nitrogen.

#### Types of Test Bars Used

Four types of specimens were cast from each heat in the following order: pressure test cylinders (Fig. 7), a solid contraction bar (Fig. 10), Navy 10B tensile test bars (Fig. 5), and a vertically-cast stepped bar (Fig. 11). All castings were made in molds of green sand having the same characteristics as that previously described. The temperature at the beginning of a pouring operation was approximately 2156 F (1180 C) and about 2102 F (1150 C) during the final stages of pouring. The Navy 10B bars were cast at approximately 2138 F (1170 C) 360 F (200 C) above the liquidus, (Fig. 1) to accentuate any effects of the ambient atmosphere employed. Remelting the stock



Fig. 8—Insulating gypsum risers used in casting hydraulic pressure cylinders.

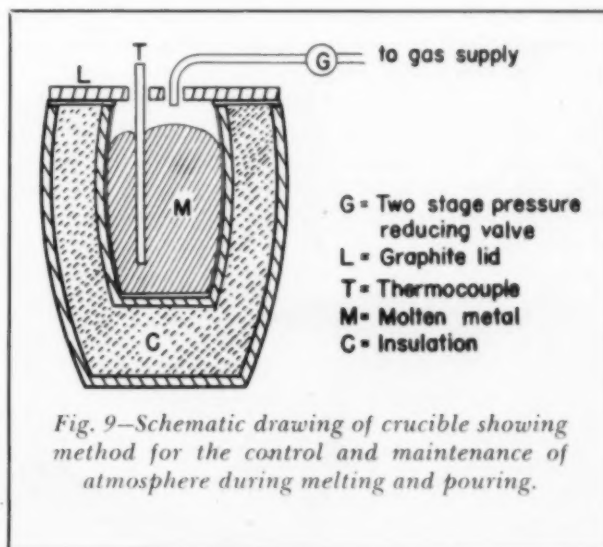


Fig. 9—Schematic drawing of crucible showing method for the control and maintenance of atmosphere during melting and pouring.

metal under various atmospheres did not significantly affect its chemical composition. All these specimens contained 94.8 to 95.0 per cent copper, 4.8 to 4.9 per cent silicon and 0.20 to 0.23 per cent iron. The hydraulic pressure test cylinders were cast four cylinders to the mold, two with insulated risers (Fig. 7), and machined as previously described. The tensile bars were machined to the same dimensions as those used in the first phase of the work. The linear contraction test results were obtained on the bar as cast (Fig. 10). A pattern of this shape has the advantage over patterns for ordinary castings in that rapping is not required in removing it from the sand. The stepped bars, used for density tests, were sectioned as shown in Fig. 11. The specimens cut from the stepped bar were surface ground to remove surface imperfections and adherent sand.

#### C. Test Procedure

The properties determined on the specimens tested in the first phase of the investigation included tensile properties, electrical resistivity, and hydraulic pressure tightness. These determinations were supplemented by indentation hardness, density and solid contraction tests in carrying out the second part of this study.

1. *Tensile properties*—All tensile tests were made in a 30,000-lb capacity Riehle hydraulic tensile testing machine, having a 15,000-lb range dial. The cross-head speed of the machine for all tests was approximately 0.1 in. per in. per min. The Templin type Baldwin-Southwark extensometer was used in recording the stress-strain diagrams. The yield strength was obtained from the stress-strain curve, using the offset method (0.2 per cent offset).

2. *Electrical resistivity*—Electrical resistivity measurements were made on the machined Navy 10B bars by a procedure conforming to the recommendations of the American Society for Testing Materials<sup>7</sup> before they were subjected to the tensile test.

3. *Indentation hardness*—Brinell hardness determinations with a 500-kg load applied to a 10-mm ball for 60 sec. were made on the threaded ends of the broken Navy 10B tensile bars, after they had been machined longitudinally to produce parallel flat faces, approximately  $\frac{5}{8}$ -in. wide.

4. *Solid contraction*—The difference between the dimension "D" of the pattern (12 in. in the present work) and the corresponding dimension of the casting (Fig. 10) was used as the criterion for determining the linear contraction of the alloy investigated.

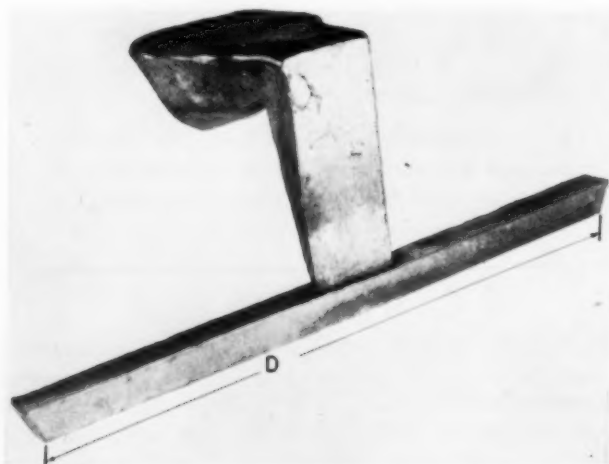


Fig. 10—Bar for determining total linear shrinkage.

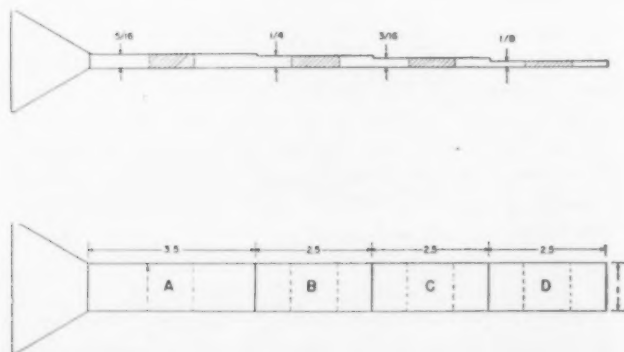


Fig. 11—Diagrammatic sketch of vertically cast stepped bar. Specimens indicated as A, B, C and D were used for determining the relationship of section thickness to density.

5. *Density*—Density determinations were made by the conventional loss of weight in water method. The results were based on tests on specimens taken from the stepped bars from positions marked A, B, C, and D (Fig. 11).

6. *Hydraulic pressure tightness*—The hydraulic pressure tests were made by applying pressure to blue-colored water which filled the cylinder, until a blue coloration was detected on the exterior of the cylinder. The pressure was noted at which the discoloration was first detected. If no evidence of leakage was found at a pressure of 300 psi maintained for one minute the pressure was then raised to 600 psi. If, after one minute at this latter pressure no leakage was observed, the casting was rated sound. All specimens thus rated were subsequently subjected to progressively increasing pressure until they leaked or the gasket failed. In no case did gaskets leak at pressures below 1700 psi.

### III. Results of Tests

#### A. Effect of composition, test bar design, and pouring temperature on alloys cast in air.

1. *Tensile properties*—The first tensile tests were conducted to ascertain the casting conditions and type of test bar which rendered the optimum tensile properties of the alloys cast in air. The results are given in Table 2.

The values for the end-gate bar are limited to the 3.4 silicon alloy, since this type of bar was discontinued in the early stages of the investigation as previously stated. All Navy 10B bars possessed a sound fracture free from any sponginess or segregations. Accordingly, the results in general show that the highest tensile values were secured with the Navy 10B bar. They further indicate in most cases that an increase in yield strengths and ultimate tensile strength accompanied a decrease in pouring temperature.

Comparison of the tensile properties of the different alloys determined with the Navy 10B bar, was limited to the results obtained on the specimens cast at approximately the same increment of temperature (216 F, 120 C) above the liquidus (Fig. 12). This figure shows a progressive increase in yield and ultimate strength and in general a decrease in reduction of area with increase in silicon content. However, the results for elongation were not so consistent. Elongation at first increased, reaching a maximum at 3.5 per cent silicon and then decreased sharply with further increases in silicon. Similar results for the ultimate tensile strength, elongation, and reduction in area were reported by Broniewski<sup>8</sup> for annealed wrought copper-silicon alloys. Broniewski's graphs show a progressive increase in elongation for silicon contents from 1 to 3 per cent and a decrease for further silicon increase.

2. *Electrical resistivity*—The relation between the electrical resistivity values and silicon content of the alloys is shown in Fig. 13. These results indicate that the resistivity increases in direct proportion to the increase in silicon content.

3. *Hydraulic pressure tightness*—In tests on specimens containing 2.0 to 3.5 per cent silicon and poured at temperatures ranging from 72 to 180 F (40-100C)



TABLE 2—EFFECT OF SILICON CONTENT, POURING TEMPERATURE AND DESIGN OF TEST BARS ON THE TENSILE PROPERTIES OF COPPER-SILICON ALLOYS

Silicon per cent	Pouring Temperature				End Gate Bar				Fin Gate Bar	
	°F	°C	Above Liquidus °F	°C	Yield Strength psi	Ultimate Strength, psi	Elonga- tion, per cent	Reduction in area, per cent	Yield Strength, psi	Ultimate Strength, psi
2.0	2120	1160	198	110	.....	.....	.....	.....	8000	32800
	2012	1100	90	50	.....	.....	.....	.....	.....	.....
2.5	2120	1160	217	120	.....	.....	.....	.....	8200	40700
	2012	1100	108	60	.....	.....	.....	.....	9000	37300
3.4	2084	1140	198	110	14400	44550	42	45	11000	47200
	1940	1060	54	30	12830	45850	31	37	12450	52000
3.9	2156	1180	306	170	.....	.....	.....	.....	12200	45200
	2012	1100	162	90	.....	.....	.....	.....	12300	42300
4.7	2156	1180	360	200	.....	.....	.....	.....	17300	32000
	1832	1000	217	120	.....	.....	.....	.....	16400	36700

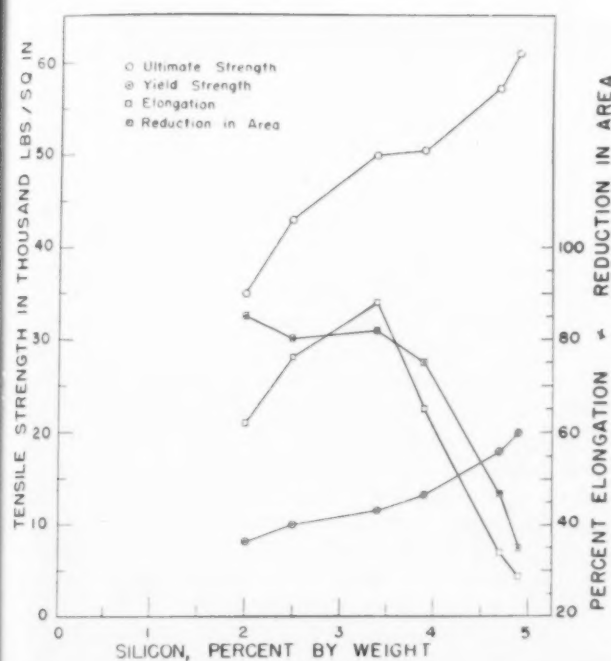


Fig. 12—Influence of silicon content on tensile properties of copper-silicon alloys, in the form of Navy 10B test bars, poured at a temperature approximately 216 F above the liquidus.

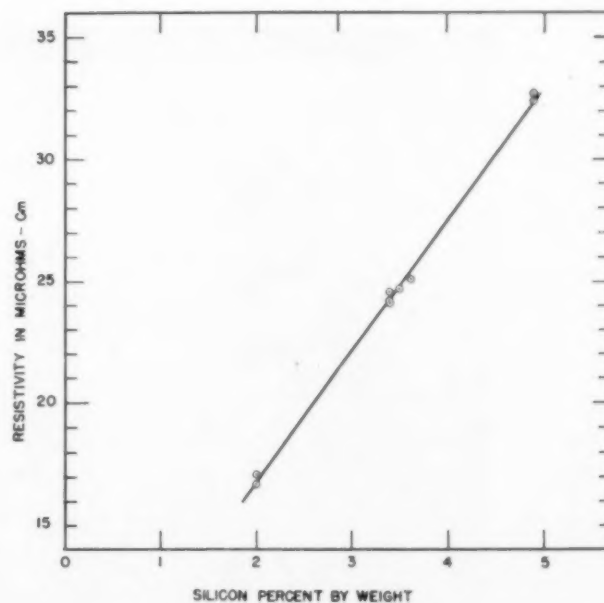


Fig. 13—Influence of silicon content on electrical resistivity values of copper-silicon alloys.

a riser contains many cavities, while its companion is practically free of porosity.

#### B. Effect of ambient atmosphere on the properties of the 4.9 per cent silicon alloy

The 4.9 per cent silicon alloy was used exclusively for determining the properties of silicon bronze melted and poured under various ambient atmospheres since it was found to have the highest tensile strength of the alloys investigated.

1. *Tensile properties*—Figure 15 shows the tensile test results obtained from Navy 10B bars cast under cover of each of the gases noted in the figure. Each point on the graph represents the average of four determinations. The castings made under atmospheres of hydrogen and carbon monoxide had appreciably lower ultimate strength, elongation, and reduction in area than those made under the other gases. The maximum tensile properties were obtained on the

above the liquidus, no specimen cast with an insulated riser showed leakage at pressures less than 600 psi while practically all the companion cylinders without the risers leaked at pressures less than 300 psi. However, it is significant that some cylinders of an alloy containing 4.2 per cent silicon cast without risers at a temperature of 180 F (100 C) above the liquidus withstood hydraulic pressures of 1000 psi or more. This result suggests that the silicon content may influence the pressure tightness of the casting. Figure 14 shows areas of maximum porosity in unetched sections of two cylinders of an alloy containing 3.5 per cent silicon, one cast with an insulated riser and the other cast without a riser and both poured at 1994 F (1090 C) approximately 126 F (70 C) above the liquidus. It is evident that the cylinder cast without

TABLE 2—EFFECT OF SILICON CONTENT, POURING TEMPERATURE AND DESIGN OF TEST BARS ON THE TENSILE PROPERTIES OF COPPER-SILICON ALLOYS

Elonga- tion, per cent	Reduction in area, per cent	Keel Bar				Navy 10B Bar			
		Yield Strength, psi	Ultimate Strength, psi	Elonga- tion, per cent	Reduction in area, per cent	Yield Strength, psi	Ultimate Strength, psi	Elonga- tion, per cent	Reduction in area, per cent
32	46	10050	35200	50	84	8400	34700	57	86
..	..	8500	36350	77	86	13150	37850	50	84
69	65	8900	40000	71	80	9950	43000	76	80
34	46	9850	43400	80	81	11000	44750	71	67
59	52	9540	47400	76	69	11700	50000	88	82
64	52	10860	51800	87	79	14900	53100	75	68
49	48	10650	39700	43	51	12300	48150	64	48
32	30	12100	49850	63	57	12700	51900	88	60
5.5	19	14800	34800	14	22.5	16100	40150	16.8	31
13.5	23.5	15100	36350	16.5	37	17900	57250	33.8	47

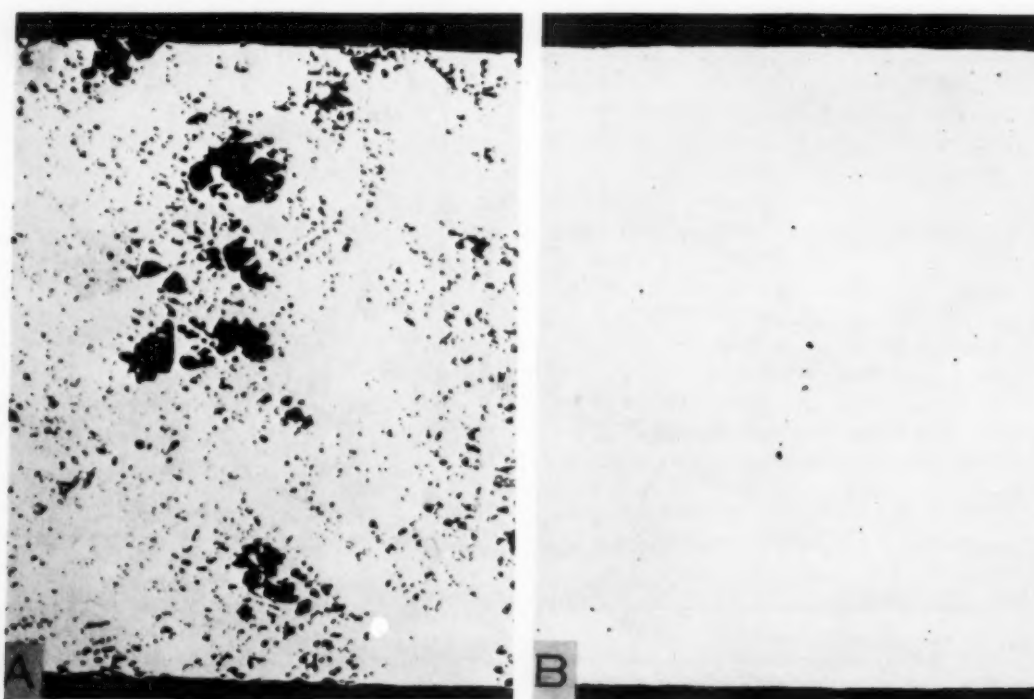


Fig. 14—Photomicrographs showing typical maximum observed porosity in cross sections of pressure test cylinders of the 3.5 per cent silicon alloy, melted and

poured in air. Pouring temperature was 126 F above the liquidus. Unetched, Mag. 28X. A, poured without insulated riser. B, poured with insulated riser.

castings made under nitrogen. The yield strength was practically unaffected by atmospheric change.

2. *Indentation hardness*—The results shown in Fig. 15 indicate that the effect of the enveloping atmospheres on the Brinell hardness number of the cast alloy was not nearly so pronounced as on the corresponding ultimate tensile strength, elongation, and reduction in area. The lowest hardness values were obtained on the specimens cast under hydrogen and carbon monoxide. Practically no difference in hardness was observed for the alloy cast under the other gases.

3. *Electrical resistivity*—The differences in resistivities among the specimens representative of various atmospheric melts were small, the maximum being

33.4 microhms-cm (for hydrogen and carbon monoxide) and the minimum 32.5 microhms-cm (for air and helium). The average resistivity value of the alloy cast under all of the gases was 33 microhms-cm.

4. *Solid contraction*—The solid contraction of the alloy was not significantly affected by the different ambient atmospheres. The contraction (pattern shrinkage allowance) was about 11/64 in. per ft. (approximately 1.4 per cent) in each case.

5. *Density*—Density determinations made on specimens cut from each of the four sections of the stepped bar (Fig. 8) cast under various atmospheres are given in Fig. 16. The density values of specimens cast under hydrogen or carbon monoxide were lower than those of specimens cast under any of the other gases. In

addition, the effect of these two gases on density was influenced considerably by the section thickness of the casting. With gases other than carbon monoxide and hydrogen, no appreciable difference in density was observed for different sections of the stepped bar. The maximum density observed was about 8.37 g/cm<sup>3</sup> obtained from a 5/16-in. section of bar cast under nitrogen, and the minimum was about 8.28 g/cm<sup>3</sup> obtained from a 3/16-in. section of bar cast under hydrogen.

The data plotted in Fig. 16 indicate that the densities of the 1/8-in. sections of the specimens cast under carbon monoxide or hydrogen were considerably greater than those for the larger sections of these bars. A similar effect, although much less pronounced was observed for hydrogen in the Bureau's study<sup>9</sup> of the characteristics of leaded red brass (Cu 85, Sn 5, Pb 5, Zn 5).

### Effect of Hydrogen

Observations made by other investigators on the effect of hydrogen on some copper alloys may be of interest. Bailey<sup>10</sup> found that hydrogen is readily soluble in many molten metals, and Prytherch<sup>11</sup> showed that while hydrogen is soluble in molten copper, its solubility is markedly reduced at the freezing point of the metal. A decided difference of opinion has been expressed concerning the practical significance of the decreased solubility of hydrogen in copper at the freezing temperature. Prytherch believes that this change in solubility results in unsoundness when copper is chill-cast. Allen<sup>12</sup> stated that, whereas commercial copper chill-cast is dense, the reverse is true for hydrogen-treated copper. However, Bailey<sup>10</sup> demonstrated that bronze (with 5 per cent Sn), treated with hydrogen and cast in chill molds contained 2.5 per cent of voids while the same bronze cast into sand molds showed about 9.2 per cent of voids. Bailey attributed this to the retention of more gas in solid solution in the case of the chill-cast bars.

The cooling rate of the various sections of the stepped bar casting should increase with a decrease in thickness of section. Assuming that the cooling rate of the 1/8-in. section approximates the rate of cooling in the chill mold, then the higher density observed for the 1/8-in. section, explained on the basis of Bailey's conclusion would indicate that more hydrogen had been retained in solid solution by the solidified metal in the 1/8-in. section than in the thicker sections.

### Effect of Carbon Monoxide

A noticeable increase in the density of the 1/8-in. section as compared with other sections of the stepped bar likewise has been demonstrated (Fig. 16) for carbon monoxide. One might presume therefore, that carbon monoxide also is soluble in molten copper-silicon alloys and partially retained in solid solution during rapid cooling, but there are no data to support such a supposition. A slight solubility of carbon monoxide in molten bronze was reported by Pearson and Baker<sup>13</sup> while practically no solubility was found by Floe and Bever<sup>14</sup>. Eastwood and Kura<sup>15</sup>, in their extensive discussion of gases in copper-base alloys,

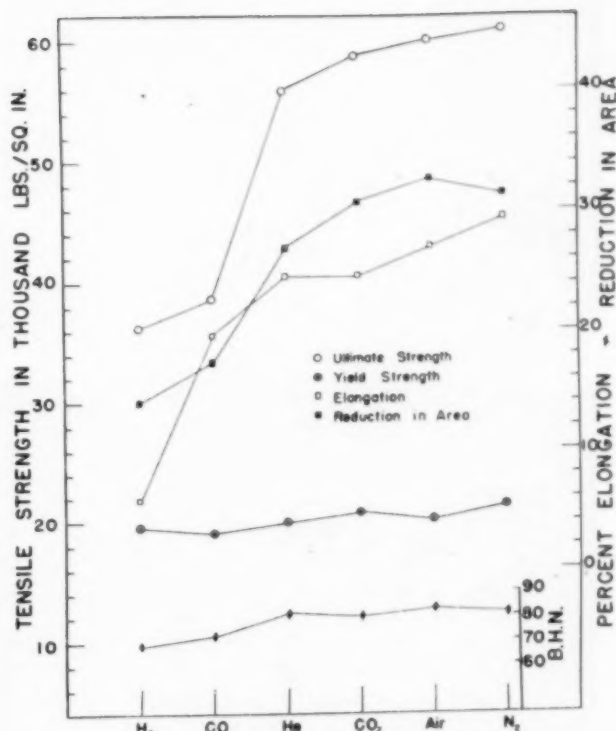


Fig. 15—Influence of different ambient atmospheres during melting and pouring on tensile properties and Brinell hardness determined on Navy 10B test bars of the 4.9 per cent silicon alloy.

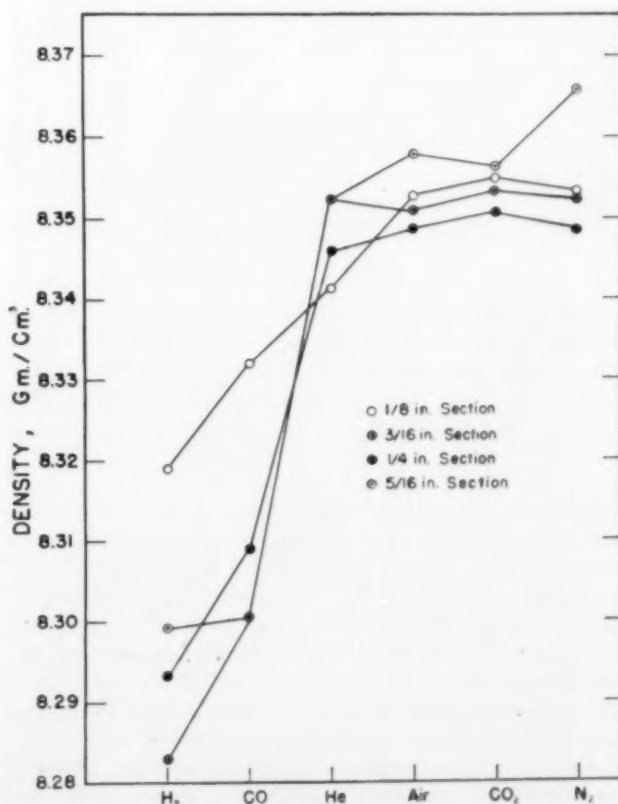


Fig. 16—Influence of various ambient atmospheres during melting and pouring on density of the 4.9 per cent silicon alloy determined on four sections of stepped bars (Fig. 11).



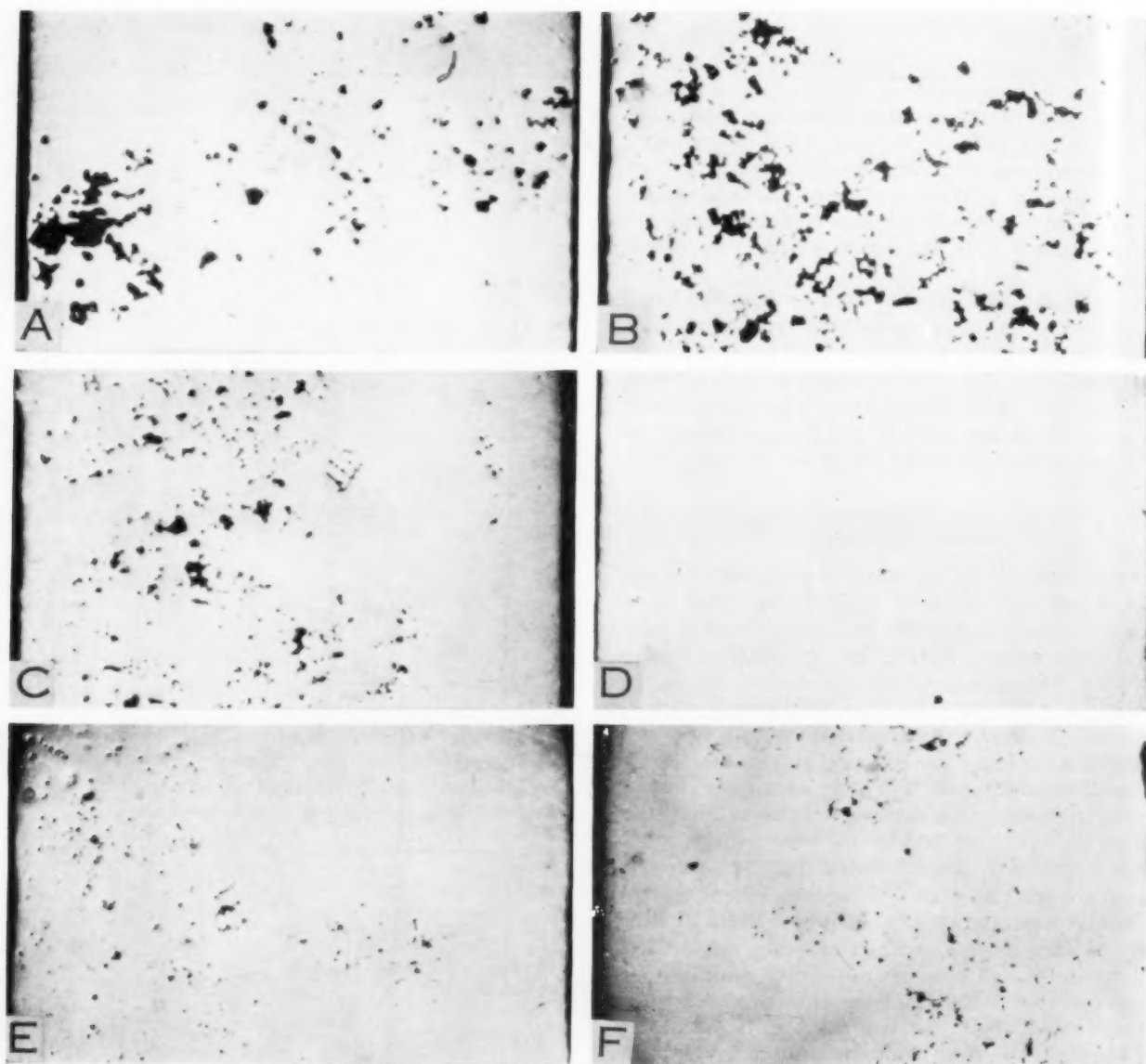


Fig. 17—Unetched photomicrographs at 28X of cross sections of pressure test cylinders of the 4.9 per cent silicon alloy cast without risers under atmosphere of: A, hydrogen; B, carbon monoxide; C, helium; D, carbon dioxide; E, air; F, nitrogen.

came to the conclusion that there is no unanimity of opinion regarding the solubility of either carbon monoxide or carbon dioxide in the molten metal. It is possible that the carbon monoxide only indirectly affects the physical properties of cast copper-silicon alloys as explained by Pearson and Baker<sup>13</sup> as well as Eastwood and Kura<sup>15</sup>, who doubt that carbon monoxide of itself causes porosity. They believe that the presence of a large proportion of carbon monoxide in the furnace atmosphere results in a lower oxygen content of the melt, and higher gas absorption of the reducing materials, such as hydrogen. It is evident that the problem of gas solubility is far from being solved and needs further study.

6. *Hydraulic pressure tightness*—The hydraulic pres-

sure tightness tests made on cylinders of the 4.9 per cent alloy cast with and without risers under different controlled atmospheres, showed except in one case that no leakage occurred at pressures up to 2000 psi. The exception was one cylinder cast without a riser under hydrogen, which leaked at 100 psi, but the duplicate cylinder did not leak under hydraulic pressures up to 2000 psi.

In view of the almost universal soundness of the cylinders containing 4.9% silicon cast without risers under various atmospheres, it was difficult to explain the adverse hydraulic test results from cylinders containing 2.0 to 3.5 per cent silicon similarly cast in air. Proper feeding during solidification is usually a prime factor in promoting soundness but in these two series of tests, the mold design was identical. However, in the tests of the 4.9 per cent silicon alloy carried out under different gases, a higher pouring temperature (approximately 360 F [200 C] above the liquidus) was used. The question arose, therefore, as to whether the difference in feeding was due to the change in the silicon content or to the variation in pouring temper-

Fig. 18  
air sh  
and b

ature and further experiments were undertaken to furnish more information on this point.

Alloys containing 2.4, 3.4 and 4.2 per cent silicon were melted in air and poured into molds without risers at temperatures of 360 F (200 C), 270 F (150 C), and 180 F (100 C) above the liquidus. The results of tests on the machined cylinders, given in Table 3, show that none of the low silicon alloys (2.4 per cent Si) passed any hydraulic pressure test, while the cylinders made of alloys containing 3.4 and 4.2 per cent silicon and poured at a temperature of 360 F (200C) above the liquidus withstood pressures up to 1,000 to 1,500 psi. Some of the 4.2 per cent silicon specimens poured at low temperatures also withstood hydraulic pressures of 1,000 psi or more (Table 3). These results indicate that both the silicon content and the pouring temperature influence the pressure tightness.

Figure 17 shows the area of maximum porosity in cross sections of the pressure test cylinders of the 4.9 per cent silicon alloy cast under different gases at 360 F (200 C) above the liquidus. These micrographs indicate that all cylinders except those cast under carbon dioxide were porous. They further demonstrate that those cylinders cast under hydrogen and carbon monoxide (Fig. 17A and B) had the highest degree of porosity. It is noteworthy that none of these cylinders leaked at pressures as high as 2,000 psi. Apparently the cavities observed did not form a continuous path along which water could pass.

These results suggest that the chief cause of leakage was not gas porosity alone but interconnected shrinkage cavities extending throughout the section of the cylinder. When adequate feeding was provided during solidification, the castings were sound as far as the hydraulic pressure test is concerned.

A recent investigation, carried out by Riddell<sup>16</sup> indicated that the pressure tightness of leaky bushings

TABLE 3—RESULTS OF HYDRAULIC PRESSURE TESTS OF MACHINED-TO-SIZE CYLINDERS CAST OF DIFFERENT COPPER-SILICON ALLOYS WITHOUT RISERS, MELTED IN AIR, AND POURED AT DIFFERENT TEMPERATURES

Alloy	Pouring Temperatures Above Liquidus		Behavior Under Pressure*		
	°C	°F	300 psi	600 psi	Failure at Pressure psi
Cu + 2.5% Si + 0.04% Fe	200	360	ns		
	200	360	ns		
	150	270	ns		
	150	270	ns		
	100	180	ns		
	100	180	ns		
Cu + 3.4% Si + 0.07% Fe	200	360	s	s	1000
	200	360	s	s	1500
	150	270	s	ns	
	150	270	s	s	1000
	100	180	ns		
	100	180	ns		
Cu + 4.2% Si + 0.08% Fe	200	360	s	s	1500
	200	360	s	s	1400
	150	270	ns		
	150	270	s	s	1300
	100	180	s	s	1500
	100	180	ns		

\* s—Sound (no leakage under pressure).

ns—Not sound (leaked under pressure).

of gun metal, valve bronze, and hydraulic bronze, was improved when they were annealed for 3 hr. at 1202 F (648 C) to 1299 F (704 C) in an air or oxygen-rich air atmosphere. The explanation advanced for this improvement is that the pores and fine shrinkage cracks in the specimens are sealed by the formation of oxides of a larger specific volume than the base metal. An effort by the authors to cure or reduce the leakage

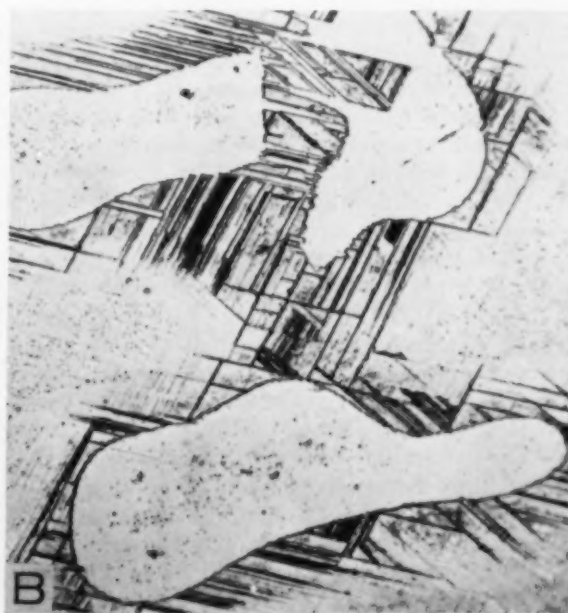
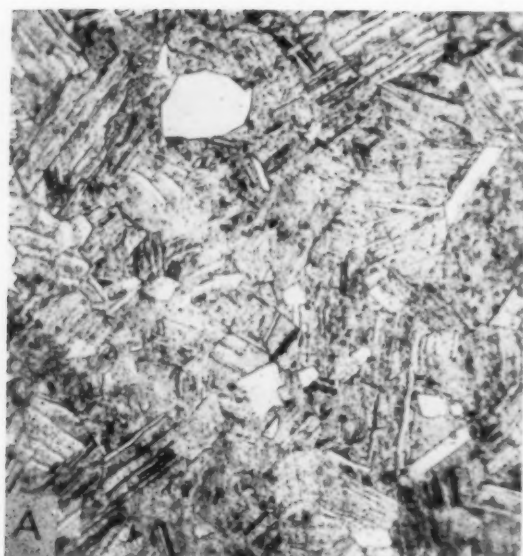
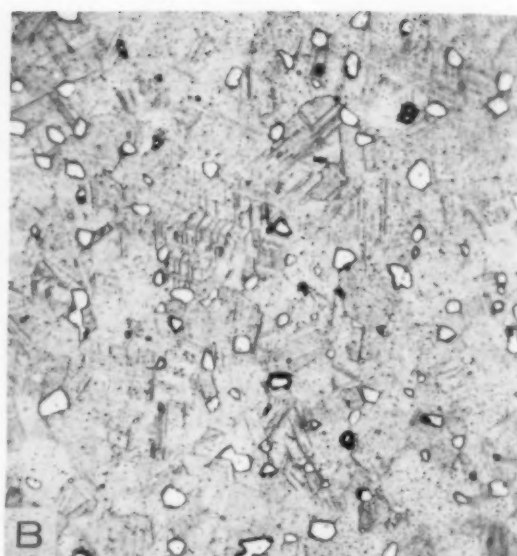


Fig. 18—Alloy containing 4.7 per cent silicon cast in air showing alpha dendrites, and precipitated massive and banded constituents interpreted as the kappa and

alpha plus kappa phases respectively. Etched in alkaline peroxide (modification of Smith's reagent). Mag. —A, 100X; B, 500X.



A, Anneal 3 hr at 1472 F, hot forged reannealed at 1472 F for 21 hr, quenched, cold forged, annealed for 192 hr at 1058 F, and quenched. Mag.—500X. Massive constituent, white under microscope in alpha ground mass interpreted as kappa phase.



B, Annealed 3 hr at 1472 F, hot forged, reannealed at 1472 F for 21 hr, quenched, cold forged, annealed 141 hr at 986 F, and quenched. Mag.—500X. Massive constituent light blue under microscope in alpha ground mass interpreted as gamma phase.

Fig. 19—Microstructure of alloy containing 4.7 per cent silicon cast in air. Etched in alkaline peroxide (modification of Smith's reagent).

of cylinders by annealing them at 1202 F (650 C) in air for 2 hr. did not prove successful.

#### IV. Microstructure

The solubility of silicon in the alpha phase according to Smith's diagram (Fig. 1) reaches a maximum at 5.3 per cent at 1548 F (842 C), decreases to 4.65 per cent at 1026 F (552 C) and is less at lower temperatures.

The structure of the alloys in the cast condition would be expected to differ from that indicated by the constitutional diagram because the cooling rate of the casting is seldom slow enough to allow equilibrium to be reached in the solid state. In the process of solidification of a copper-silicon alloy containing, for example, 4.7 per cent of silicon, the solid which first separates when the alloy begins to freeze will contain approximately 2.7 per cent silicon. If the cooling at this stage is not slow enough to allow the attainment of complete equilibrium, the central portion of each crystal will remain considerably richer in copper than the later-formed portions.

On the basis of Smith's diagram it may be assumed that the last formed crystals are alpha-plus-kappa and kappa phases. Smith states, "In castings that have not been worked, the kappa phase is largely segregated in the last part of the dendrite to solidify and is clearly outlined." During subsequent cooling under equilibrium conditions the alpha-plus-kappa and kappa phases should transform to the alpha plus gamma phase at 1026 F (552 C). However, as stated by Smith, and indicated by the work of the present

authors, the decomposition of alpha-plus-kappa to alpha-plus-gamma below 1026 F (552 C) is extremely slow unless the metal is cold worked. Thus, the structure revealed upon examination of the cast 4.7 per cent silicon alloy (Fig. 18) was interpreted in light of Smith's findings to consist of dendrites of the alpha phase and interstices filled with the banded alpha-plus-kappa and massive kappa phases. In order to further verify this conclusion two specimens taken from the threaded end of the same tensile test bar of the 4.7 per cent silicon alloy were heated at 1472 F (800 C) for 3 hr, hot forged (60 per cent reduction), given a homogenizing treatment of 21 hr at 1472 F (800 C), and water quenched. Following this treatment one specimen was cold forged (50 per cent reduction) heated to 1058 F (570 C), held at that temperature for 192 hr, and quenched in water. The second specimen was cold forged (50 per cent reduction) held at 986 F (530 C) for 141 hr and quenched in water. Figure 19 shows the structures of these two specimens etched with a modification of Smith's reagent consisting of 20 ml 3 per cent  $H_2O_2$ , 50 ml conc.  $NH_4OH$ , 5 ml 20 per cent  $NaOH$ .

In conformance with Smith's constitution diagram (Fig. 1) the white massive constituent shown in Fig. 19A was interpreted by the authors as the kappa phase. The massive constituent shown in Fig. 19B appeared light blue under the microscope and in this respect differed from that referred to in Fig. 19A. Since the specimen having the light blue particles was quenched after cold working from the alpha plus gamma range 986 F (530 C), this constituent was construed as the gamma phase. On this basis it was con-



cluded that the gamma phase was not present in the 4.7 per cent silicon alloy in the as cast condition (Fig. 18).

Microscopic examination of the 3.5 per cent silicon alloys cast in air revealed that it also contained the massive and banded constituents but in smaller proportions than observed in the 4.7 per cent silicon alloy. In the 2.0 per cent silicon alloy these phases were not detected.

It has been demonstrated in this paper that for a silicon content ranging from 2.0 to 4.9 per cent, the tensile strength of the copper-silicon alloys increased with an increase of silicon content. The elongation increased with an increase of silicon up to 3.5 per cent and with a further increase in silicon there was a sharp decrease in this property. In correlating this observation with the structures described above, it appears that the increase in strength and the decrease in elongation are associated with the increasing proportion of the massive and banded constituents.

### V. Summary

1. In this investigation a study of binary copper-silicon alloys was undertaken. In the first part of the work the influence of silicon content and pouring temperature on the physical properties was determined for a series of alloys, containing 2 to 4.9 per cent silicon, melted and cast in air. The effect of design of tensile bar on the tensile properties also was ascertained. The authors are aware, of course, that the same melt quality may not be optimum for castings of various designs and purposes. In the second part, the physical properties of a selected alloy (copper plus 4.9 per cent silicon) cast under various ambient atmospheres were studied. The atmospheres used were: hydrogen, carbon monoxide, helium, air, carbon dioxide, and nitrogen.

2. Tensile tests made using the Navy 10B, keel, fin-gate and end-gate types of bars showed that in general, the highest tensile properties were secured with the Navy 10B bar.

3. In most cases an increase in yield strength and ultimate tensile strength accompanied a decrease in pouring temperature.

4. An increase in the silicon content increased the yield and ultimate strength values and the electrical resistivity, but decreased the reduction in area. However, elongation increased with increased silicon content to 3.5 per cent but decreased sharply with further silicon additions.

5. The castings made of the 4.9 per cent silicon alloy under atmospheres of hydrogen or carbon monoxide, poured at approximately 2138 F (1170 C), possessed lower ultimate strength, elongation, and reduction in area than those made under other gases. Castings made under nitrogen showed the highest ultimate strength and elongation. The yield strength was practically unaffected by a change in atmosphere.

6. Melting and pouring under different gases had little effect on the electrical resistivity and solid contraction of the 4.9 per cent silicon alloy.

7. No appreciable difference in density was observed for the different sections of the stepped-bars cast under helium, air, carbon dioxide, and nitrogen,

respectively. The densities of the bars prepared under hydrogen and carbon monoxide were considerably lower than those cast under other ambient atmospheres. In addition, the density values of the 1/8-in. sections in these two cases were considerably greater than those of the heavier sections.

8. The results obtained with the hydraulic pressure test indicated that the chief cause of leakage of the cylinders was the presence of interconnected shrinkage cavities due to inadequate feeding.

9. The results obtained indicate that an increase in the proportion of the massive and banded constituents interpreted as the kappa and alpha plus kappa phases respectively is associated with an increase in strength and a sharp decrease in the elongation.

### Acknowledgment

Grateful acknowledgment is made to Dr. Francis A. Smith and Rosemond K. Bell of the Chemistry Division for their cooperation and to Richard H. Harwell and Richard C. Rohan for their assistance in the experimental work.

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## DISCUSSION

Chairman: R. M. BRICK, University of Pennsylvania, Philadelphia

Co-Chairman: C. A. ROBECK, Gibson & Kirk Co., Baltimore

C. S. SMITH<sup>1</sup> (Written Discussion): The extract from my 1940 paper<sup>1</sup> regarding the structure of castings might be amplified. The last part to solidify is, of course, beta, not kappa. This occurs at a temperature of 852 C but is almost immediately followed—at 842 C—by a peritectoid reaction between beta and alpha to form kappa. In effect, this results in the replacement of beta by approximately twice its volume of kappa without significant change in the composition of the contiguous alpha phase. An extremely rapid quench from this temperature would show only alpha plus kappa. Relatively slow cooling, however, enables kappa to precipitate from the alpha in those regions containing over about 4.6 per cent silicon, producing the oriented banded structure so characteristic of these alloys.

It might be added that the machinability of the alloys is greatly improved by the presence of small amounts of kappa, either massive or in the banded form.

J. E. CROWN<sup>2</sup> (Written Discussion): The data presented is of fundamental interest to the commercial producers of copper-silicon alloy sand castings. The findings reported by the authors furnish a logical basis for determining causes of and remedies for defective copper-silicon alloy sand castings which are chronically produced in commercial foundries where operating conditions are not as easily controlled as in a laboratory. Additional data pertaining to commercial copper-silicon alloy compositions such as the 92.5 per cent Copper, 3.0 per cent Silicon, 4.0 per cent Zinc and 0.75 per cent Iron used at the Naval Gun Factory would be of practical value.

The Fig. 10B test coupon is used at the Naval Gun Factory for testing production heats of bronze alloys, such as copper-tin, copper-tin-zinc and copper-silicon-zinc alloy. The tension test results obtained at the Naval Gun Factory have been consistent in bars from the same heat and a good measure of the quality of the molten metal from heat to heat. Naval Gun Factory experience, therefore, supports the author's selection of the Fig. 10B test coupon for testing foundry practice variations.

R. A. COLTON<sup>3</sup>: I would like first to compliment Col. Krynitsky for undertaking work of this type. The commercial copper-silicon alloys have been much maligned, undeservedly, I think, since they can be a very useful group of alloys.

I would like to point out some of our experiences with them that may be of interest. In regard to test bar molds for silicon bronzes, we have tried most of the conventional types and have found that the test bar mold giving the best properties is the so-called "two-bar" aluminum mold. I believe this type bar meets ASTM specifications, and since the tensile strengths run to 50,000 or 60,000 psi and the elongations up to 80 per cent, it is an advantageous one to use.

With the commercial alloys, such as Herculoy, the melting practice is rather simple. The furnace should be charged with clean foundry scrap, or with ingot, and the metal melted with an oxidizing flame, using no flux or cover. Care should be taken not to overheat the metal, and an effort made to pour castings below 2250 F. If the melting atmosphere is correct, and the metal temperature kept low, there should be less trouble with these alloys than usually found with tin bronzes.

I know of one case where Herculoy is being used in a hydraulic valve for a Watson-Stillman press. This is a rather heavy casting, of a 1-in. wall section, and has to withstand 6,000 lb internal pressure. When silicon bronze castings do leak, and occasionally they do, it is usually from inadequate feeding, causing interdendritic porosity, and rarely from gas evolution. I believe that Herculoy is a useful metal for high pressure valves since castings can be made pressure tight.

The complaint is occasionally made that the silicon bronzes do not machine well. I would like to point out, with regard to the commercial silicon bronzes, that most of them are two-phase alloys, as the micrographs showed, and as with many two-phase alloys, machinability is fair. They do not machine as well as 85-5-5-5, but they can be machined fairly well and are

good casting alloys and can be well recommended for many engineering applications because of superior mechanical properties and corrosion resistance.

I know of one shop up in New York State that uses the silicon bronzes instead of 85-5-5-5 for running off castings when there is no specific alloy required. They like the silicon bronzes better because they come out of the sand clean, make a better looking casting and have far superior properties.

E. M. SMITH<sup>4</sup>: The authors are to be commended very highly for the excellent presentation of the considerable amount of data presented in their paper.

Additional discussion of density values is desirable, however, and reference is made to Fig. 16. Restricting discussion to melts prepared under hydrogen, it is shown on the graph that the density of the  $\frac{3}{16}$ -in. section is least, contrary to expectations.

The viewpoint has been presented by the authors that on the basis of Bailey's work there were definite indications that chill casting tended to increase the density, and slow cooling tended to make the castings more porous. It is requested that the authors expand on the significance of the apparent discrepancy in the density of the  $\frac{3}{16}$ -in. section. I would expect that the density should be greater if chilling alone will affect the density.

With reference to the data presented in Fig. 16 and the photomicrographs in Fig. 17, density values and amount of porosity appear to be incompatible. Roughly computing the porosity on the basis of density there is a difference of about eight points for eight hundred between the density of melts prepared under hydrogen and carbon monoxide a maximum porosity of about one per cent should be expected. The photomicrographs indicate that melting under carbon dioxide results in the least porous or most dense casting, that is "D" in Fig. 17. The data that are plotted in the preceding Fig. 16 indicated, on the other hand, that air, CO<sub>2</sub>, and nitrogen are practically equivalent with respect to the effect of atmosphere in density.

Now, do the authors have density values determined on those specimens shown in the photomicrographs?

Although the least density plotted in Fig. 16 is 8.28 gr per cc and suggests that one per cent porosity is the maximum obtained, the photomicrographs certainly indicate that porosity substantially in excess of one per cent may be obtained.

MR. KRYNITSKY: No density determinations were made on the pressure-test cylinders, only on the stepped bars. With regards to Mr. Smith's remark concerning a discrepancy observed for the density values of the  $\frac{3}{16}$ -in. section I will ask Mr. Saunders, the co-author of the paper to answer this question.

MR. SAUNDERS: The density determinations made on the stepped bar sections under the hydrogen were run several times. When we first saw the apparent discrepancy we tried to rationalize it on the basis of rate of chill, and it is obvious that we cannot. As far as the experimental error is concerned, each time that the experiments were repeated, as far as the density determinations were concerned, the same type of thing was observed, and I have no concrete evidence or even a suggestion as to why that should occur.

CHAIRMAN BRICK: Was the  $\frac{3}{16}$ -in. section always lowest in density?

MR. SAUNDERS: Yes, the original size of that specimen that we first used was  $\frac{3}{16}$ -in. thick. After having been cut out of the central portion of the stepped bar itself it left a piece perhaps  $\frac{3}{4}$  in. long and  $\frac{1}{2}$  in. wide. We used small samples. We did not want to use too small samples because we used the loss-of-weight-in-water method of determination which perhaps might not be too accurate. We actually used two different samples for checking the values and we reground them on an emery wheel in between determinations to slightly change the values by perhaps two grams.

There was some conjecture as to possibly some type of inclusion effect, but in each case the skin caused by the sand was ground off, and an attempt was made to get well below any obvious surface imperfections. But on slicing a sample in half, of course, it is difficult to see whether or not there are any tiny holes without a careful microscopic examination because a cutting wheel can sometimes smear the surface.

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<sup>4</sup> Battelle Memorial Institute, Columbus, Ohio.

E. L. DABLER<sup>2</sup>: I would like to suggest that there might be a critical value of  $\frac{3}{16}$  in. and that you turn your attention to surface tension and fluidity effects. Leading momentarily away from the thinking of the paper, to surface tension, fluidity, turbulence and motion of the liquid metal back and forth in the mold before solidification, you should find a given rate of flow of metal at the center compared to that at the mold face, for a given section. You would expect these factors to affect density, through their effect on temperature equilibrium alone.

H. J. ROAST<sup>3</sup>: Silicon bronze was used a great deal in the States and in Canada during the war for very critical applications. We used it on the railroads for the highest form of steam valve, where the temperature was very high on account of the high pressure of the steam.

There are always, of course, things against any alloy. The silicon bronze of about  $3\frac{1}{2}$  per cent silicon will undoubtedly gas in an oil-fired furnace with great facility if you have the wrong conditions. The worst gasing I have ever seen has existed with the silicon alloy. But there is no need to have it if you keep your oxidizing atmosphere in the right condition.

There is one thing we found, namely, that in casting in green sand the moisture content of the sand is very critical. We cast most of our work in dry sand. But if we had a green sand mold with some dry cores we would find that where we had a strip of green sand with too much moisture then when we took the casting out and machined it, the air was filled with brick dust material. It ran in our case to about  $\frac{1}{32}$  in., but it was completely soft and looked exactly like brick dust. That came from an excess of moisture over 6 per cent in the sand and always came when that condition existed. But there is no need to have this condition.

As to the pressure tests, I would confirm the speaker who said that 6,000 lb is readily met. We have made many such tests on the usual cylinder. This cast cylinder had a 1-in. wall originally when cast. It was machined by removing  $\frac{1}{3}$  from the inside,  $\frac{1}{3}$  from the outside, leaving only the center core. We made many castings that had no leaks when tested at 6,000 lb pressure. I noticed the casting temperature was given by one discussor as not higher than 2200 F. Wherever possible, we cast at 1900 F.

The test bar we used throughout was the keel-block test bar. This test bar brings out one of the useful features of silicon alloys, namely, that in the keel bar,  $1\frac{1}{2}$  in. square, you have a perfectly uniform grain, or almost perfectly uniform. There is very little finer grain on the outside when cast in a dry sand mold. That is not the case with the tin bronzes and accounts for the good pressure results which you get from the use of silicon bronze.

MR. STERN: With reference to the practice of melting, all our melts for pouring test specimens in the first part of the investigation were made under oxidizing atmosphere in an induction furnace. We found it was absolutely impossible to get a good alloy if we used a reducing atmosphere. The addition of charcoal to the melt, while preparing stock metal, would cause such severe gasing that when the ingot was finally poured it resembled a sponge.

J. I. MEDOFF<sup>4</sup>: In Fig. 16, density vs. size, your highest curve seemed to be for the  $\frac{5}{16}$ -in. bar. In line with chilling im-

proving the density, would that fit in with the same reasoning that the  $\frac{1}{8}$  in. is positively the highest?

MR. KRYNITSKY: In the case of  $\text{CO}_2$ , air and nitrogen, yes, the density values were highest for the  $\frac{5}{16}$ -in. section, but for those atmospheres the differences in density values of different sections are too small to draw any definite conclusions.

MR. SAUNDERS: As a matter of reiteration, the thought explaining the higher density for the  $\frac{1}{8}$ -in. section for the CO and hydrogen heats is that the chilling action, being more rapid for a thin section, retained the gases in a state of solid solution, which would not decrease the density.

G. P. HALLIWELL<sup>5</sup>: My remarks have only an indirect bearing on the data in this paper. Periodically foundrymen will encounter gassed castings of silicon bronze. The usual deoxidizers will not correct this condition, mainly because the source of trouble is a reducing gas. The addition of a deoxidizer to such a melt removes the oxygen and aggravates this condition by disturbing the hydrogen-oxygen equilibrium. Lithium is reputed as having an affinity for hydrogen as well as oxygen. Recently we attempted to evaluate its chemical activity in this respect. Ordinary silicon bronze (95-1-4) melted in an oil-fired crucible furnace and cast into green sand molds produced sound test bars with excellent properties. When lithium was added to the same pot of liquid metal (150 lb) the risers puffed and the resultant castings were porous with very poor properties. This does not mean that lithium poisoned the silicon bronze, but rather that it was an excellent deoxidizer, and reduced the oxygen content far below that necessary for equilibrium. Since this condition is characterized by an inverse relationship, additional hydrogen could be dissolved by the metal. Such hydrogen could come from the following reaction.



The water in the above reaction is that contained in the sand, which upon contact with the molten copper forms steam and copper oxide and liberates hydrogen. The latter could easily be sufficient to establish equilibrium conditions or even supersaturate the melt. In the case under consideration, hydrogen from the reaction was not required, since the metal when poured into a hot (400 F) cast iron mold puffed badly. Gas was evolved which burnt with an almost invisible flame, being tinged only with the green of the copper.

Silicon bronzes once gassed are extremely difficult to degas and the only dependable solution to the problem is to pig the metal and then remelt under oxidizing conditions.

#### AUTHOR'S CLOSURE

MR. KRYNITSKY (*Authors' Closure*): The authors appreciate the written contributions made on this discussion by Dr. Cyril Stanley Smith and Mr. J. E. Crown.

In interpreting the microstructure of the Cu plus 4.7 per cent Si alloy, as cast, on the basis of Dr. Smith's diagram, the authors have considered the product of the peritectoid reaction which occurs at 842 C.

The authors agree that it would be of practical value to obtain additional data pertaining to commercial copper-silicon alloy compositions. Accordingly, in the authors' opinion, as the next step in this research, a systematic study of the series of ternary alloys, particularly Cu-Si-Ni and Cu-Si-Zn alloys would be advisable.

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# A LABORATORY EVALUATION OF SOME AUTOMOTIVE CAST IRONS

By

Arthur B. Shuck\*

## ABSTRACT

*Cupola melted cast irons are examined to evaluate the effect of composition and cross section on mechanical and thermal properties, wear resistance and microstructure. Test bars, varying in cross section from one-fourth of an inch square to one and one-half inches square, were cast in green sand molds for each of the irons investigated. Several carbon-silicon base compositions were melted in each type of alloy to determine the effect of carbon and silicon content.*

## Introduction

THE EFFECT of cross section on the mechanical properties and metallurgy of cast iron often is not appreciated by the design engineer. The purpose of this paper is to summarize the results of an investigation of the properties of plain and alloy cast irons which have been used in piston rings, steam packings and pressure seals. The work was carried out in the Foundry and Research Laboratory of the Koppers Company, Inc., Piston Ring Division.

A variety of cast iron alloys have been used in piston rings and pressure packings in an attempt to produce wear resistant materials for applications where ordinary cast iron is inadequate. The extreme section sensitivity of some of the alloys makes it necessary to produce them in four or five carbon-silicon bases. This investigation had as its immediate objective the evaluation of some of the more regularly produced irons to eliminate those showing inferior properties and to gain a clearer knowledge of the effects of varying cross section.

## Experimental Procedure

All experimental work was carried out on horizontally-cast test bars, the pattern for which is shown in Fig. 1. These consisted of six 10-in., straight bars of approximately square cross section. The cross sectional dimensions of the different bars on the pattern varied as follows: (A) 1½ in. (B) 1 in., (C) ¾ in., (D) ½ in., (E) ⅜ in. and (F) ¼ in. The cross sectional contour of the bars is shown in Fig. 2. The patterns were mounted on a 16-inch square pattern

plate and the molds were made on a squeeze machine in 2¾-in. depth flasks using production foundry sand. The molds were stacked four high and poured from a runner box common to all six sprues. Chokes were provided in the gates of the four larger bars and a run-off riser was used at the extreme end of all of the bars to remove the first metal and prevent chill effects.

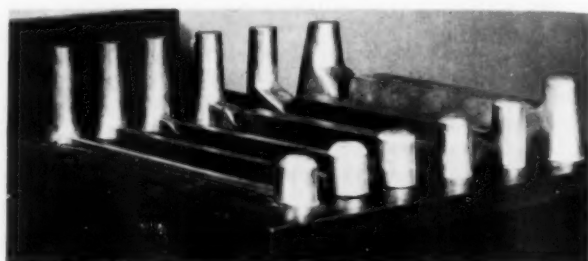


Fig. 1—Test coupon patterns for section sensitivity tests.

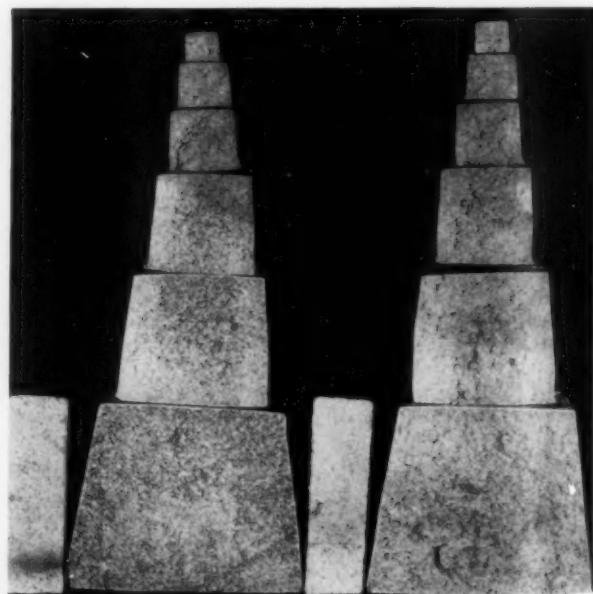


Fig. 2—Fractures of test coupons.

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These patterns produced metal more closely approximating that found in piston rings than the standard 0.875-in. and 1.20-in. A.S.T.M. test bars which are cast vertically and poured through a large sprue cup. The surface-volume ratios of the square section bars are more nearly those of the castings represented than those of round bars, producing more representative cooling rates. All test castings were x-rayed and those showing unsoundness were discarded.

### Chemical Analysis

Drilling and spectrographic samples were prepared on each section size from each heat. Carbon analyses were run on each sample to show variation in combined carbon content with varying cross section. Composite analyses are reported on silicon, sulphur, phosphorus, manganese and the alloys.

### Mechanical Properties

Tensile strengths on the three larger sections were obtained from standard 1/2-in. diameter test bars. Tension test specimens from the smaller cross sections were scaled down proportionately using a 3/8-in. diameter test specimen from the 1/2-in. casting, a 1/4-in. diameter test specimen from the 3/8-in. casting and a 0.180-in. diameter specimen from the 1/4-in. casting. Each tensile strength recorded was the average of three tests. Rockwell hardness was obtained on flat sections milled from the test coupons. Elastic modulus and impact strength were obtained on miniature rectangular bars measuring 4.0 x 0.260 x 0.140 in. The bars were loaded beamwise between centers 2.785 in. apart for elastic modulus. The results were plotted and a secant modulus of elasticity was computed at 50 per cent of the breaking load. Comparative impact tests were made on the bars in an inch-pound Izod impact tester.

### Wear Tests

Dry wear tests were carried out on the machine shown in Fig. 3. The test specimen was a rectangular block measuring 0.500 x 0.500 x 0.285 in. When the casting section was below this size, specimens were built up by riveting. A 0.500 x 0.285-in. surface was finished to between 5 and 10 micro-inches root-mean-square<sup>1</sup>. This surface was run brake-shoe-wise against a standardized 3 1/2-in. diameter cast iron drum. The surfaces of the drums were prepared by polishing out the machining thread marks with 320-grit emery cloth and finishing with grade 0 polishing paper. Both the specimens and the drums were carefully cleaned with carbon tetrachloride to remove all oil and grit. They were then weighed to the nearest milligram and assembled in the test rig with a 5-lb. load. The drums were rotated at 1725 rpm without lubrication for a period of one hour, after which the specimens and drums were removed from the test machine, reweighed and the wear was recorded as milligrams weight loss by the drums and the specimens. The tests were run

in duplicate for each material and were repeated if 10 milligram agreement was not obtained.

### Coefficient of expansion and thermal critical points

A Bristol-Rockwell dilatometer was used in the study of thermal expansion. Heating rate and cooling rate was 400 F per hour for all specimens with a soaking period of one hour at 1600 F. Coefficient of expansion was obtained between 100 F and 800 F. In the cast irons studied, this portion of the curve was approximately a straight line. Critical points were determined from the sharp contraction and expansion on heating and cooling. The permanent expansion of the 3-in. specimens after this cycle is also recorded.

In alloy 6, which is heat treated, a dilatometric study of the heat treatment was made. This consisted of a duplication of the standard heat treatment on the dilatometer, the specimen being quenched in place on the dilatometer. This gave information on the transformations and the permanent expansion taking place during the heat treatment. Further consideration will be given to this study under the section on alloy 6.

### Microstructures

Microscopic studies of the materials were made from specimens taken from the middle of the 10-in. long bars. Photographs were taken at a point midway between the center of the cross section and the cope side for comparison. Graphite size and type are described in accordance with the Tentative Recommended Practice for Evaluating the Microstructure of Graphite in

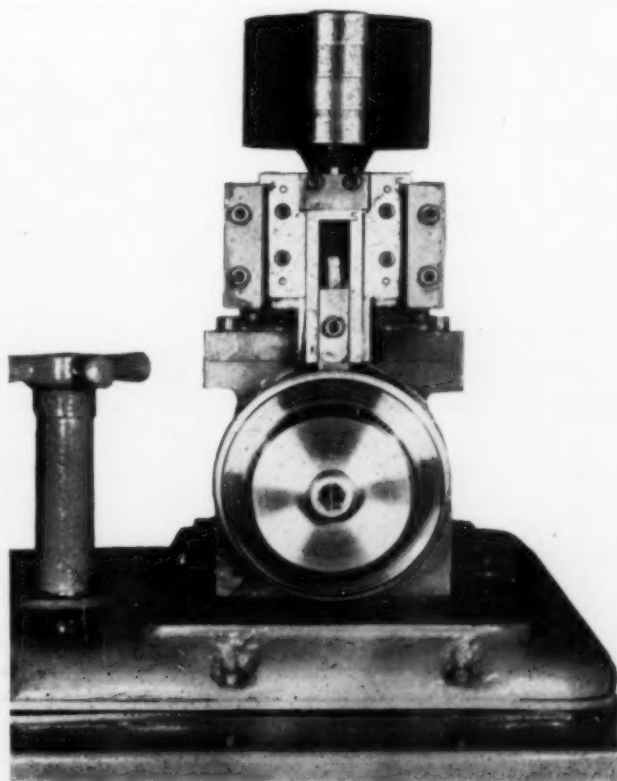


Fig. 3—Drum and specimen assembly dry wear test.

<sup>1</sup> An expression of surface finish or roughness in millionths of an inch as the square root of the average of squares of the deflection of a diamond-tipped tracer moved over the surface.

Gray Iron, A.S.T.M. Des. A 247-T. Studies of the matrices were made visually at 100X, 500X and 1000X over the entire cross section.

The following types of cast iron alloys were examined and are reported in this paper:

Type	Alloy	Number of Carbon-Silicon-Bases
1.	No Alloy	5
2.	Nickel-Chromium	4
3.	Nickel-Molybdenum	2
4.	Chromium-Molybdenum	4
5.	Molybdenum	2
6.	Nickel-Molybdenum-Chromium (Heat Treated)	4

The alloys were melted as part of production cupola runs in a 42-in. front slagging cupola equipped with blast heating and drying equipment. Blast temperatures were between 400 F and 600 F with a moisture content of less than 3 gr per cu. ft. Forehearth temperatures were measured at between 2750 F and 2900 F. Ladle inoculation was not used to prevent undercooling. The alloys were produced by additions of ferro-alloy at the pouring ladle and some of these show a slight inoculating effect. The metal was transferred from the forehearth to the molds in 400-lb. covered ladles. Temperature loss in the pouring ladles was between 100 F and 150 F.

TABLE I-A—CHEMICAL COMPOSITION—TYPE I—UNALLOYED CAST IRON

Heat No.	Cross Section, in. sq.	T.C.	G.C.	C.C.	Si	S	P	Mn	Cr
250A	1½	3.71	3.33	0.33					
B	1	3.67	3.07	0.60					
C	¾	3.69	3.04	0.65	2.69	0.060	0.51	0.50	0.14
D	½	3.75	3.03	0.72					
E	⅜	3.71	3.05	0.66					
F	¼	3.75	2.90	0.85					
285A	1½	3.55	3.11	0.44					
B	1	3.53	2.99	0.54					
C	¾	3.57	2.93	0.64	2.52	0.071	0.40	0.54	0.12
D	½	3.58	2.90	0.68					
E	⅜	3.56	2.87	0.69					
F	¼	3.57	2.84	0.73					
252A	1½	3.53	2.88	0.65					
B	1	3.57	2.77	0.80					
C	¾	3.52	2.78	0.74	2.05	0.072	0.37	0.58	0.14
D	½	3.61	2.81	0.80					
E	⅜	3.61	2.80	0.81					
F	¼	3.62	2.78	0.84					
253A	1½	3.48	2.83	0.65					
B	1	3.53	2.88	0.65					
C	¾	3.62	2.85	0.77	1.80	0.065	0.35	0.59	0.14
D	½	3.57	2.76	0.81					
E	⅜	3.57	2.79	0.78					
F	¼	3.58	2.62	0.96					
383A	1½	3.44	2.76	0.68					
B	1	3.41	2.61	0.80	1.57	0.062	0.24	0.58	0.08
C	¾	3.43	2.67	0.76					
D	½	Specimen ½ in. sq. and under were not machinable.							
E	⅜								
F	¼								

TABLE I-B—MECHANICAL PROPERTIES—TYPE I—UNALLOYED CAST IRON

Heat No.	Cross Section, in. sq.	Average Tensile Strength, psi	Hardness Rb	Elastic Modulus	Impact Strength in.-lb.	Wear Tests, loss, mg		
						Spec. Wear	Drum Wear	Total Wear
250A	1½	25,000	64	4.6x10 <sup>6</sup>	4.0	31.5	36.3	67.8
B	1	20,700	69	4.7x10 <sup>6</sup>	3.7	27.5	32.6	60.1
C	¾	24,000	84	7.1x10 <sup>6</sup>	3.0	38.7	32.2	70.9
D	½	27,000	91	9.3x10 <sup>6</sup>	2.5	31.1	31.2	62.3
E	⅜	35,000	93	10.3x10 <sup>6</sup>	2.5	35.4	31.2	66.6
F	¼	37,200	97	12.1x10 <sup>6</sup>	2.7	52.8	27.4	80.2
285A	1½	20,600	67	5.2x10 <sup>6</sup>	3.5	60.3	26.1	86.4
B	1	21,500	73	8.9x10 <sup>6</sup>	3.0	46.0	29.2	75.2
C	¾	27,500	87	9.5x10 <sup>6</sup>	3.0	41.3	34.4	75.7
D	½	34,300	95	13.2x10 <sup>6</sup>	3.0	50.5	28.1	78.7
E	⅜	35,000	99	13.2x10 <sup>6</sup>	3.5	53.8	25.5	79.3
F	¼	43,200	102	13.7x10 <sup>6</sup>	3.3	56.2	24.1	80.3
252A	1½	19,900	77	6.5x10 <sup>6</sup>	3.0	38.0	31.8	69.8
B	1	20,000	81	6.9x10 <sup>6</sup>	3.0	25.7	28.6	54.3
C	¾	25,000	88	8.3x10 <sup>6</sup>	2.7	35.9	31.5	67.4
D	½	32,100	94	8.3x10 <sup>6</sup>	3.0	45.6	24.9	70.5
E	⅜	35,900	96	11.5x10 <sup>6</sup>	2.5	59.2	19.4	78.7
F	¼	43,900	100	14.9x10 <sup>6</sup>	3.0	56.8	29.3	86.1
253A	1½	27,100	82	10.2x10 <sup>6</sup>	3.3	34.1	33.1	67.2
B	1	25,700	88	9.1x10 <sup>6</sup>	3.0	36.7	33.8	70.5
C	¾	29,900	92	10.2x10 <sup>6</sup>	3.3	36.0	35.9	71.9
D	½	35,900	97	12.9x10 <sup>6</sup>	2.9	64.2	25.5	89.7
E	⅜	43,100	101	13.8x10 <sup>6</sup>	3.4	75.0	39.8	114.8
F	¼	38,000	105	19.3x10 <sup>6</sup>	3.3	78.1	47.2	125.3
283A	1½	33,000	84	11.2x10 <sup>6</sup>	3.8	30.6	38.3	68.9
B	1	33,200	89	11.5x10 <sup>6</sup>	3.6	41.8	34.6	76.4
C	¾	35,100	94	12.3x10 <sup>6</sup>	3.0	43.5	45.4	88.9
D	½							
E	⅜							
F	¼							

Test specimens less than ½-in. square were not machineable



TABLE 1-C—MICROSTRUCTURES AND THERMAL PROPERTIES—TYPE 1—UNALLOYED CAST IRON

Heat No.	Cross Section, in. sq.	Graphite AFA-ASTM	Matrix	Microstructure	Remarks	Coefficient of Expansion Per °F.	Lower Critical Heating, °F.	Permanent Expansion, in 3 in., in.
250A	1½	3-A	Ferrite—50% Pearlite—50%	Secondary Ferrite <sup>2</sup>				
B	1	3-4-A	Ferrite—25% Pearlite—75%	Secondary Ferrite <sup>2</sup>				
C	¾	4-5-AB	Ferrite—10% Pearlite—90%	Secondary Ferrite <sup>2</sup>				
D	½	4-5-AB	Ferrite—2% Pearlite—98%	Secondary Ferrite <sup>2</sup>				
E	⅜	5-AB	Ferrite—2% Pearlite—98%	Secondary Ferrite <sup>2</sup>		6.44x10 <sup>-6</sup>	1410	0.026
F	¼	6-7-AB	Pearlite—100%					
285A	1½	3-A	Ferrite—70% Pearlite—30%	Secondary Ferrite				
B	1	4-A	Ferrite—30% Pearlite—70%	Secondary Ferrite				
C	¾	5-A	Ferrite—25% Pearlite—75%	Secondary Ferrite				
D	½	6-A	Ferrite—2% Pearlite—98%	Secondary Ferrite		6.07x10 <sup>-6</sup>	1400	0.0184
E	⅜	6-7-A	Pearlite—100%					
F	¼	7-8-AB	Pearlite—97% Ferrite—3%	Eutectiform Ferrite <sup>2</sup>				
252A	1½	3-A	Ferrite—10% Pearlite—90%	Secondary Ferrite				
B	1	4-A	Ferrite—2% Pearlite—98%	Secondary Ferrite				
C	¾	5-A	Ferrite—2% Pearlite—98%	Secondary Ferrite		6.35x10 <sup>-6</sup>	1440	0.021
D	½	5-AB	Pearlite—100%					
E	⅜	6-8-AD	Pearlite—100%					
F	¼	6-8-D	Pearlite—97% Ferrite—3%	Eutectiform Ferrite				
253A	1½	3-4-A	Ferrite—5% Pearlite—95%	Secondary Ferrite				
B	1	4-5-A	Pearlite—100%					
C	¾	5-A	Pearlite—100%			6.20x10 <sup>-6</sup>	1420	0.018
D	½	5-6-B	Pearlite—100%					
E	⅜	6-7-B	Pearlite—98% Ferrite—2%	Eutectiform Ferrite				
F	¼	8-D	Pearlite—98% Ferrite—2%	Eutectiform Ferrite				
283A	1½	3-4-A	Massive cementite	Cementite needles				
B	1	4-5-A	Pearlite—100%					
C	¾	5-A	Pearlite—100%					
D	½	5-6-B	Pearlite—90% Ledeburite—10%	Eutectiform Ferrite		6.20x10 <sup>-6</sup>	1420	0.018
E	⅜	6-7-B	Pearlite—30% Ledeburite—70%					
F	¼	8-D	Pearlite—10% Ledeburite—90%					

<sup>2</sup> The term "secondary ferrite" will be used throughout to designate the ferrite associated with normal flake graphite; "eutectiform ferrite" to designate ferrite occurring with eutectiform (Type D) graphite.

#### Analytical Data

##### Type 1 Unalloyed Cast Iron

Type 1 alloy is an unalloyed gray cast iron. Field data has shown it to be highly wear resistant in applications where lubrication is good or moderate, provided it possesses a satisfactory microstructure. In the following test the five carbon-silicon bases are as they were taken from the cupola. The chromium content is residual from the scrap iron and the steel used in the charge. See Tables 1-A, 1-B and 1-C.

**Microstructure.** The unalloyed type 1 iron is pearlitic, the coarseness of the pearlite increases with the cross-sectional dimensions of the casting. Tables 1-D and 1-E show the types of graphite, occurrence of massive carbides and distribution of primary ferrite pearlite and secondary ferrite according to cross-sections. See Fig. 4, 5, 6, 7 and 8.

#### Observations

1. In each case where eutectiform ferrite and Type D graphite occurred in the cross-section, total wear

increased to above 80 milligrams and specimen wear was in excess of 50 milligrams.

2. Wear decreased as graphite flake size increased in the absence of primary ferrite.

3. Wear increased slightly with the occurrence of large amounts of secondary ferrite in the larger cross-sections.

4. Tensile strength, hardness and elastic modulus increased with decreasing cross-section associated with a generally finer structure of the iron and the increased combined carbon content in the smaller cross-sections. This is effective only to the point where massive carbide or Type D graphite begins to form.

#### Type 2 Nickel-Chromium Alloy Cast Iron

Type 2 alloy is a nickel-chromium alloy used in

more severe applications than Type 1 alloy. The alloy content of the four different materials listed is similar; the difference being in the carbon and silicon content. See Tables 2-A, 2-B and 2-C.

**Microstructures.** There is insufficient alloy in this material to change the pearlitic matrix structure. The principal benefit of the alloy addition was to stabilize the combined carbon content of the material. Only 10 per cent primary ferrite occurs in Heat No. 341-A in the 1½-in. cross-section, the chromium stabilizing the combined carbon. The relatively high nickel content prevents the formation of massive carbides in the smaller cross-sections. Type D graphite occurred only in ¼-in. square, cross-section F. This cast iron was the least section sensitive of the irons tested. See Tables 2-D and 2-E and Fig. 9, 10, 11 and 12.

TABLE 1-D—OCCURRENCE OF GRAPHITE TYPES OF MASSIVE CEMENTITE IN VARIOUS CROSS-SECTIONS\*—TYPE 1—UNALLOYED CAST IRON

Heat No.	Types A and B Graphite	Type D Graphite	Massive Carbides
	Cross Section	Cross Section	Cross Section
250	A, B, C, D, E, F		
285	A, B, C, D, E, F	F	
252	A, B, C, D, E, F	F	
253	A, B, C, D, E	E, F	F
283	A, B, C	C, D, E, F	D, E, F

\* Cross Section: A—1½ in. sq.; B—1 in. sq.; C—¾ in. sq.; D—½ in. sq.; E—⅜ in. sq.; F—¼ in. sq.

TABLE 1-E—MATRIX IN VARIOUS CROSS-SECTIONS\*—TYPE 1—UNALLOYED CAST IRON

Heat No.	Secondary Ferrite and Pearlite	Pearlite Only Cross Section	Eutectiform Ferrite and Pearlite
	Cross Section		
250	A, B, C, D, E		
285	A, B, C, D	E	F
282	A, B, C	D, E	F
253	A	B, C, D,	E, F
283		A, B	C, D, E, F

\* Cross Section: A—1½ in. sq.; B—1 in. sq.; C—¾ in. sq.; D—½ in. sq.; E—⅜ in. sq.; F—¼ in. sq.

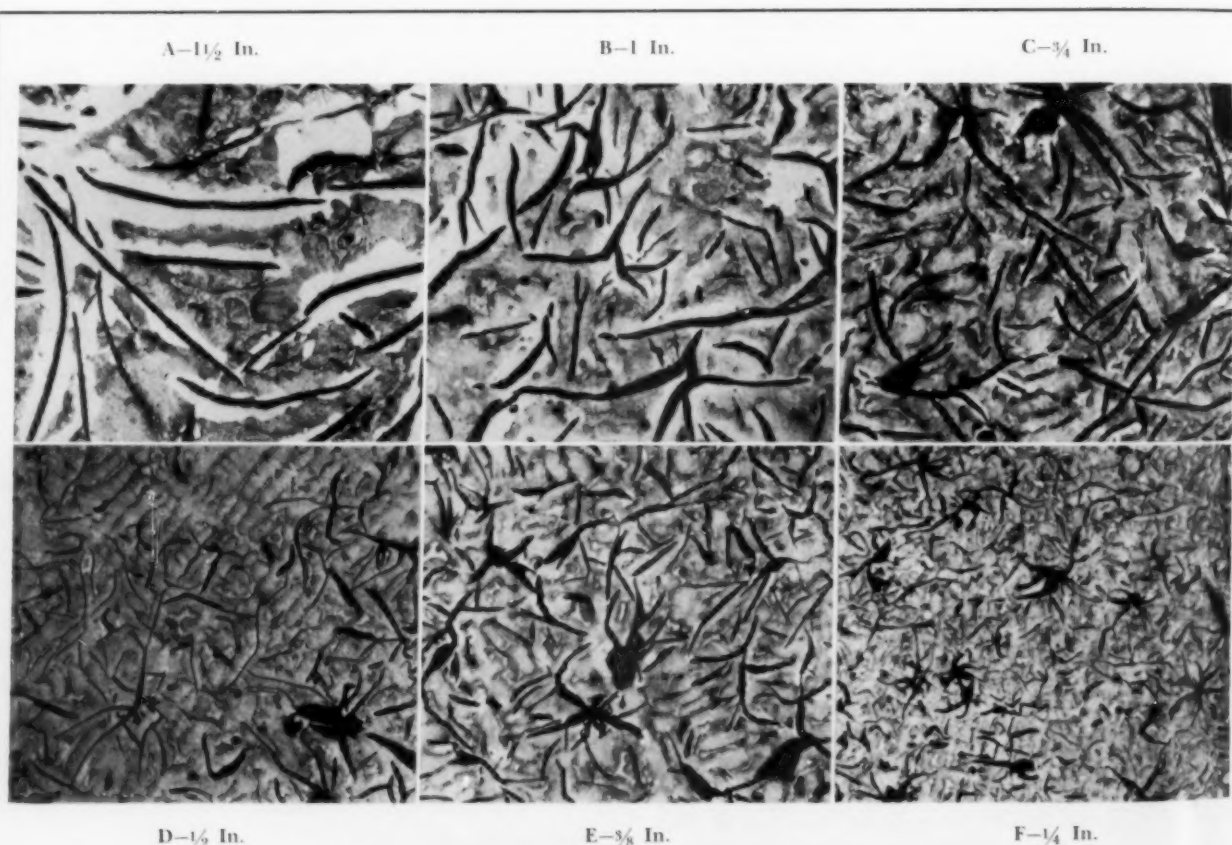


Fig. 4—Unalloyed iron, Heat No. 250, Mag. 100X, picral etch.

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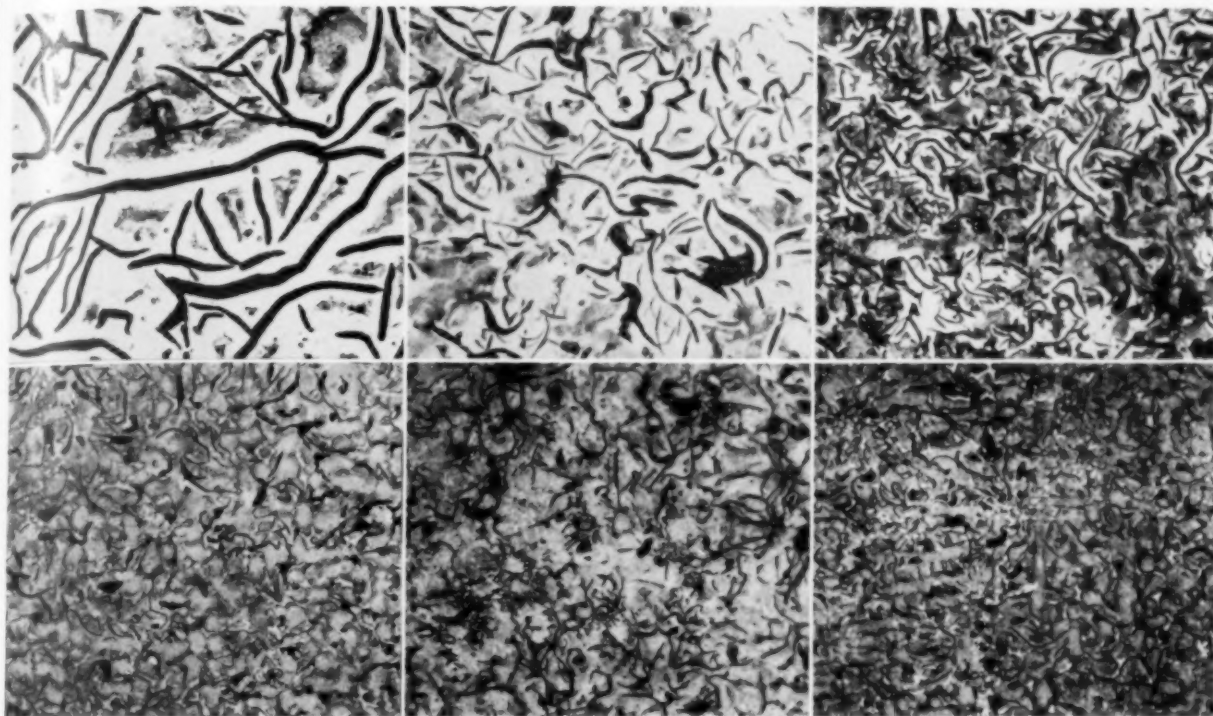
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B-1 In.

C-¾ In.



D-½ In.

E-¾ In.

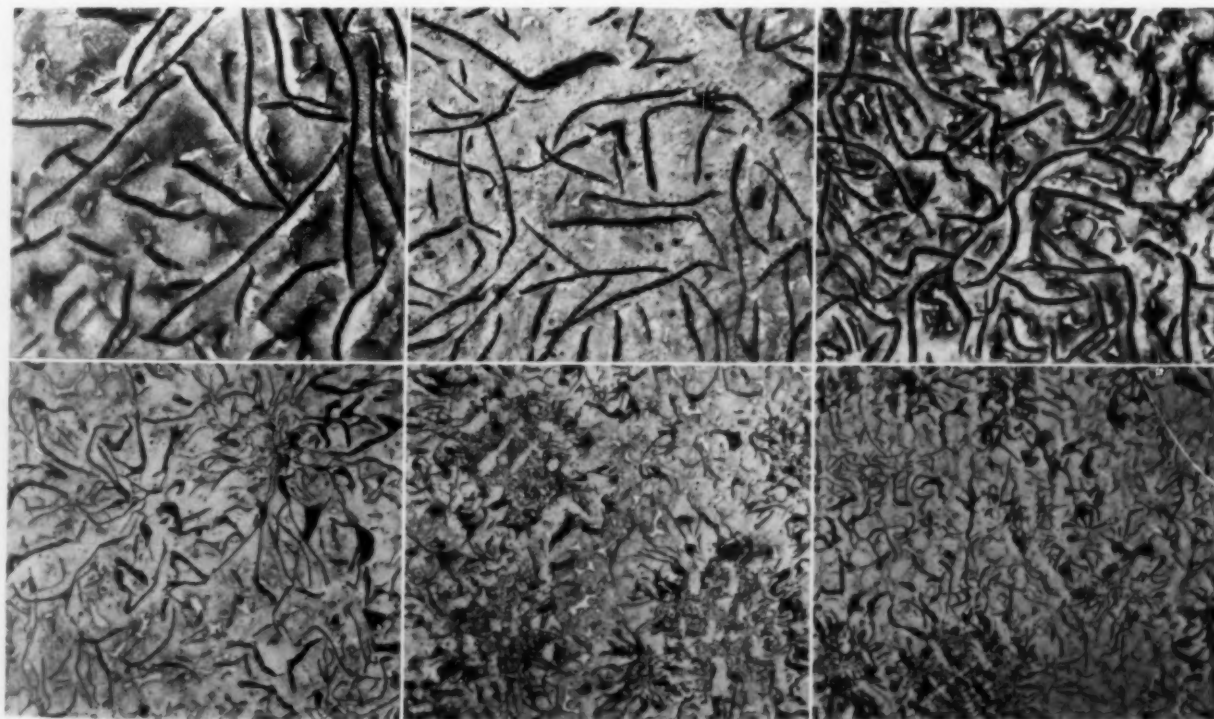
F-¼ In.

Fig. 5—Type 1, unalloyed iron, Heat No. 285. Mag. 100X, picral etch.

A-1½ In.

B-1 In.

C-¾ In.



D-½ In.

E-¾ In.

F-¼ In.

Fig. 6—Type 1, unalloyed iron, Heat No. 252. Mag. 100X, picral etch.



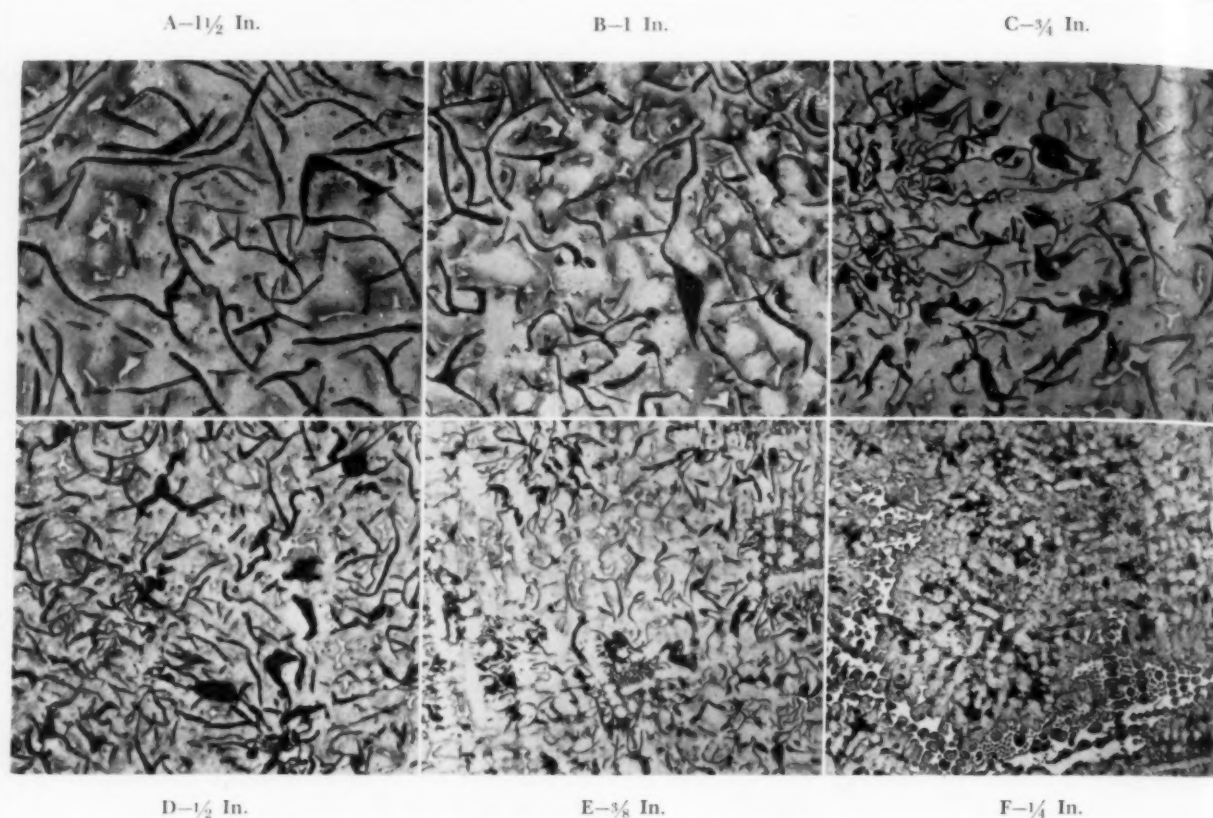


Fig. 7—Type 1, unalloyed iron, Heat No. 253. Mag. 100X, picral etch.

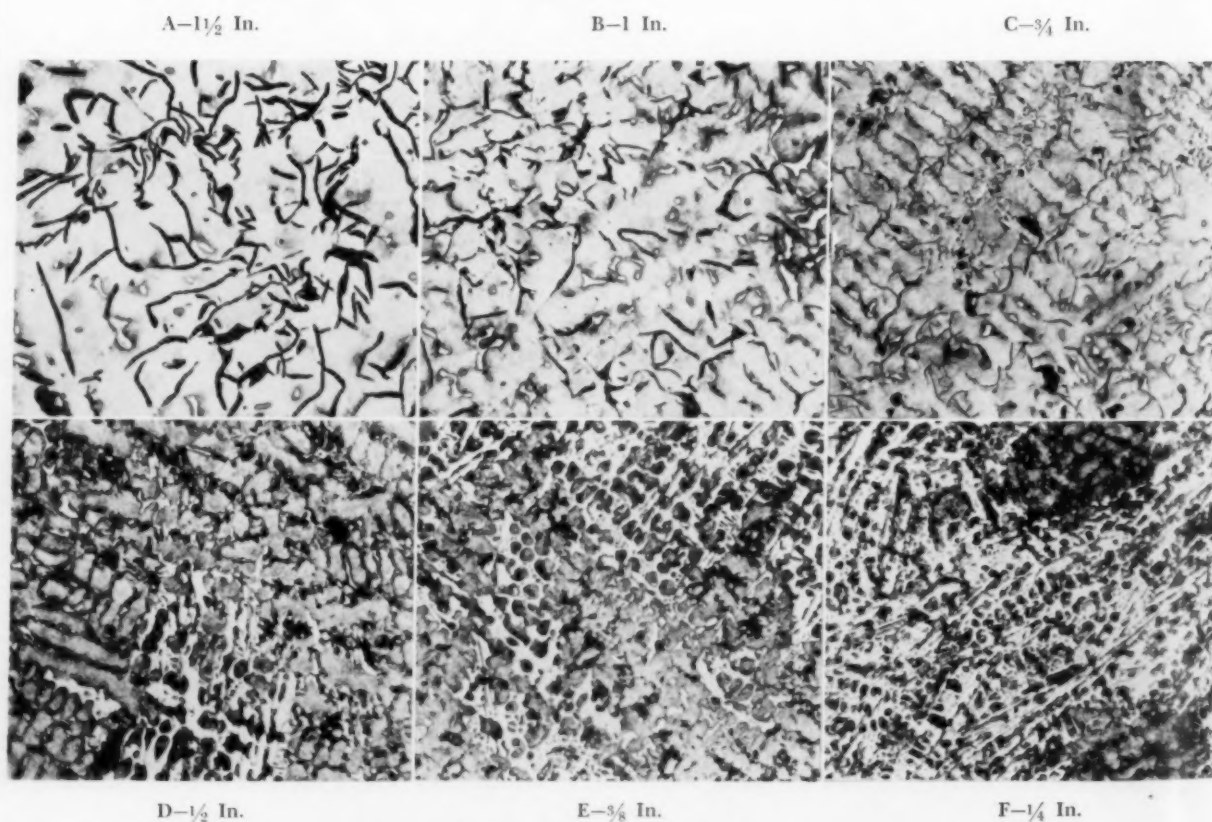


Fig. 8—Type 1, unalloyed iron, Heat No. 383. Mag. 100X, picral etch.

**Observations**

1. The conclusions reached on the effect of the micro-structure on the wear and physical properties of type 1 alloy are also true in type 2 alloy.

2. The chromium and nickel alloy additions tend to stabilize the pearlitic phase and to increase the section sensitivity, reducing the amount of eutectiform and secondary ferrite and the massive carbide formation.

3. A comparison of the tensile strength of type 2 alloy with the strength of the unalloyed material in the proper cross-section is shown in Table 2-F. The average increase in strength with the addition of the chromium-nickel alloy was approximately 26 per cent above that of unalloyed iron.

4. Because of the greater uniformity of micro-structure and smaller tendency for the formation of primary ferrite in the heavy sections and eutectiform

TABLE 2-A—TYPE 2—CHROMIUM NICKEL ALLOY CAST IRON

Heat No.	Cross Section, in. sq.	T.C.	G.C.	C.C.	Si	S	P	Mn	Cr	Ni
341A	1½	3.71	3.25	0.46				0.60	0.32	1.17
B	1	3.69	3.13	0.56				0.60	0.31	1.17
C	¾	3.71	3.14	0.57	3.06	0.072	0.46	0.61	0.31	1.18
D	½	3.73	3.10	0.65				0.62	0.29	1.22
E	⅜	3.77	3.10	0.67				0.61	0.29	1.25
F	¼	No bar ran.								
378A	1½	3.67	3.10	0.57					0.30	1.19
B	1	3.78	3.15	0.63					0.30	1.06
C	¾	3.75	3.06	0.69	2.44	0.074	0.26	0.59	0.27	1.11
D	½	3.79	3.13	0.66					0.29	1.13
E	⅜	3.79	3.07	0.72					0.28	1.09
F	¼	3.77	3.01	0.76					0.28	1.08
342A	1½	3.55	2.87	0.68				0.54	0.35	1.43
B	1	3.59	2.87	0.72				0.53	0.33	1.43
C	¾	3.59	2.83	0.76	2.13	0.077	0.33	0.57	0.35	1.48
D	½	3.59	2.82	0.77				0.56	0.35	1.45
E	⅜	3.62	2.87	0.75				0.58	0.37	1.50
F	¼	3.58	2.76	0.82				0.57	0.37	1.44
371A	1½	3.33	2.62	0.71				0.53	0.37	1.44
B	1	3.35	2.62	0.73				0.53	0.34	1.40
C	¾	3.41	2.56	0.85	1.61	0.079	0.34	0.55	0.35	1.34
D	½	3.43	2.54	0.89				0.54	0.33	1.43
E	⅜	3.41	2.54	0.87				0.54	0.32	1.51
F	¼									

TABLE 2-B—MECHANICAL PROPERTIES—TYPE 2—CHROME-NICKEL ALLOYS

Heat No.	Cross Section, in. sq.	Average Tensile Strength, psi	Hardness Rb	Elastic Modulus	Impact Strength, in.-lb.	Wear Tests, loss, mg.		
						Spec. Wear	Drum Wear	Total Wear
341A	1½	25,800	87	9.2x10 <sup>6</sup>	2.2	32.5	37.1	69.6
B	1	26,100	85	8.1x10 <sup>6</sup>	2.0	35.3	29.8	65.1
C	¾	28,300	90	8.9x10 <sup>6</sup>	2.1	35.6	39.7	75.3
D	½	30,600	92	9.1x10 <sup>6</sup>	2.2	55.4	33.2	88.6
E	⅜	45,100	97	9.1x10 <sup>6</sup>	2.0	51.0	28.9	80.0
F	¼							
378A	1½	25,000	79	6.5x10 <sup>6</sup>	2.7	31.1	39.7	70.8
B	1	31,500	88	7.8x10 <sup>6</sup>	2.5	29.3	30.1	59.4
C	¾	31,500	94	8.9x10 <sup>6</sup>	3.0	34.7	34.8	69.5
D	½	41,000	100	9.6x10 <sup>6</sup>	3.3	42.9	35.4	78.3
E	⅜	42,100	101	10.4x10 <sup>6</sup>	3.0	46.1	28.3	74.4
F	¼	54,100	104	12.6x10 <sup>6</sup>	3.5			
342A	1½	24,900	88	7.0x10 <sup>6</sup>	2.5	35.2	37.7	72.9
B	1	27,900	89	9.4x10 <sup>6</sup>	2.7	36.1	38.4	74.5
C	¾	40,400	95	11.4x10 <sup>6</sup>	3.0	31.9	40.6	72.5
D	½	41,800	99	13.5x10 <sup>6</sup>	3.0	41.8	39.2	81.0
E	⅜	44,200	101	13.4x10 <sup>6</sup>	3.0	47.3	27.1	74.4
F	¼			16.2x10 <sup>6</sup>				
371A	1½	39,600	90	12.1x10 <sup>6</sup>	3.5	24.7	51.1	75.8
B	1	42,600	93	13.5x10 <sup>6</sup>	3.0	35.9	51.2	87.1
C	¾	46,400	99	15.0x10 <sup>6</sup>	3.0	31.4	39.7	71.1
D	½	51,300	102	14.7x10 <sup>6</sup>	3.0	35.7	38.8	74.5
E	⅜	52,500	104	16.5x10 <sup>6</sup>	4.0	40.4	35.5	74.9
F	¼							

TABLE 2-C—MICROSTRUCTURE AND THERMAL PROPERTIES—TYPE 2—CHROME-NICKEL ALLOYS

Heat No.	Cross Section, in. sq.	Graphite AFA-ASTM	Microstructures		Coefficient of Expansion, per °F	Lower Critical Heating, °F	Permanent Expansion in 3 In. In.
			Matrix	Remarks			
341A	1½	3-A	Pearlite—90% Ferrite—10%				
B	1	3-4-A	Pearlite—95% Ferrite—5%				
C	¾	4-A	Pearlite—98% Ferrite—2%				
D	½	5-6-AB	Pearlite—98% Ferrite—2%		6.27x10 <sup>-6</sup>	1480	0.019
E	¾	5-7-AE	Pearlite—100%				
F	¼						
378A	1½	4-A	Pearlite—98% Ferrite—2%				
B	1	4-A	Pearlite—100%				
C	¾	4-5-A	Pearlite—100%		6.18x10 <sup>-6</sup>	1410	0.015
D	½	5-AB	Pearlite—100%				
E	¾	5-6-AB	Pearlite—100%				
F	¼	7-8-AD	Pearlite—95% Ferrite—5%	Eutectiform ferrite			
342A	1½	2-3-A	Pearlite—100%				
B	1	4-A	Pearlite—100%				
C	¾	4-5-A	Pearlite—100%		6.58x10 <sup>-6</sup>	1405	0.012
D	½	5-6-AB	Pearlite—100%				
E	¾	6-7-AB	Pearlite—100%				
F	¼	7-8-AD	Pearlite—98% Ferrite—2%	Eutectiform ferrite			
371A	1½	3-A	Pearlite—100%				
B	1	4-A	Pearlite—100%				
C	¾	4-5-AB	Pearlite—100%		6.80x10 <sup>-6</sup>	1380	0.009
D	½	6-A	Pearlite—100%				
E	¾	6-8-ABD	Pearlite—99% Ferrite—1%	Eutectiform ferrite			
F	¼						

TABLE 2-D—OCCURRENCE OF GRAPHITE TYPES AND MASSIVE CARBIDE IN VARIOUS CROSS-SECTIONS\* AND ANALYSES OF TYPE 2—CHROMIUM-NICKEL ALLOYS

Heat No.	Flake Graphite Cross Section	Type D, Graphite Cross Section	Massive Carbides Cross Section
341	A, B, C, D, E, (F?)	None	None
378	A, B, C, D, E, F	F	None
342	A, B, C, D, E, F	F	None
371	A, B, C, D, E, F	F	None

Section A—1½ in. sq.; B—1 in. sq.; C—¾ in. sq.; D—½ in. sq.; E—¾ in. sq.; F—¼ in. sq.

TABLE 2-E—MATRIX CONDITION IN VARIOUS CROSS-SECTIONS\* AND ANALYSIS OF TYPE 2—CHROMIUM-NICKEL ALLOY

Heat No.	Primary Ferrite and Pearlite Cross Section	Pearlite Only Cross Section	Eutectiform Ferrite and Pearlite Cross Section
341	A, B, C, D	E (F?)	
378	A	B, C, D, E	F
342		A, B, C, D, E	F
371		A, B, C, D	E, (F?)

Section A—1½ in. sq.; B—1 in. sq.; C—¾ in. sq.; D—½ in. sq.; E—¾ in. sq.; F—¼ in. sq.

TABLE 2-F—COMPARISON OF TENSILE STRENGTH (PSI) OF TYPE 2—CHROMIUM-NICKEL ALLOY

Cross Section in. sq.	A Type 2 Material	B Type 1 Material	A/B
¼	45,100	37,200	1.21
¾	42,000	35,000	1.20
½	41,800	32,000	1.30
¾	46,400	35,100	1.31
Averages	43,800	36,300	1.26

TABLE 2-G—COMPARISON OF ELASTIC MODULUS—TYPE 2—CHROMIUM-NICKEL ALLOYS AND TYPE 1—UNALLOYED CAST IRON

Cross Section in. sq.	A Type 2 Material	B Type 1 Material	A/B
¼	9.1x10 <sup>6</sup>	10.3x10 <sup>6</sup>	0.88
¾	10.4x10 <sup>6</sup>	13.2x10 <sup>6</sup>	0.79
½	13.5x10 <sup>6</sup>	8.3x10 <sup>6</sup>	1.60
¾	15.0x10 <sup>6</sup>	10.2x10 <sup>6</sup>	1.47
Averages	11.9x10 <sup>6</sup>	10.5x10 <sup>6</sup>	1.18



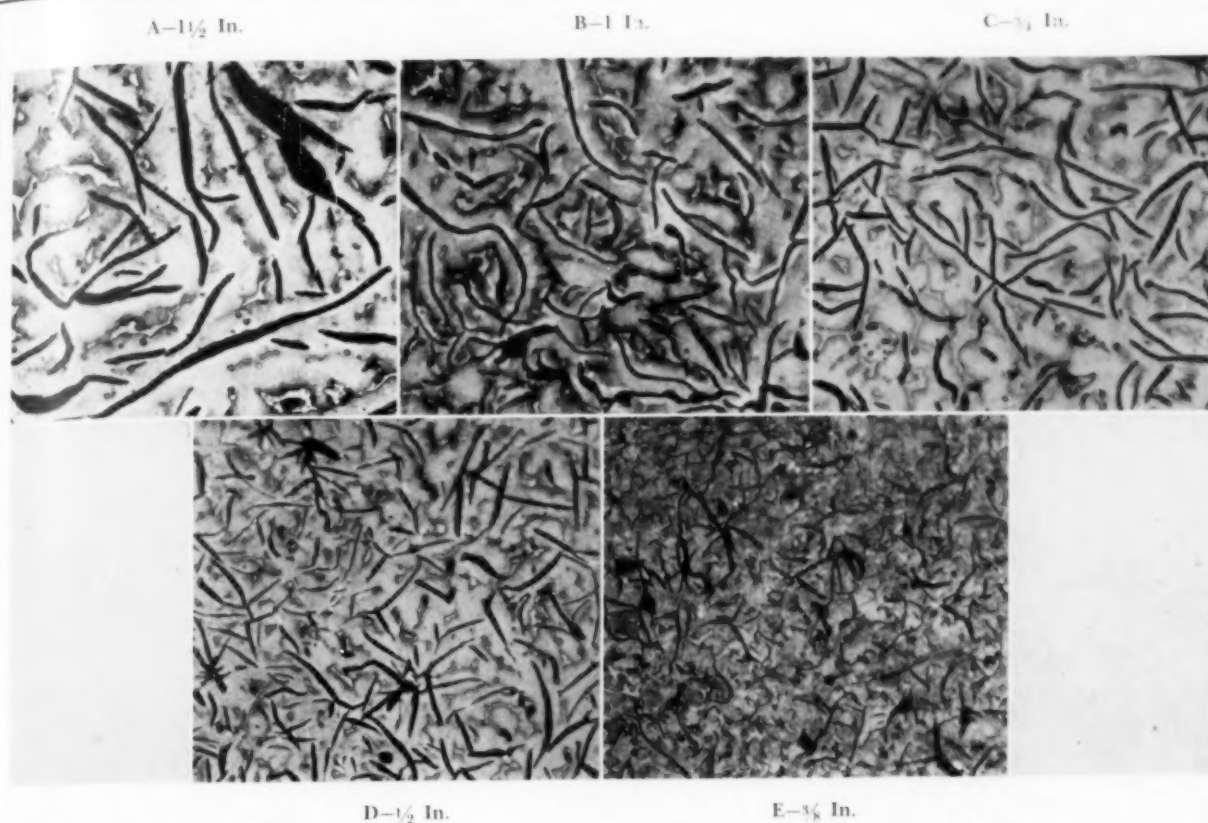


Fig. 9—Type 2, nickel-chromium alloy iron, Heat No. 341. Mag. 100X, picral etch.

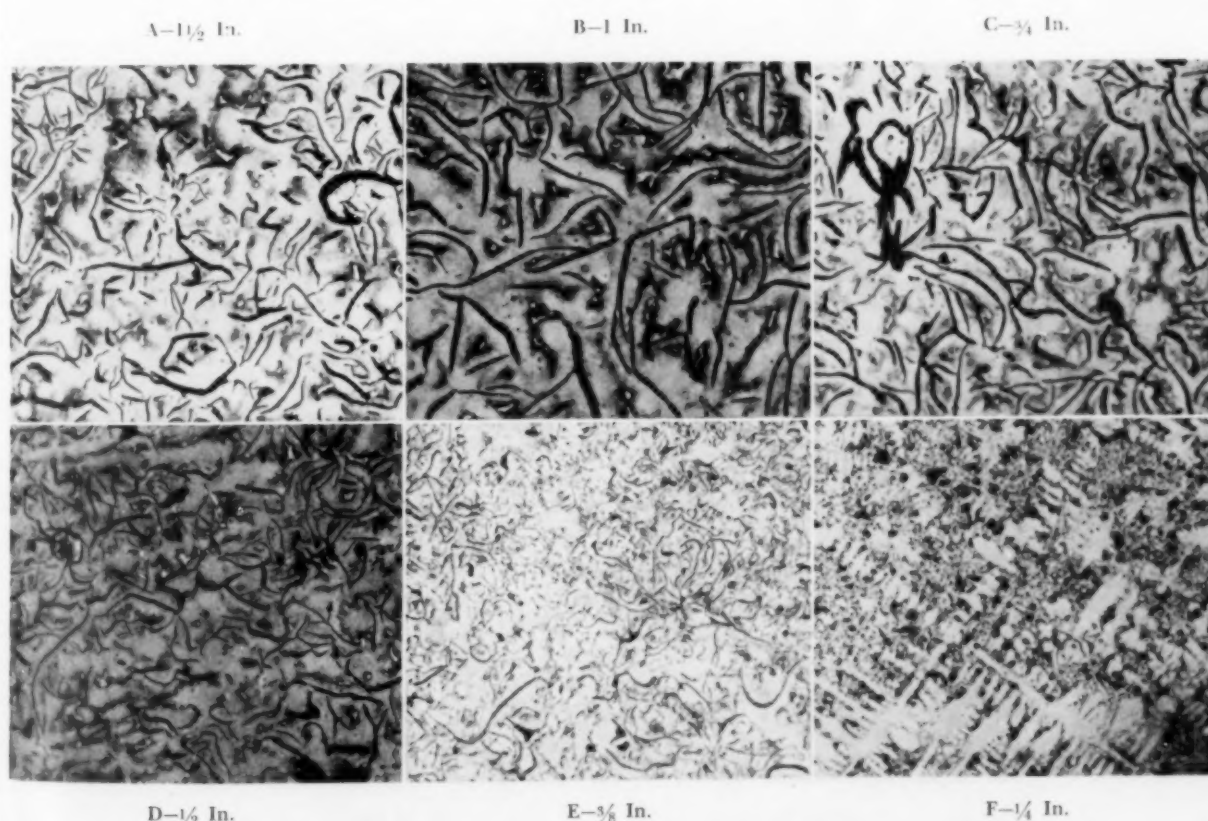


Fig. 10—Type 2, nickel-chromium alloy iron, Heat No. 378. Mag. 100X, picral etch.

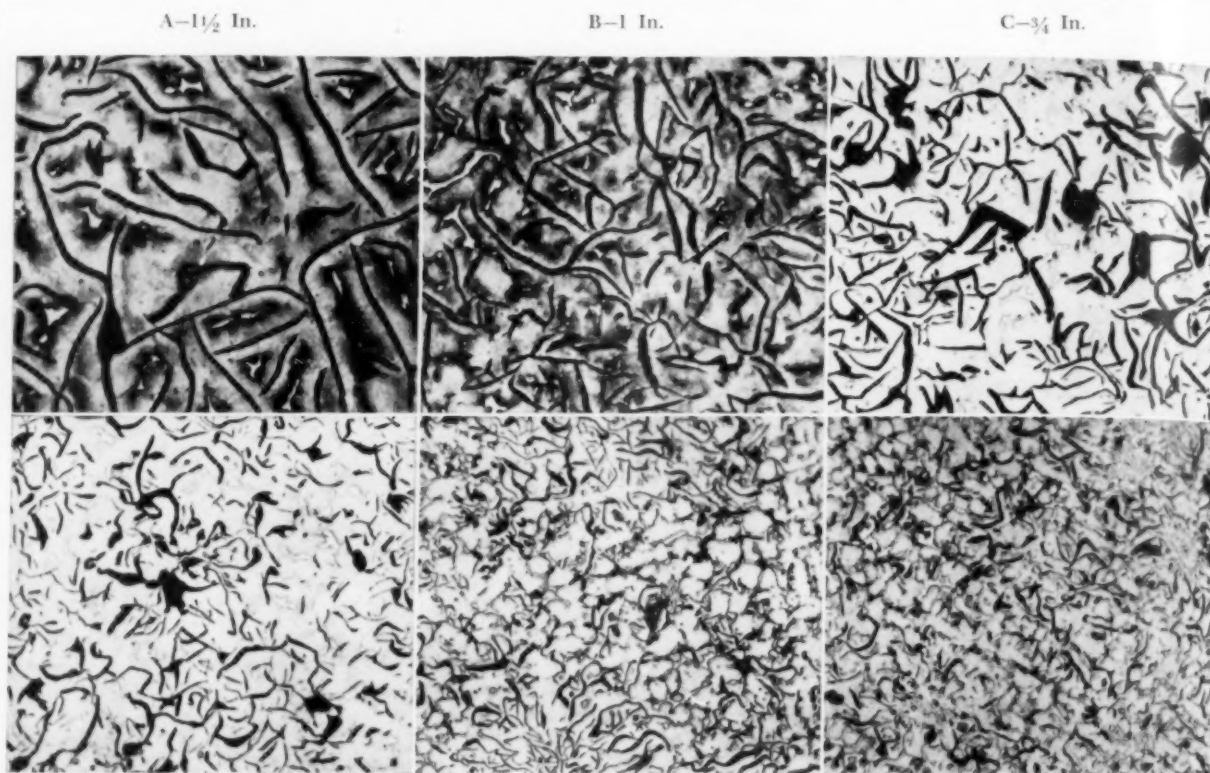


Fig. 11—Type 2, nickel-chromium alloy iron, Heat No. 342. Mag. 100X, picral etch.

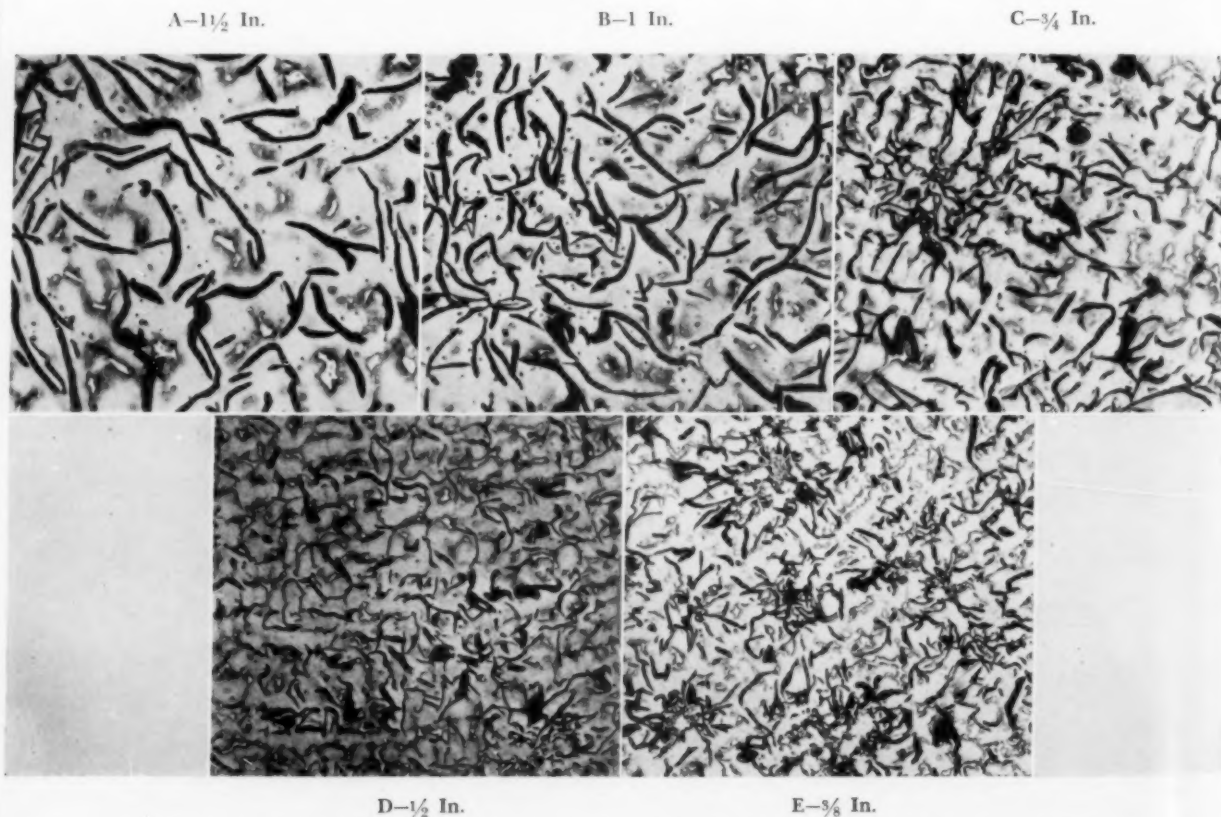


Fig. 12—Type 2, nickel-chromium alloy iron, Heat No. 371. Mag. 100X, picral etch.

TABLE 3-A—CHEMICAL COMPOSITION—TYPE 3—NICKEL-MOLYBDENUM ALLOY

Heat No.	Cross Section, in. sq.	T.C.	G.C.	C.C.	Si	S	P	Mn	Ni	Mo
373A	1½	3.28	2.73	0.55					1.30	0.51
B	1	3.31	2.75	0.56					1.38	0.50
C	¾	3.35	2.71	0.64	2.37	0.073	0.34	0.54	1.45	0.47
D	½	3.40	2.67	0.73					1.37	0.49
E	⅜	3.37	2.65	0.72					1.37	0.47
F	¼	Non-machinable								
344A	1½	3.41	2.95	0.46				0.55	1.79	0.66
B	1									
C	¾	3.43	2.69	0.74	1.96	0.078	0.30	0.55	1.77	0.68
D	½	3.44	2.66	0.78				0.54	1.84	0.58
E	⅜	3.42	2.69	0.73				0.54	1.75	0.58
F	¼	3.43	2.60	0.83				0.55	1.79	0.59

TABLE 3-B—MECHANICAL PROPERTIES—TYPE 3—NICKEL-MOLYBDENUM ALLOY

Heat No.	Cross Section, in. sq.	Average Tensile Strength, psi	Hardness Rb	Elastic Modulus	Impact Strength in.-lb.	Wear Tests, loss, mg.		
						Spec. Wear	Drum Wear	Total Wear
373A	1½	34,500	88	16.0x10 <sup>6</sup>	3.0	40.2	35.3	75.5
B	1	38,000	95	16.5x10 <sup>6</sup>	3.0	37.4	33.0	70.4
C	¾	41,900	98	16.0x10 <sup>6</sup>	4.3	36.0	36.7	72.7
D	½	51,000	99	17.4x10 <sup>6</sup>	4.5	33.1	41.6	74.7
E	⅜	47,500	104	16.7x10 <sup>6</sup>	4.5	36.5	39.3	75.8
F	¼	Not machinable						
344A	1½	31,300	91	13.4x10 <sup>6</sup>	3.0	28.4	36.2	64.6
B	1	No test casting available						
C	¾	42,800	96	14.2x10 <sup>6</sup>	6.0	27.9	38.7	66.6
D	½	51,800	103	15.6x10 <sup>6</sup>	5.0	31.2	39.1	70.3
E	⅜	52,200	106	16.9x10 <sup>6</sup>	5.8	35.7	44.0	79.7
F	¼	Not machinable						

TABLE 3-C—MICROSTRUCTURES AND THERMAL PROPERTIES—TYPE 3—NICKEL-MOLYBDENUM ALLOY

Heat No.	Cross Section, in. sq.	Graphite AFA-ASTM	Matrix	Remarks	Coefficient of Expansion, Per °F.	Lower Critical Heating, °F.	Permanent Expansion in 3 In., In.
373A	1½	3-A 100%	Ferrite—10% Pearlite—90%				
B	1	4-5-A	Ferrite—5% Pearlite—95%				
C	¾	5-6-A	Ferrite—5% Pearlite—95%				
D	½	6-A	Ferrite—10% Acicular—5% Pearlite—85%		6.59x10 <sup>-6</sup>	1410	0.015
E	⅜	6-8-A	Ferrite—5% Acicular—60% Pearlite—35%				
F	¼	6-8-AD	Ferrite—5% Acicular—90% Pearlite—5%	Eutectiform ferrite			
344A	1½	3-4-A	Ferrite—2% Pearlite—98%				
B	1						
C	¾	5-A	Acicular—90% Pearlite—10%		6.38x10 <sup>-6</sup>	1400	0.012
D	½	5-8-AD	Acicular—100%				
E	⅜	6-8-AD	Acicular—100%				
F	¼	7-8-AD	Acicular—100%				



TABLE 3-D—DISTRIBUTION OF MATRIX CONSTITUENTS—VARIOUS CROSS-SECTIONS OF TYPE 3—NICKEL-MOLYBDENUM ALLOY

Heat No.	Pearlite and Secondary Ferrite	Acicular + Pearlite + Ferrite	Acicular + Pearlite	Acicular Only
373	A, B, C	D, E, F	—	—
344	A	(B?)	C	D, E, F

TABLE 3-E—DISTRIBUTION OF GRAPHITE IN VARIOUS CROSS-SECTIONS—TYPE 3—NICKEL-MOLYBDENUM ALLOY

Heat No.	Flake Graphite Cross-Section	Type D Graphite Cross-Section	Massive Carbide
373	A, B, C, D, E, F	F	—
344	A, C, D, E, F	E, F	—

TABLE 3-F—COMPARISON OF TENSILE STRENGTH, ELASTIC MODULUS AND IMPACT RESISTANCE IN TYPE 3—NICKEL-MOLYBDENUM ALLOY AND TYPE 1—UNALLOYED CAST IRON

Heat No.	Section	Tensile Strength		Elastic Modulus		Impact Strength	
		A Type 3	B Type 1	C Type 3	D Type 1	Type 3	Type 1
373	1/2	51,000	32,100	17.4x10 <sup>6</sup>	8.3x10 <sup>6</sup>	4.5	3.0
344	3/4	42,800	35,100	14.2x10 <sup>6</sup>	10.2x10 <sup>6</sup>	6.0	3.0
Averages		45,800	33,600	15.8x10 <sup>6</sup>	9.3x10 <sup>6</sup>	5.3	3.0

ferrite and Type D graphite formation in the smaller sections, wear was more uniform than it was in the unalloyed cast iron. Type 1 cast iron ordinarily showed only one of the sections entirely pearlitic while the matrices of the type 2 alloys have three or more sections entirely pearlitic and only a very small amount of ferrite in the other sections.

5. The modulus of elasticity is higher in the type 2 alloyed material than in the type 1 unalloyed cast iron in the larger cross-sections. Comparison is given in Table 2-G.

#### Type 3 Nickel Molybdenum Alloy Cast Iron

The third alloy tested was a nickel-molybdenum

alloy containing 1.30 to 1.85 per cent nickel and 0.50 to 0.70 per cent molybdenum. This is acicular material used to withstand severe impact. Chemical analyses, mechanical properties, and microstructures are shown in Tables 3-A, 3-B, 3-C, 3-D, 3-F.

**Microstructures.** Sufficient nickel and molybdenum are present in the type 3 alloy to produce an acicular matrix structure in the as-cast condition in the proper cross-sections. In Heat No. 373, with a silicon content of 2.73 per cent, the acicular structure was produced in all sections less than 1/2 in. square, in Heat No. 344 with 1.96 per cent silicon and approximately 1.75 per cent nickel, the acicular structure was produced in all sections under 3/4 in. square. This matrix gives

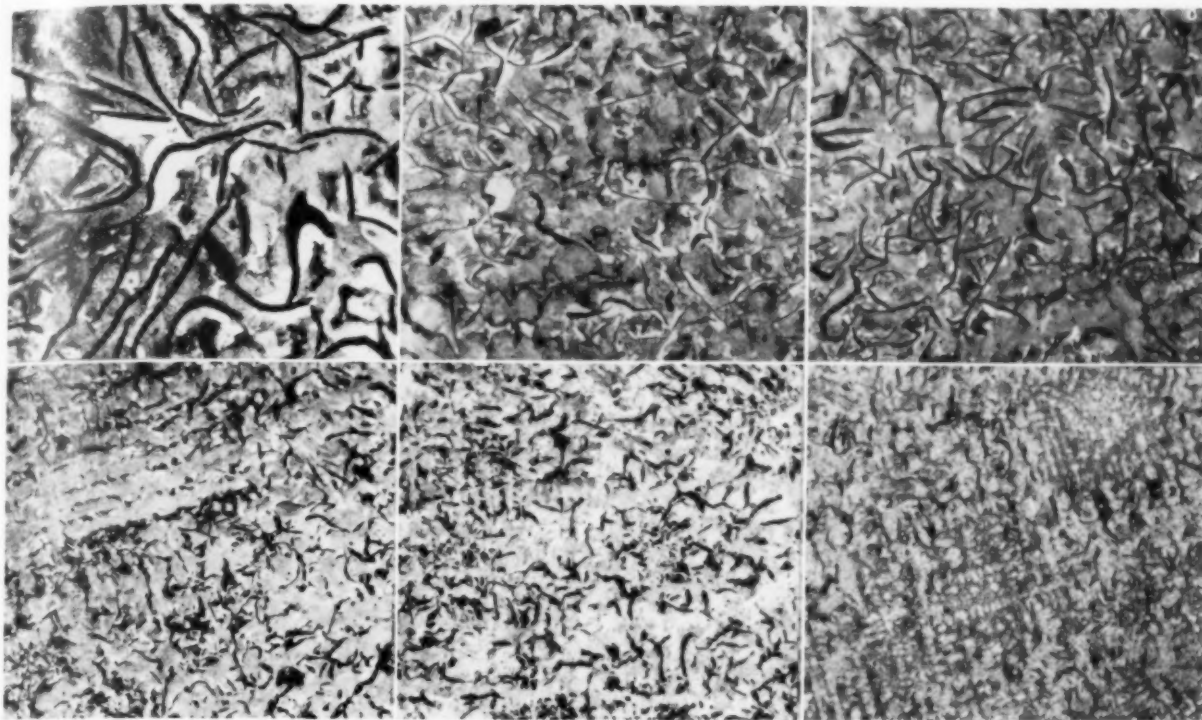
TABLE 4-A—CHEMICAL COMPOSITION—TYPE 4—CHROMIUM MOLYBDENUM ALLOY

Heat No.	Cross Section, in sq.	T.C.	G.C.	C.C.	Si	S	P	Mn	Cr.	Mo
310A	1 1/2	3.65	3.20	0.45				0.59	0.36	0.56
B	1	3.66	3.08	0.48				0.62	0.35	0.56
C	3/4	3.68	3.05	0.63				0.63	0.32	0.47
D	1/2	3.76	3.05	0.71	3.20	0.053	0.55	0.62	0.32	0.45
E	3/8	3.79	3.11	0.68				0.62	0.34	0.44
F	1/4	3.80	3.12	0.68				0.62	0.31	0.41
311A	1 1/2	3.59	2.97	0.62						
B	1	3.43	2.75	0.68						
C	3/4	3.55	2.84	0.71	2.95	0.064	0.43	0.51	0.35	0.34
D	1/2	3.55	2.88	0.67						
E	3/8	3.59	2.90	0.69						
F	1/4	3.58	2.80	0.78						
312A	1 1/2	3.34	2.70	0.60						
B	1	3.36	2.73	0.63						
C	3/4	3.38	2.69	0.69						
D	1/2	3.40	2.67	0.73	2.33	0.067	0.34	0.60	0.31	0.25
E	3/8	3.39	2.63	0.76						
F	1/4	3.39	2.65	0.74						
313A	1 1/2	3.71	3.03	0.68						
B	1	3.78	3.00	0.78						
C	3/4	3.77	3.01	0.76	1.72	0.078	0.31	0.63	0.46	0.41
D	1/2	3.77	2.98	0.79						
E	3/8	3.77	2.87	0.80						
F	1/4	Unmachinable								

A-1½ In.

B-1 In.

C-¾ In.



D-½ In.

E-⅜ In.

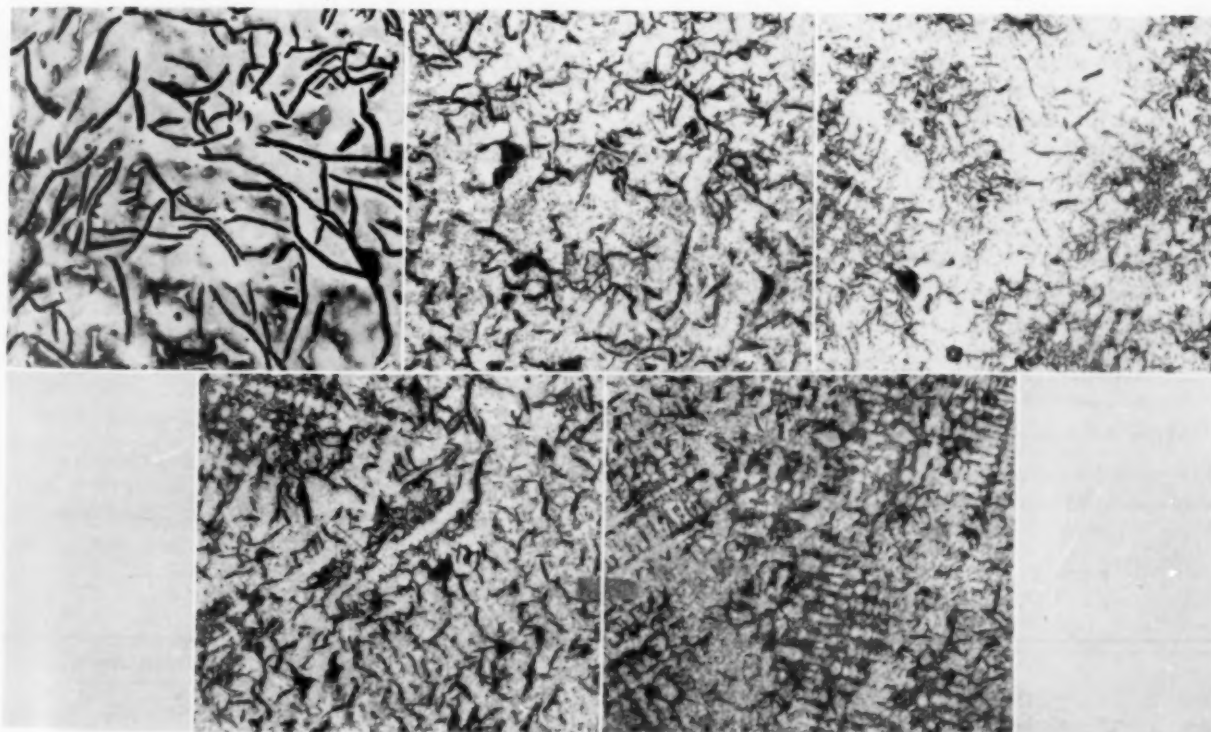
F-¼ In.

Fig. 13—Type 3, nickel-molybdenum alloy iron, Heat No. 373. Mag. 100X, picral etch.

A-1½ In.

B-1 In.

C-¾ In.



D-½ In.

E-⅜ In.

Fig. 14—Type 3, nickel-molybdenum alloy iron, Heat No. 344. Mag. 100X, picral etch.

TABLE 4-B—MECHANICAL PROPERTIES—TYPE 4—CHROMIUM MOLYBDENUM ALLOY

Heat No.	Cross Section, in. sq.	Average Tensile Strength, psi	Hardness Rb	Elastic Modulus	Impact Strength, in.-lb.	Spec. Wear	Wear Test, loss, mg.	
							Drum Wear	Total Wear
310A	1½	26,300	90	8.7x10 <sup>6</sup>	2.5	27.5	34.6	62.1
B	1	23,200	83	9.7x10 <sup>6</sup>	2.0	43.3	34.1	77.4
C	¾	26,900	93	10.8x10 <sup>6</sup>	2.3	42.6	37.6	80.2
D	½	43,900	96	12.0x10 <sup>6</sup>	2.5	32.9	44.9	77.6
E	¾	45,400	98	12.4x10 <sup>6</sup>	2.3	42.6	37.6	87.3
F	¼	48,600	101	12.5x10 <sup>6</sup>	2.5	103.8	34.6	138.4
311A	1½	24,500	83	11.5x10 <sup>6</sup>	2.3			
B	1	26,200	86	10.2x10 <sup>6</sup>	2.5	41.9	33.1	75.0
C	¾	24,700	89	11.2x10 <sup>6</sup>	2.5	40.4	29.0	69.4
D	½	38,400	93	12.6x10 <sup>6</sup>	2.5	44.0	32.5	76.5
E	¾	46,800	96	14.1x10 <sup>6</sup>	2.7	66.7	31.3	98.0
F	¼	42,200	102	16.6x10 <sup>6</sup>	2.5	98.2	27.1	125.3
312A	1½	36,400	90	9.3x10 <sup>6</sup>	3.5	33.6	33.4	67.0
B	1	38,200	94	12.7x10 <sup>6</sup>	3.5	50.1	23.7	73.0
C	¾	41,400	96	13.4x10 <sup>6</sup>	3.0	46.8	32.8	79.6
D	½	50,500	98	13.4x10 <sup>6</sup>	3.2	48.5	37.7	88.2
E	¾	50,100	99	15.0x10 <sup>6</sup>	3.0			
F	¼	63,800	104	16.2x10 <sup>6</sup>	3.8	53.7	37.8	91.5
313A	1½	29,700	87	8.5x10 <sup>6</sup>	3.5	37.3	31.6	68.9
B	1	26,900	90	10.7x10 <sup>6</sup>	3.5	30.9	33.5	64.4
C	¾	38,100	96	11.1x10 <sup>6</sup>	4.0	42.6	41.5	84.1
D	½	52,000	97			33.2	47.8	81.0
E	¾	50,900	102			36.6	57.3	93.9
F	¼	Not machineable						

the alloy an increased resistance to impact over the pearlitic irons. Both Type 1 and Type D graphite were observed in this iron. The types of graphite and matrix structure according to cross-section and analysis are listed in Table 3-D and 3-E. See Fig. 13 and 14.

#### Observations

1. Wear tests show the material to be about equal to the unalloyed iron in resistance to wear.

2. Table 3-F is a comparison of tensile strength, elastic modulus, impact resistance for type 3 material and type 1 unalloyed cast iron.

3. The structure of the Type 3 alloy is very sensitive to changes in cross-section. Ordinarily the acicular structure is produced in cross-sections of different sizes by varying the amount of nickel and molybdenum added. In a 1-in. cross-section it is necessary to add more nickel and molybdenum than in a ¾-in. cross-section in order to produce the completely acicular matrix structure.

#### Type 4 Chrome Molybdenum Alloy Cast Iron

The type 4 alloy contains from 0.25 to 0.50 per cent of molybdenum and 0.30 to 0.60 per cent of chrom-

ium. Both the chromium and molybdenum actively stabilize the carbides and by holding silicon to a low level, graphitization is reduced. This iron has been successfully used in applications where temperatures up to 800 F are encountered and excessive expansion and creep is observed with the unalloyed iron.

Tables 4-A, 4-B, 4-C show the chemical composition and mechanical properties and microstructures in four different carbon-silicon levels of this alloy.

**Microstructure.** Graphite and carbide distribution by cross-section is shown in Table 4-D.

TABLE 4-G—COMPARISON OF PERMANENT EXPANSION IN TYPE 4—CHROMIUM MOLYBDENUM ALLOY AND TYPE 1—UNALLOYED CAST IRON

Heat No.	Permanent Growth in 3 In., in. Type 4 Material	Permanent Growth in 3 In., in. Type 1 Material
310	0.0105	0.0260
311	0.0135	0.0184
312	0.0121	0.0210
313	0.0074	0.0180

TABLE 4-F—COMPARISON OF TENSILE STRENGTH AND ELASTIC MODULUS IN TYPE 4—CHROMIUM MOLYBDENUM ALLOY AND TYPE 1—UNALLOYED CAST IRON

Heat No.	Cross Section, in. sq.	Tensile Strength, psi		Elastic Modulus	
		A Type 4 Material	B Type 1 Material	C Type 4 Material	D Type 1 Material
310	¼	48,600	37,200	12.5x10 <sup>6</sup>	12.1x10 <sup>6</sup>
311	¾	46,800	35,000	14.1x10 <sup>6</sup>	13.2x10 <sup>6</sup>
312	½	50,500	32,100	13.4x10 <sup>6</sup>	8.3x10 <sup>6</sup>
313	¾	38,100	29,900	11.1x10 <sup>6</sup>	12.3x10 <sup>6</sup>



TABLE 4-C—MICROSTRUCTURE AND THERMAL PROPERTIES—TYPE 4—CHROMIUM MOLYBDENUM ALLOY

Heat No.	Cross Section, in. sq.	Graphite AFA-ASTM	Microstructure		Coefficient of Expansion, Per °F.	Lower Critical, °F.	Permanent Expansion in 3 In., In.
			Matrix	Remarks			
310A	1½	2-3-A	Pearlite-60% Secondary Ferrite-40%				
B	1	3-A	Pearlite-85% Ferrite-15%				
C	¾	3-4-A	Pearlite-95% Ferrite-5%		6.54x10 <sup>-6</sup>	1500	0.0105
D	½	4-AB	Pearlite-99% Ferrite-1%				
E	⅜	5-6-AB	Pearlite-100%				
F	¼	50% 7-8-AD 50% 4-B					
311A	1½	2-3-A	Pearlite-85% Ferrite-15%				
B	1	3-A	Pearlite-90% Ferrite-10%				
C	¾	3-4-A	Pearlite-99% Ferrite-1%		6.20x10 <sup>-6</sup>	1500	0.0135
D	½	5-AB	Pearlite-98% Ferrite-2%	Eutectiform ferrite			
E	⅜	5-6-AB	Pearlite-99% Ferrite-1%	Eutectiform ferrite Eutectiform ferrite			
F	¼	7-8-AD	Pearlite-99% Ferrite-1%	Eutectiform ferrite			
312A	1½	4-5-A	Pearlite-98% Ferrite-2%				
B	1	4-5-A	Pearlite-100%		6.05x10 <sup>-6</sup>	1490	0.0121
C	¾	5-6-A	Pearlite-100%				
D	½	6-7-ABD	Pearlite-98% Ferrite-2%	Eutectiform ferrite			
E	⅜	7-ABD	Pearlite-98% Ferrite-2%	Cellular phosphide			
F	¼	7-8-DE	Pearlite-95% Ferrite-5%	Cellular phosphide			
313A	1½	3-4-A	Pearlite-100%				
B	1	4-A	Pearlite-100%				
C	¾	5-A	Pearlite-100%		6.0x10 <sup>-6</sup>	1420	0.0074
D	½	5-6-AB	Pearlite-98% Ferrite-2%	Eutectiform ferrite			
E	⅜	6-8-ABD	Pearlite-92% Ferrite-2%	Eutectiform ferrite			
F	¼	7-8-D	Massive Carbide-6% Pearlite-10% Acicular-25% Ferrite-20% Massive Carbide-45%	Eutectiform ferrite			

TABLE 4-D—DISTRIBUTION OF GRAPHITE AND MASSIVE CARBIDE IN VARIOUS CROSS-SECTIONS OF TYPE 4—CHROMIUM MOLYBDENUM ALLOY

Heat No.	Flake Graphite Present in Cross-Section	Whorl Graphite Present in Cross-Section	Secondary Graphite Present in Cross-Section	Massive Carbide in Cross-Section
310	A, B, C, D, E, F	D, E	F	
311	A, B, C, D, E, F	D, E	F	
312	A, B, C, D, E	D, E	F	
313	A, B, C, D	D, E	F	E, F

TABLE 4-E—DISTRIBUTION OF MATRIX CONSTITUENTS IN VARIOUS CROSS-SECTIONS—TYPE 4—CHROMIUM MOLYBDENUM ALLOY

Heat No.	Ferrite + Pearlite	Pearlite Only	Pearlite + Eutectiform ferrite	Pearlite + Acicular + Secondary Ferrite
310	A, B, C, D	E, F		
311	A, B, C		D, E, F	
312	A	B, C	D, E, F	
313		A, B, C,	D, E	F

The type of matrix by cross-section is as shown in Table 4-E. See Fig. 15, 16, 17 and 18.

#### Observations

1. The material shows a decided tendency to form eutectiform ferrite and graphite. This type of struc-

ture gives a fairly high hardness, strength, and modulus of elasticity, but very poor wearing properties.

2. A comparison of strength and elastic modulus for type 4 alloy and the unalloyed cast iron in the proper cross-sections is as in Table 4-F.

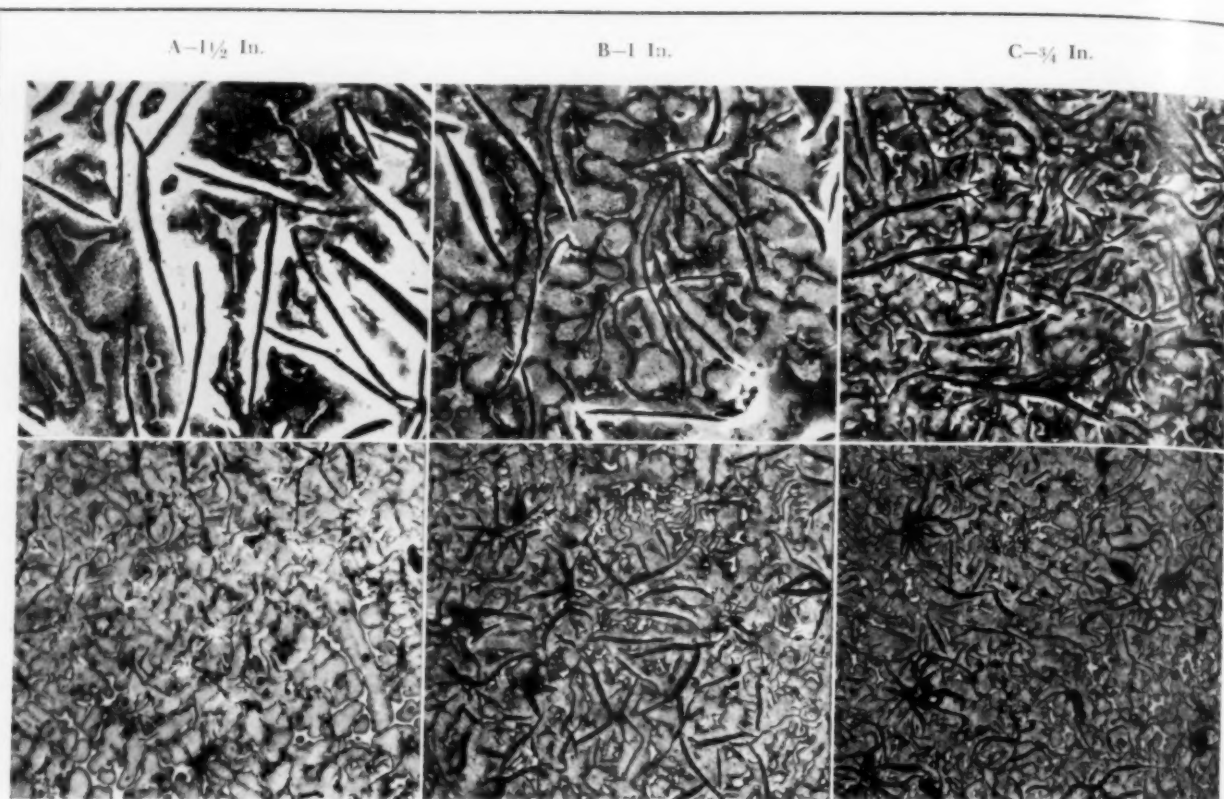


Fig. 15—Type 4, chromium-molybdenum alloy iron, Heat No. 310. Mag. 100X, picral etch.

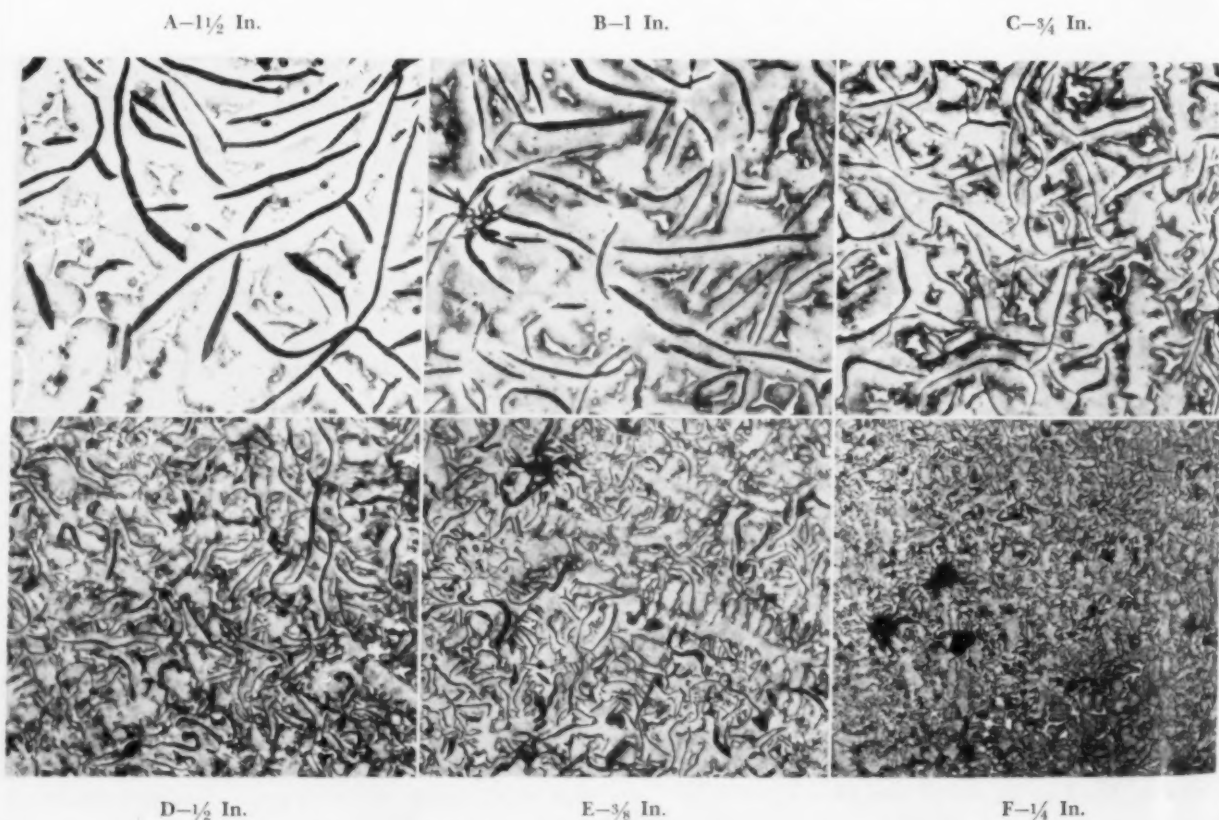


Fig. 16—Type 4, chromium-molybdenum alloy iron, Heat No. 311. Mag. 100X, picral etch.

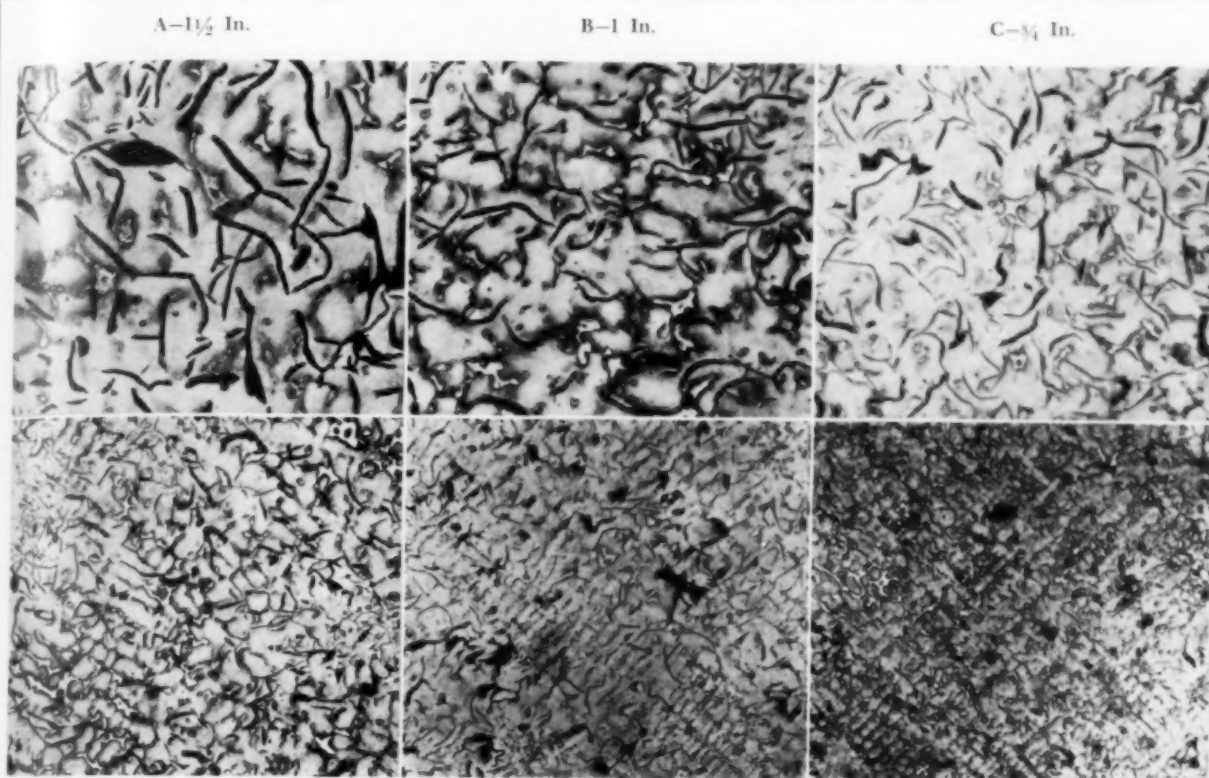


Fig. 17—Type 4, chromium-molybdenum alloy iron, Heat No. 312. Mag. 100X, picral etch.

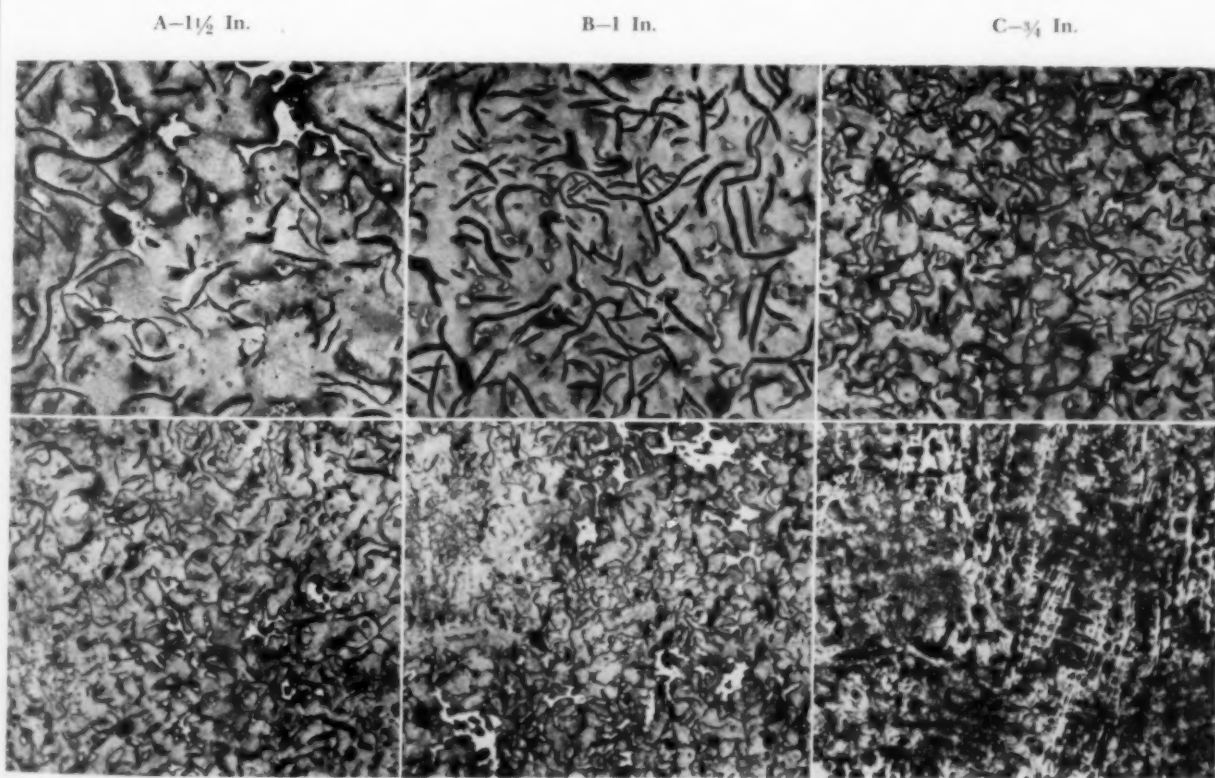
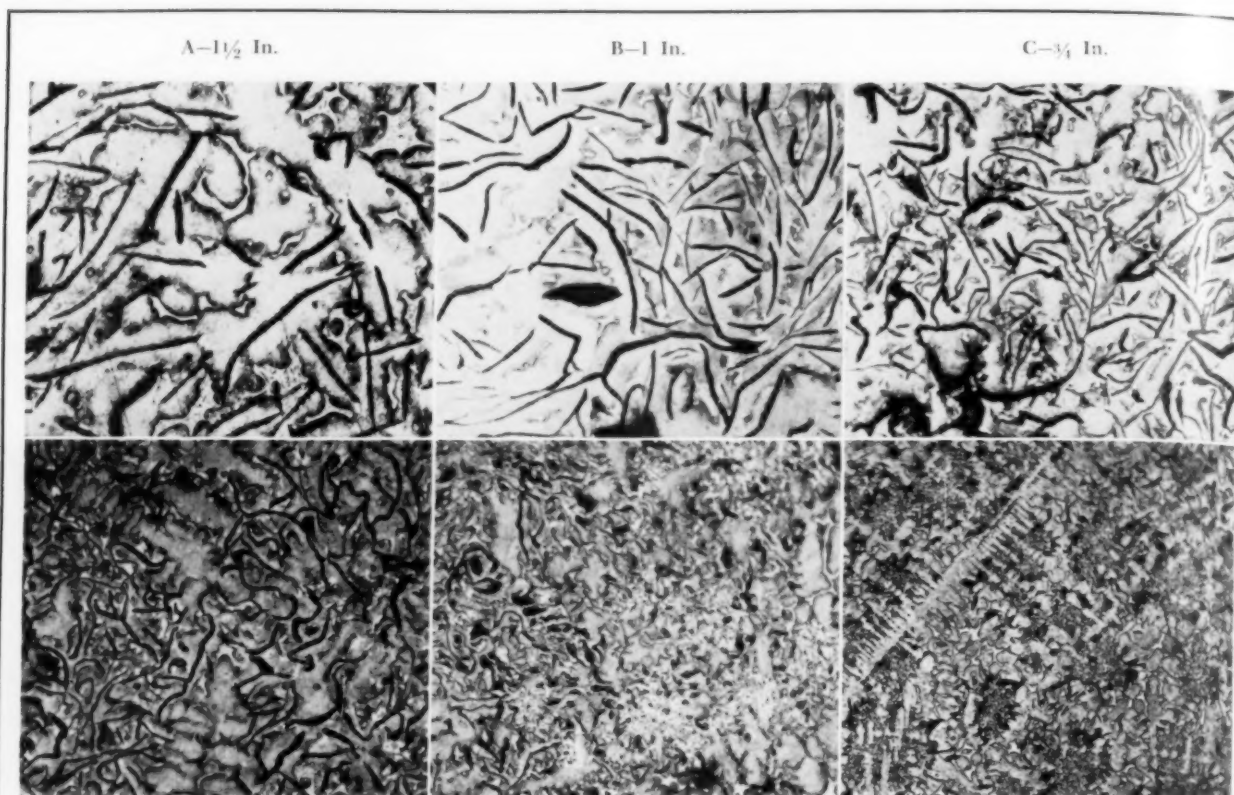


Fig. 18—Type 4, chromium-molybdenum alloy iron, Heat No. 313. Mag. 100X, picral etch.



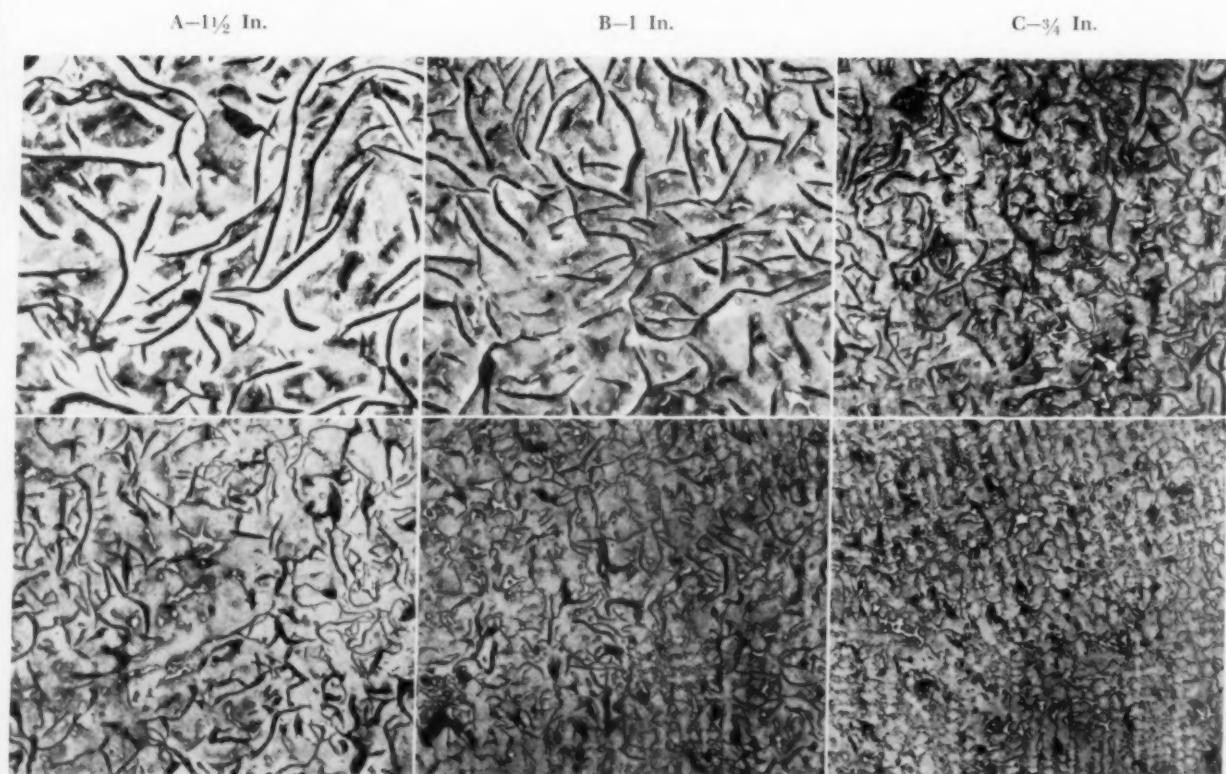


D-1/2 In.

E-3/8 In.

F-1/4 In.

Fig. 19—Type 5, molybdenum alloy iron, Heat No. 286. Mag. 100X, picral etch.



D-1/2 In.

E-3/8 In.

F-1/4 In.

Fig. 20—Type 5, molybdenum alloy iron, Heat No. 278. Mag. 100X, picral etch.

3. Dilatometer studies lead to the conclusion that the resistance of this material to growth is as much as a function of the silicon content as of the alloy content. A comparison of the permanent increase in length after heating from room temperature to 1600 F at a rate of 400 F per hr, holding at 1600 F for two

hours, and cooling at the rate of 400 F per hr for type 4 alloy and type 1 unalloyed cast iron is given in Table 4-G.

The average growth of the type 4 material was only 53 per cent of that of the unalloyed material.

4. Wear was greater on the type 4 material than

TABLE 5-A—CHEMICAL COMPOSITION OF TYPE 5—MOLYBDENUM ALLOY CAST IRON

Heat No.	Cross Section, in. sq.	T.C.	G.C.	C.C.	Si	S	P	Mn	Mo
286A	1½	3.55	3.11	0.44				0.57	0.21
B	1	3.53	2.99	0.54				0.55	0.17
C	¾	3.57	2.93	0.64				0.56	0.19
D	½	3.58	2.90	0.68	2.52	0.071	0.40	0.54	0.17
E	⅜	3.57	2.87	0.70				0.56	0.17
F	¼	3.57	2.84	0.73				0.55	0.18
287A	1½	3.41	2.90	0.51				0.58	0.21
B	1	3.48	2.85	0.63				0.55	0.21
C	¾	3.46	2.79	0.67	2.02	0.047	0.36	0.57	0.19
D	½	3.50	2.74	0.76				0.54	0.18
E	⅜	3.53	2.77	0.76				...	...
F	¼	3.49	2.72	0.77				0.56	0.18

TABLE 5-B—MECHANICAL PROPERTIES OF TYPE 5—MOLYBDENUM ALLOY CAST IRON

Heat No.	Cross Section, in. sq.	Average Tensile Strength, psi	Hardness Rb	Elastic Modulus	Impact Strength, in.-lb.	Spec. Wear	Wear Test, loss, mg.	
							Drum Wear	Total Wear
286A	1½	18,700	74	4.4x10 <sup>6</sup>	3.2	54.0	26.3	80.3
B	1	18,300	76	6.2x10 <sup>6</sup>	2.5	64.2	29.6	93.8
C	¾	26,900	89	10.2x10 <sup>6</sup>	3.0	57.1	29.3	86.4
D	½	30,900	96	10.4x10 <sup>6</sup>	3.0	54.5	31.1	85.6
E	⅜	47,600	99	10.8x10 <sup>6</sup>	3.0	70.0	25.4	95.1
F	¼	48,500	99	13.1x10 <sup>6</sup>	3.0	123.2	26.1	149.3
287A	1½	25,000	79	10.3x10 <sup>6</sup>	4.0	38.9	32.2	71.1
B	1	31,500	88	10.4x10 <sup>6</sup>	3.7	35.5	38.7	74.2
C	¾	31,500	94	11.8x10 <sup>6</sup>	3.7	39.9	35.7	75.6
D	½	41,000	100	13.5x10 <sup>6</sup>	3.0	41.2	35.4	76.6
E	⅜	42,100	101	13.7x10 <sup>6</sup>	3.3	51.5	30.9	82.4
F	¼	54,100	104	15.0x10 <sup>6</sup>	3.3	53.6	29.1	82.7

TABLE 5-C—MICROSTRUCTURES AND THERMAL PROPERTIES OF TYPE 5—MOLYBDENUM ALLOY CAST IRON

Heat No.	Cross Section, in. sq.	Graphite AFA-ASTM	Matrix	Remarks	Coefficient of Expansion, per °F.	Lower Critical, °F	Expansion in 3 In., in.
286A	1½	2-3-A	Ferrite—50% Pearlite—50%				
B	1	3-4-A	Pearlite—85% Ferrite—15%				
C	¾	4-5-A	Pearlite—95% Ferrite—5%				
D	½	5-AB	Pearlite—98% Ferrite—2%	Eutectiform ferrite	6.30x10 <sup>-6</sup>	1500	0.0145
E	⅜	6-8-ABD	Pearlite—95% Ferrite—5%	Eutectiform ferrite			
F	¼	7-ABD	Pearlite—95% Ferrite—5%				
287A	1½	3-A	Pearlite—80% Ferrite—20%				
B	1	4-A	Pearlite—90% Ferrite—10%				
C	¾	4-5-A	Pearlite—98% Ferrite—2%		6.35x10 <sup>-6</sup>	1480	0.0112
D	½	5-A	Pearlite—100%				
E	⅜	6-AE	Pearlite—98% Ferrite—2%				
F	¼	7-8-ED	Pearlite—97% Ferrite—3%	Steadite in cell form			

on the unalloyed cast iron because of the unsatisfactory microstructures containing large amounts of eutectiform ferrite and Type D graphite.

#### Type 5 Molybdenum Alloy Cast Iron

The type 5 alloy contains approximately 0.20 per cent molybdenum. This was not sufficient to change the pearlitic character of the matrix structure but did harden the material and produce a somewhat higher elastic modulus than was obtained from the unalloyed materials. Tables 5-A, B and C give the chemical composition, mechanical properties and microstructures for this type of iron in two different carbon-silicon bases.

**Microstructures.** The molybdenum tends to harden the matrix and stabilize the pearlitic phase. The graphite type, distribution, and matrix structure according to cross-section is shown in Tables 5-D and 5-E.

This material was very sensitive in that pearlite without the occurrence of primary or secondary ferrite occurred only in cross-section D of Heat No. 278. See Fig. 19. and 20.

#### Observations

1. Both microstructures and wear tests indicate that this material does not show the resistance to wear of a good grade of pearlitic unalloyed iron.
2. Hardness, tensile strength and elastic modulus were higher in this material than in the unalloyed iron. Tensile strength and elastic modulus are compared in Tables 5-F for the two metals.
3. Shock resistance and rupture work factor were approximately the same in this material as in an unalloyed material.

#### Type 6 Chromium Nickel Molybdenum Alloy (Heat Treated)

This material was developed originally to meet the requirements of aircraft piston rings. In a 1/8-in. cross-section casting, a tensile strength minimum of 68,000 psi was maintained on tensile test specimens from each heat of this material. Chemical properties, physical properties as heat treated, and microstructures are shown in Tables 6-A, 6-B and 6-C. See Fig. 21, 22, 23, 24 and 25.

TABLE 5-D—DISTRIBUTION OF GRAPHITE IN VARIOUS SECTIONS OF TYPE 5—MOLYBDENUM ALLOY CAST IRON

Heat No.	Flake Graphite	Whorl Graphite	Type D Graphite
286	A, B, C, D, E, F	E, F	F
287	A, B, C, D, E	—	E, F

**Microstructures.** Microstructural studies were made on the material in both the as-cast and fully heat treated conditions. Because of the large amounts of graphite present, the matrices of the larger sections contain considerable ferrite in the as-cast conditions. Heat treatment redissolved carbon from the carbides and from the graphite, and the material quenches to produce a martensitic structure with a small amount of retained austenite. Drawing at 900 F or more converts the retained austenite to an acicular material which breaks down with the martensite to a fine spheroidal material. Any degree of hardness from Rockwell "C" 60 to Rockwell "C" 15 may be produced in this material by varying the draw temperatures from 750 F to 1250 F.

#### Foundry Practice

For sections less than 1/4 in. square, special cupola charges are ordinarily made. These are 20 to 25 per cent steel, 25 per cent pig iron and 50 to 55 per cent return type 6 scrap. Alloy additions are made to the ladle to produce the required analysis. Inoculation is made by the addition of 12 to 16 oz per 100 lb of nickel—70 per cent silicon—30 per cent alloy with a small addition of 85 per cent ferrosilicon. The inoculation should be such that the chill is just removed from the section being cast. This means that a small inoculation is added to the heavy sections and the greater inoculation to the smaller sections and insures production of the tightest structure without massive carbides. Inoculation has not been found necessary in sections above 3/8 in. square.

#### Heat Treatment

Objectives of heat treatment are as follows:

1. To produce a uniform matrix free of ferrite.
2. To eliminate or reduce the size of any massive particles.
3. To control hardness.
4. To increase physical properties such as tensile strength, yield strength, and ultimate strength ratio.

TABLE 5-E—DISTRIBUTION OF MATRIX CONSTITUENTS IN VARIOUS SECTIONS OF TYPE 5—MOLYBDENUM ALLOY CAST IRON

Heat No.	Primary Ferrite + Pearlite	Pearlite Only	Pearlite + Secondary Ferrite
286	A, B, C	—	D, E, F
287	A, B, C	D	E, F

TABLE 5-F—COMPARISON OF TENSILE STRENGTH AND ELASTIC MODULUS IN TYPE 5—MOLYBDENUM ALLOY CAST IRON AND TYPE 1 UNALLOYED CAST IRON

Heat No.	Cross Section, in. sq.	Tensile Strength		Elastic Modulus	
		A Type 5 Material	B Type 1 Material	C Type 5 Material	D Type 1 Material
286	3/8	47,600	35,000	10.8x10 <sup>6</sup>	13.2x10 <sup>6</sup>
287	1/2	41,000	32,100	13.5x10 <sup>6</sup>	8.3x10 <sup>6</sup>



TABLE 6-A—CHEMICAL COMPOSITION OF TYPE 6 (HEAT TREATED)—CHROMIUM-NICKEL-MOLYBDENUM ALLOY

Heat No.	Cross Section, in sq.	T.C.	G.C.	C.C.	Si	S	P	Mn	Cr	Mo	Ni	Quench
343A	1½	3.55	2.68	0.87				0.61	0.25	0.92	0.93	Oil
B	1	3.65	2.75	0.90				0.60	0.23	0.95	0.88	Oil
C	¾	3.69	2.95	0.74	3.25	0.067	0.34	0.60	0.24	0.98	0.98	Oil
D	½	3.70	2.93	0.77				0.61	0.25	0.94	0.92	Air
E	⅜	3.67	2.95	0.72				0.62	0.24	0.95	0.95	Air
F	¼	3.62	2.92	0.70				0.60	0.24	0.90	0.93	Air
401A	1½	3.44	2.91	0.53				0.55	0.23	1.24	0.99	Oil
B	1	3.54	2.85	0.69				0.52	0.22	1.22	1.09	Oil
C	¾	3.60	2.85	0.75	2.82	0.068	0.38	0.52	0.23	1.20	1.05	Oil
D	½	3.63	2.80	0.83				0.49	0.20	1.16	1.15	Air
E	⅜	3.62	2.77	0.86				0.49	0.21	1.18	1.05	Air
F	¼	3.62	2.77	0.85				0.48	0.23	1.14	1.10	Air
346A	1½	3.44	2.40	1.04				0.57	0.31	1.26	1.23	Oil
B	1	3.54	2.55	0.99				0.56	0.31	1.30	1.30	Oil
C	¾	3.53	2.60	0.87				0.58	0.30	1.24	1.27	Oil
D	½	3.56	2.68	0.88	2.14	0.074	0.32	0.55	0.30	1.28	1.33	Air
E	⅜	3.55	2.70	0.85				0.58	0.30	1.18	1.26	Air
F	¼	3.52	2.75	0.77				0.58	0.31	1.20	1.30	Air
345A	1½	3.41	2.80	0.61				0.60	0.28	1.41	1.09	Oil
B	1	3.52	2.61	0.91				0.58	0.28	1.44	1.10	Oil
C	¾	3.54	2.68	0.86				0.59	0.28	1.37	1.20	Oil
D	½	3.59	2.73	0.86	2.03	0.074	0.29	0.56	0.25	1.34	1.15	Air
E	⅜	3.60	2.70	0.90				0.57	0.27	1.30	1.19	Air
F	¼	3.62	2.67	0.95				0.55	0.27	1.33	1.12	Air

TABLE 6-B—MECHANICAL PROPERTIES OF TYPE 6 (HEAT TREATED)—CHROMIUM-NICKEL-MOLYBDENUM ALLOY

Heat No.	Cross Section, in sq.	Average Tensile Strength, psi	Hardness Rb	Elastic Modulus	Impact Strength, in.-lb.	Spec. Wear	Wear Tests, loss, mg.	
							Drum Wear	Total
343A	1½	21,000	71	9.3x10 <sup>6</sup>	1.0	13.9	57.9	71.8
B	1	21,600	97	9.0x10 <sup>6</sup>	1.2	15.9	52.9	68.8
C	¾	30,200	100	10.3x10 <sup>6</sup>	1.5	16.1	63.0	79.1
D	½	35,300	102	10.7x10 <sup>6</sup>	2.0	24.6	45.2	69.6
E	⅜	48,700	103	10.2x10 <sup>6</sup>	2.5	25.5	40.6	66.1
F	¼	57,200	26C	11.0x10 <sup>6</sup>	2.0	36.7	44.2	80.9
401A	1½	22,300	25C	10.9x10 <sup>6</sup>	3.5	15.3	69.1	84.4
B	1	20,250	25C	10.6x10 <sup>6</sup>	1.2	17.8	52.2	70.0
C	¾	26,800	23C	10.3x10 <sup>6</sup>	1.5	11.8	104.8	116.3
D	½	35,400	27C	11.6x10 <sup>6</sup>	2.0	21.1	54.4	75.5
E	⅜	54,800	27C	13.4x10 <sup>6</sup>	2.5	27.9	63.8	91.7
F	¼	70,200	31C	15.0x10 <sup>6</sup>	3.0	34.1	51.2	85.3
346A	1½	41,500	102	11.0x10 <sup>6</sup>	2.0	14.3	50.6	64.9
B	1	43,400	103	15.3x10 <sup>6</sup>	2.0	20.8	49.4	70.2
C	¾	47,400	104	14.5x10 <sup>6</sup>	3.0	19.8	61.5	81.3
D	½	49,900	106	15.6x10 <sup>6</sup>	3.3	27.9	62.2	89.1
E	⅜	62,800	29C	15.6x10 <sup>6</sup>	3.0	27.6	74.9	102.5
F	¼	68,400	28C	15.8x10 <sup>6</sup>	3.5	...	...	...
345A	1½	38,200	102	11.6x10 <sup>6</sup>	1.5	27.4	55.6	84.0
B	1	40,600	104	12.1x10 <sup>6</sup>	1.8	19.7	60.6	80.3
C	¾	42,700	106	14.1x10 <sup>6</sup>	2.0	24.2	48.7	72.9
D	½	56,500	106	15.7x10 <sup>6</sup>	2.5	30.5	47.7	78.2
E	⅜	49,300	106	15.8x10 <sup>6</sup>	2.5	34.3	42.5	76.8
F	¼	54,900	106	16.5x10 <sup>6</sup>	3.0	...	...	...

Heat treatment is accomplished in the following steps:

1. Castings are charged into a controlled atmosphere furnace operating at about 1550 F.
2. They are heated from 1550 F to 1750 F in from 30 min. to 1 hr.
3. The temperature is maintained in 1750 F for 2 hr. or longer if carbide breakdown is necessary.

4. Quenching is accomplished either by an air blast or by oil quenching. Sections ½ in. and under may be air quenched while those above ½ in. must be oil quenched to produce a uniform martensitic matrix.

5. The final step is drawing from 900 to 1150 F depending on the desired hardness.

Several precautions are necessary in the heat treating of cast iron:

on the unalloyed cast iron because of the unsatisfactory microstructures containing large amounts of eutectiform ferrite and Type D graphite.

#### Type 5 Molybdenum Alloy Cast Iron

The type 5 alloy contains approximately 0.20 per cent molybdenum. This was not sufficient to change the pearlitic character of the matrix structure but did harden the material and produce a somewhat higher elastic modulus than was obtained from the unalloyed materials. Tables 5-A, B and C give the chemical composition, mechanical properties and microstructures for this type of iron in two different carbon-silicon bases.

**Microstructures.** The molybdenum stabilizes the matrix and stabilizes the pearlite and graphite type, distribution, and morphology according to cross-section is shown in Table 5-D.

This material was very sensitive to heat treatment without the occurrence of primary ferrite. Primary ferrite occurred only in cross-section D of Type 5, Fig. 19, and 20.

#### Observations

1. Both microstructures and mechanical properties show that this material does not show the characteristics of a good grade of pearlitic unalloyed cast iron.
2. Hardness, tensile strength and elastic modulus were higher in this material than in unalloyed cast iron. Tensile strength and elastic modulus are compared in Tables 5-F for the two materials.
3. Shock resistance and rupture energy were approximately the same in this material as in unalloyed material.

#### Type 6 Chromium Nickel Molybdenum Alloy Cast Iron (Heat Treated)

This material was developed originally to meet the requirements of aircraft piston rings. In section casting, a tensile strength of 40,000 psi was maintained on tensile test specimens of each heat of this material. Chemical composition and mechanical properties as heat treated, and microstructures are shown in Tables 6-A, 6-B and 6-C. See Fig. 21, 22, 23, 24 and 25.

TABLE 5-D—DISTRIBUTION OF GRAPHITE IN VARIOUS SECTIONS OF TYPE 5—MOLYBDENUM ALLOY CAST IRON

Heat No.	Flake Graphite	Whorl Graphite	Type D Graphite
286	A, B, C, D, E, F	E, F	F
287	A, B, C, D, E	—	E, F

**Microstructures.** Microstructural studies were made on the material in both the as-cast and fully heat treated conditions. Because of the large amounts of graphite present, the matrices of the larger sections contain considerable ferrite in the as-cast conditions. Heat treatment redissolved carbon from the carbides and from the graphite, and the material quenches to produce a martensitic structure with a small amount of retained austenite. Drawing at 900 F or more converts the retained austenite to an acicular material which breaks down with the martensite to a fine spheroidal material. Any degree of hardness from Rockwell "C" 60 to Rockwell "C" 15 may be obtained.

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strength, yield strength, and ultimate strength ratio.

TABLE 5-E—DISTRIBUTION OF MATRIX CONSTITUENTS IN VARIOUS SECTIONS OF TYPE 5—MOLYBDENUM ALLOY CAST IRON

Heat No.	Primary Ferrite + Pearlite	Pearlite Only	Pearlite + Secondary Ferrite
286	A, B, C	—	D, E, F
287	A, B, C	D	E, F

TABLE 5-F—COMPARISON OF TENSILE STRENGTH AND ELASTIC MODULUS IN TYPE 5—MOLYBDENUM ALLOY CAST IRON AND TYPE 1 UNALLOYED CAST IRON

Heat No.	Cross Section, in. sq.	Tensile Strength		Elastic Modulus	
		A Type 5 Material	B Type 1 Material	C Type 5 Material	D Type 1 Material
286	$\frac{3}{8}$	47,600	35,000	$10.8 \times 10^6$	$13.2 \times 10^6$
287	$\frac{1}{2}$	41,000	32,100	$13.5 \times 10^6$	$8.3 \times 10^6$

TABLE 6-A—CHEMICAL COMPOSITION OF TYPE 6 (HEAT TREATED)—CHROMIUM-NICKEL-MOLYBDENUM ALLOY

Heat No.	Cross Section, in sq.	T.C.	G.C.	C.C.	Si	S	P	Mn	Cr	Mo	Ni	Quench
343A	1½	3.55	2.68	0.87				0.61	0.25	0.92	0.93	Oil
B	1	3.65	2.75	0.90				0.60	0.23	0.95	0.88	Oil
C	¾	3.69	2.95	0.74	3.25	0.067	0.34	0.60	0.24	0.98	0.98	Oil
D	½	3.70	2.93	0.77				0.61	0.25	0.94	0.92	Air
E	⅜	3.67	2.95	0.72				0.62	0.24	0.95	0.95	Air
F	¼	3.62	2.92	0.70				0.60	0.24	0.90	0.93	Air
401A	1½	3.44	2.91	0.53				0.55	0.23	1.24	0.99	Oil
B	1	3.54	2.85	0.69				0.52	0.22	1.22	1.09	Oil
C	¾	3.60	2.85	0.75	2.82	0.068	0.38	0.52	0.23	1.20	1.05	Oil
D	½	3.63	2.80	0.83				0.49	0.20	1.16	1.15	Air
								0.49	0.21	1.18	1.05	Air
								0.48	0.23	1.14	1.10	Air
								0.57	0.31	1.26	1.23	Oil
								0.56	0.31	1.30	1.30	Oil
								0.58	0.30	1.24	1.27	Oil
								0.55	0.30	1.28	1.33	Air
								0.58	0.30	1.18	1.26	Air
								0.58	0.31	1.20	1.30	Air
								0.60	0.28	1.41	1.09	Oil
								0.58	0.28	1.44	1.10	Oil
								0.59	0.28	1.37	1.20	Oil
								0.56	0.25	1.34	1.15	Air
								0.57	0.27	1.30	1.19	Air
								0.55	0.27	1.33	1.12	Air

D)—CHROMIUM-NICKEL-MOLYBDENUM ALLOY

Impact length, in.-lb.	Spec. Wear	Wear Tests, loss, mg.	
		Drum Wear	Total
1.0	13.9	57.9	71.8
1.2	15.9	52.9	68.8
1.5	16.1	63.0	79.1
2.0	24.6	45.2	69.6
2.5	25.5	40.6	66.1
2.0	36.7	44.2	80.9
3.5	15.3	69.1	84.4
1.2	17.8	52.2	70.0
1.5	11.8	104.8	116.3
2.0	21.1	54.4	75.5
2.5	27.9	63.8	91.7
3.0	34.1	51.2	85.3
2.0	14.3	50.6	64.9
2.0	20.8	49.4	70.2
3.0	19.8	61.5	81.3
3.3	27.9	62.2	89.1
3.0	27.6	74.9	102.5
3.5	...	...	...
1.5	27.4	55.6	84.0
1.8	19.7	60.6	80.3
2.0	24.2	48.7	72.9
2.5	30.5	47.7	78.2
2.5	34.3	42.5	76.8
3.0	...	...	...

Heat treatment is accomplished in the following steps:

1. Castings are charged into a controlled atmosphere furnace operating at about 1550 F.
2. They are heated from 1550 F to 1750 F in from 30 min. to 1 hr.
3. The temperature is maintained in 1750 F for 2 hr. or longer if carbide breakdown is necessary.

4. Quenching is accomplished either by an air blast or by oil quenching. Sections ½ in. and under may be air quenched while those above ½ in. must be oil quenched to produce a uniform martensitic matrix.
5. The final step is drawing from 900 to 1150 F depending on the desired hardness.

Several precautions are necessary in the heat treating of cast iron:



TABLE 6-C—MICROSTRUCTURES AND THERMAL PROPERTIES OF TYPE 6—CHROMIUM-NICKEL-MOLYBDENUM ALLOY

Heat No.	Cross Section, in. sq.	Graphite AFA-ASTM	Microstructure		Heat Treatment	Thermal Analysis		
			Matrix	Remarks		Coefficient of Expansion, in. per °F.	Critical (Heating) ° F.	Expansion in 3 in., in.
343A	1½	2-3-A	Ferrite—75% Pearlite—25%	Secondary ferrite	As Cast	6.26x10 <sup>-6</sup>	1420	0.018
B	1	3-A	Ferrite—50% Pearlite—50%	Secondary ferrite				
C	¾	3-4-A	Ferrite—10% Pearlite—90%	Secondary ferrite				
D	½	4-5-A	Ferrite—5% Acicular—90% Pearlite—5%	Secondary ferrite				
E	⅜	5-6-AB	Acicular—100%	—				
F	¼	6-AB	Acicular—100%	—				
343A	1½	2-3-A	Drawn Martensite—100%	—	1550°-1750° F, in 30 min., 1750° F 2 hr., Quench, Draw at 1070° F	6.26x10 <sup>-6</sup>	1420	0.018
B	1	3-A	Drawn Martensite—100%	—				
C	¾	3-4-A	Drawn Martensite—100%	—				
D	½	4-5-A	Drawn Martensite—100%	—				
E	⅜	5-6-AB	Drawn Martensite—100%	—				
F	¼	6-AB	Drawn Martensite—100%	—				
401A	1½	3-A	Pearlite—50% Ferrite—50%	Secondary ferrite	As Cast	6.35x10 <sup>-6</sup>	1420	0.0094
B	1	3-5-A	Ferrite—30% Pearlite—70%	Secondary ferrite				
C	¾	4-A	Ferrite—5% Acicular—80% Pearlite—15%	Secondary ferrite Secondary ferrite Secondary ferrite				
D	½	5-6-A	Acicular—100%	—				
E	⅜	5-6-A	Acicular—100%	—				
F	¼	7-8-ABD	Acicular—100%	—				
346A	1½	2-3-A	Ferrite—20% Pearlite—80%	Secondary ferrite	As-Cast	6.35x10 <sup>-6</sup>	1420	0.0094
B	1	4-5-A	Pearlite—40% Acicular—60%	—				
C	¾	4-5-A	Pearlite—5% Acicular—95%	—				
D	½	5-6-A	Acicular—100%	—				
E	⅜	5-7-AB	Acicular—100%	—				
F	¼	7-8-AB	Acicular—100%	—				
346A	1½	2-3-A	Drawn Martensite—100%	—	1550° F. 1750° in 30 min. 1750° F, 2 hr., Quench Draw at 1070° F	6.35x10 <sup>-6</sup>	1420	0.0094
B	1	4-5-A	—	—				
C	¾	4-5-A	—	—				
D	½	5-6-A	—	—				
E	⅜	5-7-AB	—	—				
F	¼	7-8-AB	—	—				
345A	1½	2-3-A	Ferrite—10% Pearlite—90%	Considerable carbide in small islands Cellular phosphide	As-Cast	6.18x10 <sup>-6</sup>	1410	0.010
B	1	4-A	Pearlite—40% Acicular—60%					
C	¾	4-5-A	Pearlite—5% Acicular—95%					
D	½	5-6-A	Acicular—100%					
E	⅜	5-7-ABD	Acicular—100%					
F	¼	8-ABD	Acicular—100%					
345A	1½	2-3-A	Drawn Martensite	Carbide islands	1550° F. 1750° in 30 min., 1750° F, 2 hr., Quench, Draw at 1070° F	6.18x10 <sup>-6</sup>	1410	0.010
B	1	4-A	—	Carbide islands				
C	¾	4-5-A	—	Carbide islands				
D	½	5-6-A	—	—				
E	⅜	5-7-ABD	—	—				
F	¼	8-ABD	—	—				

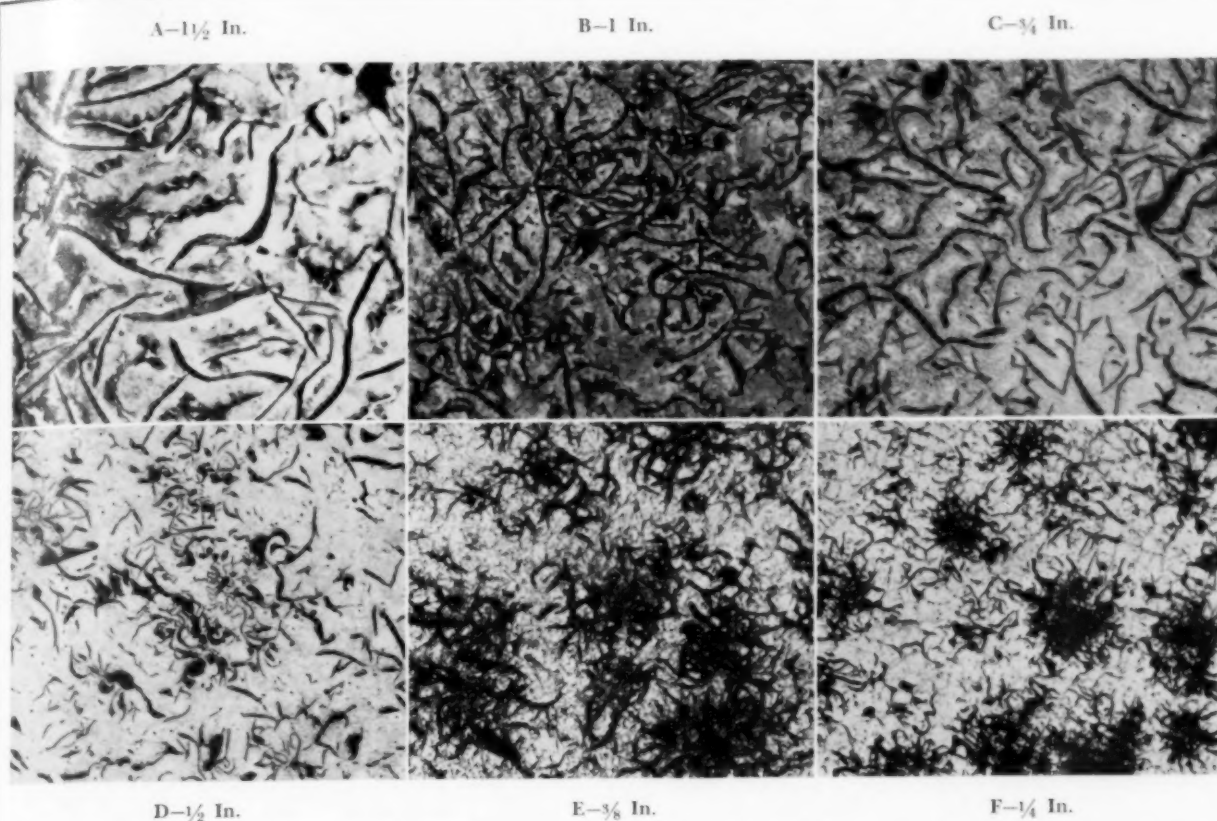


Fig. 21—Type 6, nickel-chromium-molybdenum alloy iron, Heat No. 345, as cast. Mag. 100X, picral etch.

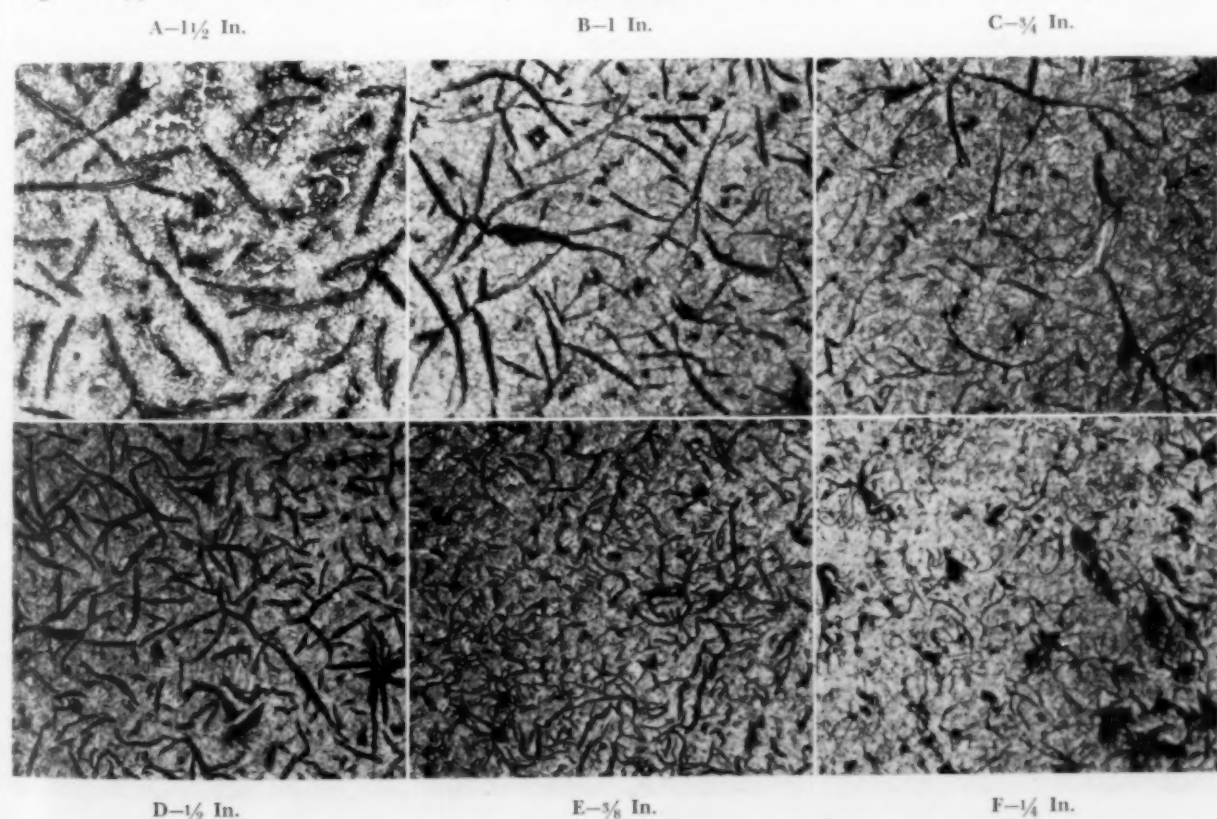


Fig. 22—Type 6, chromium-nickel-molybdenum alloy iron, Heat No. 343, heat treated. Mag. 100X, picral etch.

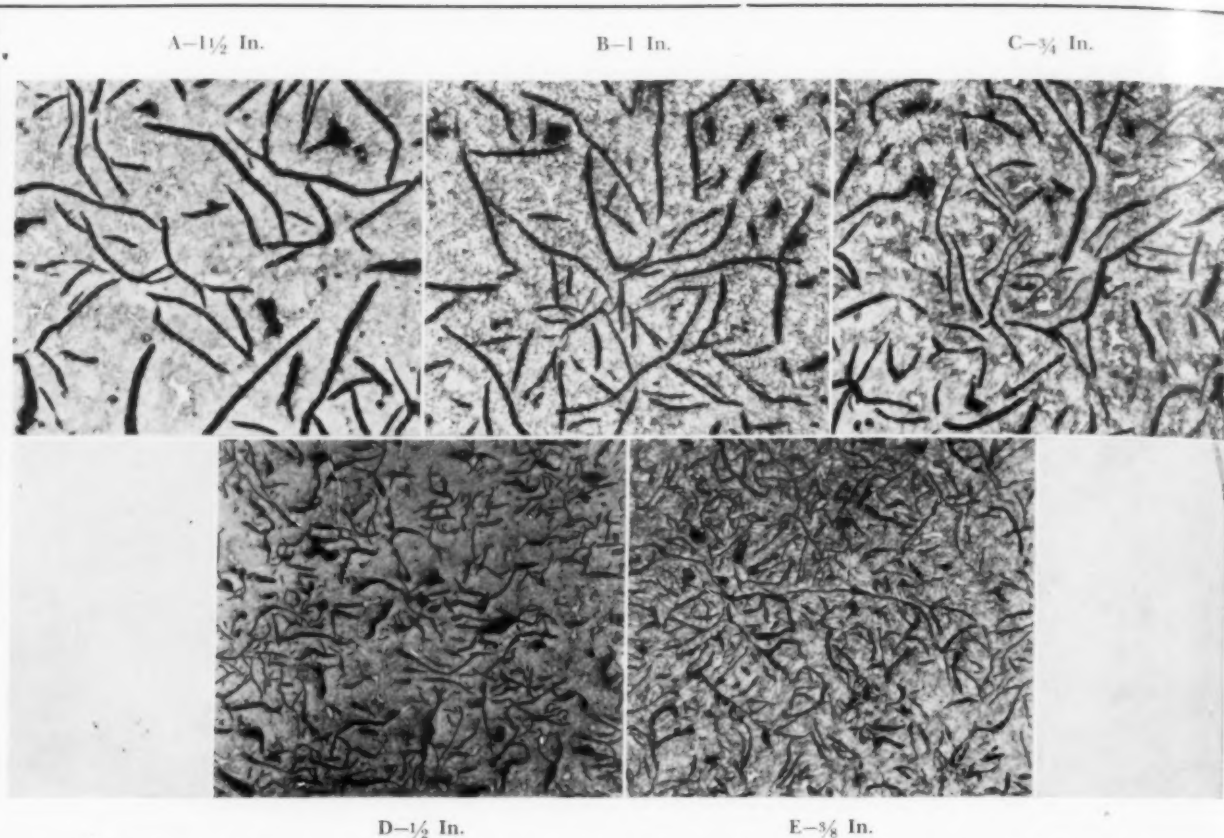


Fig. 23—Type 6, nickel-chromium-molybdenum alloy iron, Heat No. 401, heat treated. Mag. 100X, picral etch.

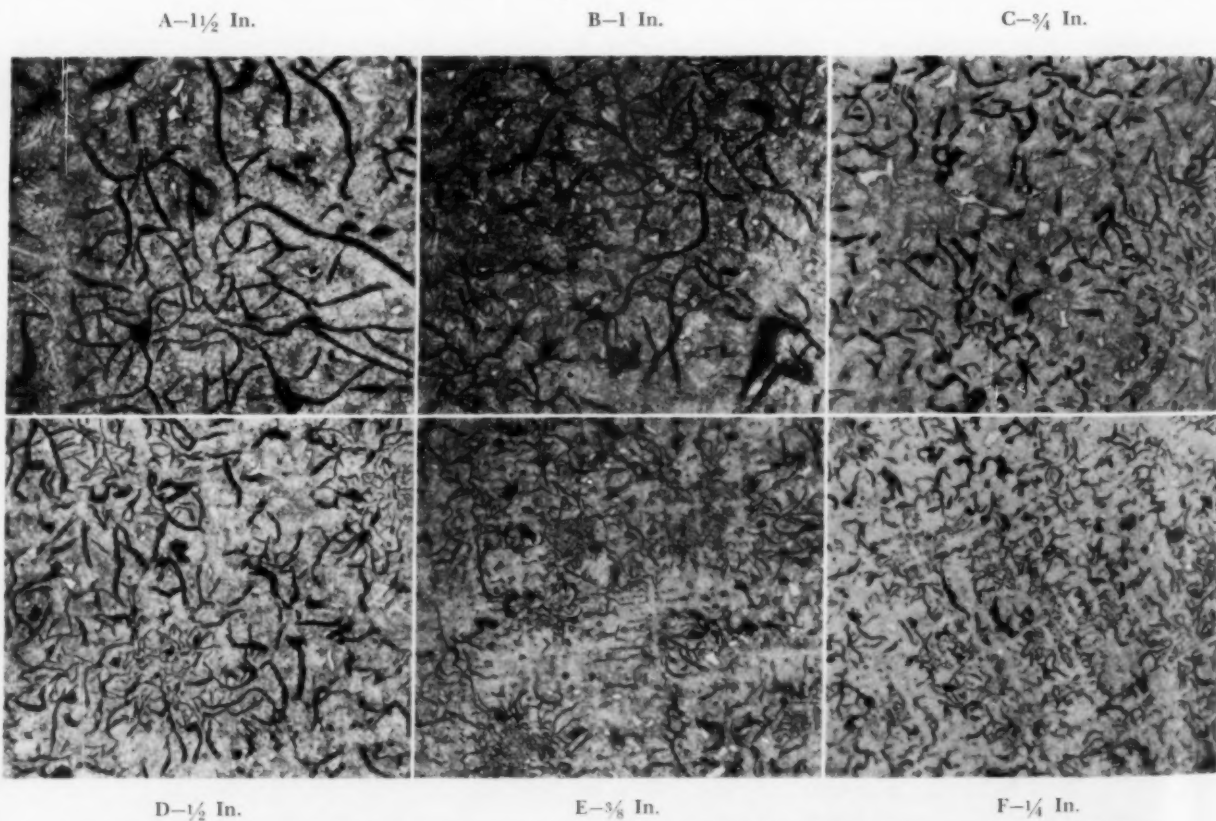


Fig. 24—Type 6, nickel-chromium-molybdenum alloy iron, Heat No. 345, as cast. Mag. 100X, picral etch.

TAB

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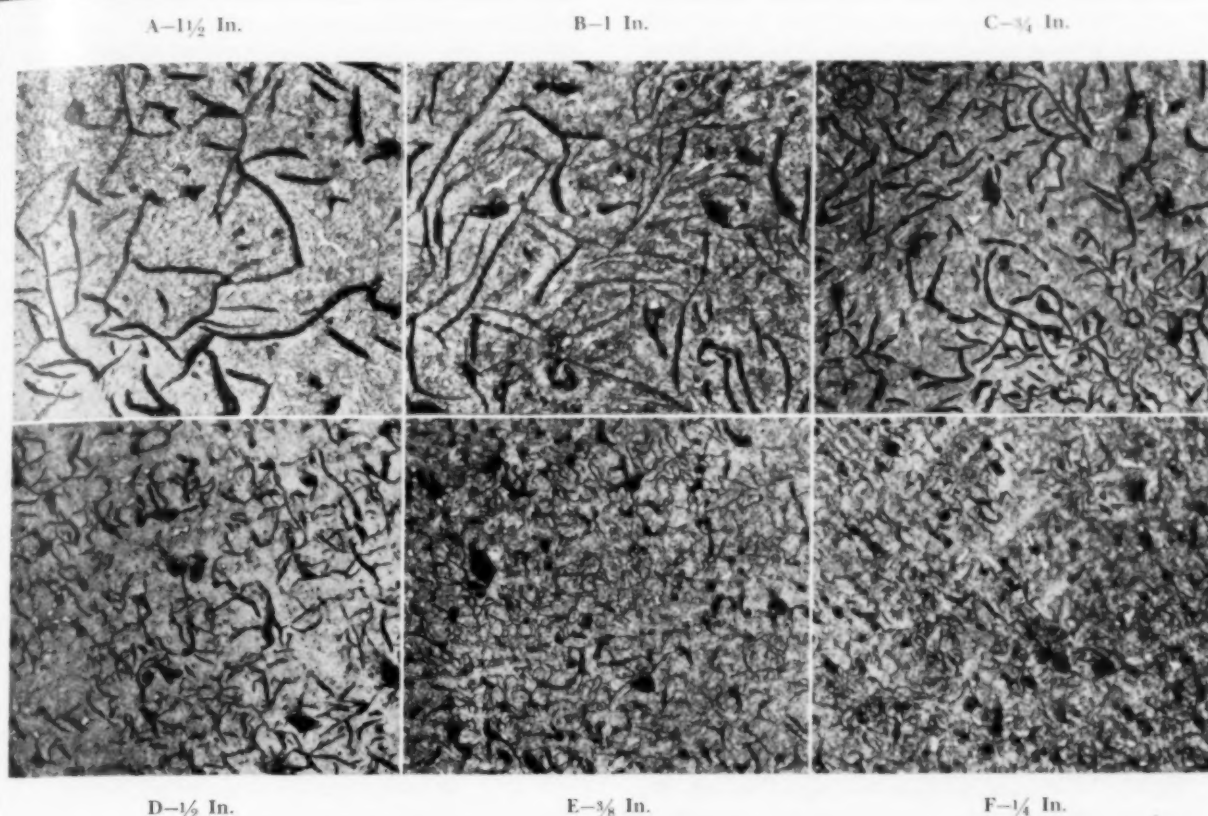


Fig. 25—Type 6, nickel-chromium-molybdenum alloy iron, Heat No. 346, heat treated. Mag. 100X, picral etch.

TABLE 6-D—COMPARISON OF TENSILE STRENGTH AND ALLOY AND TYPE I—UNALLOYED CAST IRON

Heat No.	Cross Section, in sq.	Tensile Strength, psi		Specimen Wear, loss, mg.	
		Type 6 A	Type 1 B	Type 6 C	Type 1 D
343	1/4	70,200	37,200	34.1	52.8
401	3/8	62,800	35,000	27.6	53.8
346	1/2	56,500	32,100	24.2	45.6
345	1	40,600	33,200	19.7	41.8

(1) To prevent surface decarburization, a gas atmosphere must be used which would be carburizing to steel. "Neutral" atmospheres do not protect cast irons;

(2) Cast iron graphitizes and the carbides tend to migrate to the austenite grain boundaries when cooling takes place at certain critical rates within the austenitic temperature range. It is important that no delay take place in the transfer of the castings from the furnace to the quench.

Dilatometric studies of the heat treatment are shown in Fig. 26. These give the following information:

1. The austenitizing temperature of the material (heating rate 700 F per min.) is 1490 F.
2. Little permanent expansion takes place during the heating and cooling periods.
3. Martensite formation takes place during the quench at below 450 F.

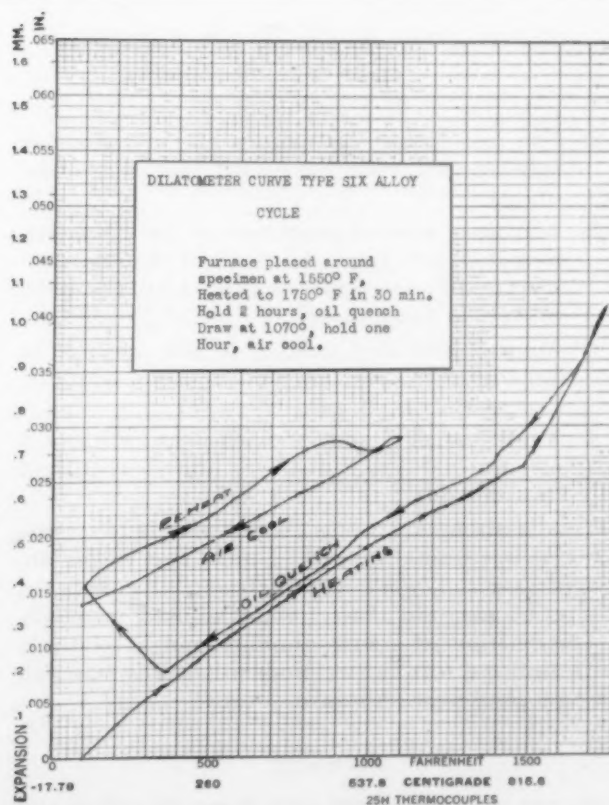


Fig. 26—Heat treatment dilatometer curve for Type 6 alloy.

4. Most of the permanent expansion (growth) of the material takes place during the quench and is due to the replacement of the dense austenitic structure by martensite.
5. Further expansion takes place between 109 F and 900 F during the draw. This is caused by breakdown of the retained austenite in the quenched structure.
6. Coalescence of carbides and the destruction of martensite at between 900 and 1000 F causes a slight contraction.
7. Contraction during cooling from 1070 F is proportional to the decrease in temperature and shows a coefficient of contraction of  $6.02 \times 10^{-6}$  F.
8. Permanent expansion during heat treatment was 0.57 per cent.

Ordinarily heat treatment draw temperatures are adjusted to produce a Rockwell "C" hardness of approximately 33. However, in this case a standard treatment was given causing a considerable variation in hardness. For this reason the mechanical properties in the heavy cross-sections are far below that found on materials of higher hardness.

### Observations

1. The matrix of the heat treated material is uniformly a drawn martensite regardless of the cross-section of the as-cast matrix condition. Therefore, physical properties are largely the result of—
  - a. The degree of hardness of the matrix material.
  - b. Graphite size, amount and distribution.
  - c. Carbide and steadite distribution.
  - d. The analysis balance of the material.
2. The low physical properties of Heat No. 343 indicate that excessively high silicon is undesirable.
3. The poor physical properties of Heat No. 345, sizes D, E and F are the result of the un-inoculated graphite structure and the cellular carbide-phosphide distribution.
4. Alloying and heat treating of this material in large sections does not produce a fine graphite structure.
5. Tensile strength and specimen wear for a type 6 material and type 1 unalloyed iron are compared in Table 6-D.

Tensile strength on the alloyed and heat treated material was from 1.22 to 1.82 times that of the unalloyed cast iron, while wear was reduced to between 47 and 65 per cent that of the unalloyed materials. A combination of coarse graphite and a relatively hard, uniform matrix was produced in this material, giving a good wear resistance.

### Summary

Six cast iron alloy types, each in several carbon and silicon analyses, were examined for the effect of analysis and cross section on the mechanical properties and

microstructures. The change in microstructure produced by variation of the cross section and analysis effected the mechanical properties of the material. Some of these effects are noted as follows:

1. Tensile strength increases with the refinement of the matrix and graphite structure produced by decrease in cross section. Low tensile strength may accompany very fine graphite when it is in the interdendritic form (AFA Type D).
2. Hardness increases with decreasing cross section.
3. Elastic modulus increases somewhat in the smaller cross sections with the refinement of the graphite and matrix structures. Very low elastic moduli occur in materials containing large quantities of primary ferrite. Materials containing AFA Type D graphite generally show a high elastic modulus.
4. Wear is influenced by microstructure and indirectly by casting cross section.
  - a. High rates of wear were observed in specimens containing the degenerate eutectic ferrite. Very small amounts of this structure influenced wear resistance adversely.
  - b. Secondary ferrite influenced wear resistance unfavorably but not to the extent of the degenerate structure.
  - c. Wear resistance is increased in specimens showing a coarse graphite structure, while it decreased in specimens having a fine graphite structure.
  - d. Pearlitic, acicular and drawn martensitic structures of the same hardness and having the same graphite flake size and distribution produced comparable wear resistance.
  - e. With a given graphite structure, increasing matrix hardness produced increasing resistance to wear.
5. The nickel-chromium alloy (Type 2) could be cast in a greater range of cross sections with a uniform pearlitic matrix structure than any of the other alloys tested.
6. Impact resistance of the acicular nickel-molybdenum alloy (Type 3) was two to three times greater than that of corresponding pearlitic materials.
7. Molybdenum and molybdenum-chromium additions (alloy Types 4 and 5) decreased the cross-sectional interval in which uniform pearlitic matrices could be produced. These alloys displayed the degenerate ferrite-interdendritic graphite structures in the smaller cross sections.
8. An increased tensile strength, hardness, elastic modulus and wear resistance was obtained in the heat treated, nickel-molybdenum-chromium alloy (Type 6).

### DISCUSSION

Chairman: A. E. SCHUH, U. S. Pipe & Foundry Corp., Burlington, N. J.

Co-Chairman: F. G. SEFING, International Nickel Co., Inc., New York

J. S. VANICK (*Written Discussion*):<sup>1</sup> Foundrymen are perhaps more conscious of the influence of cooling rate on their products than any other metal producing group, and the comprehensive

<sup>1</sup> International Nickel Co., Inc., New York

study which the author presents, provides a long-lasting reservoir of wholesome information. Any discussion must be confined to a selection of only one or two items. It is particularly important to know that an iron testing 20,000 psi in the standard test bar sections, actually possesses 40,000 psi tensile strength in piston ring sections. The differences in rates of wear or in modulus or strength might be separately considered by some, but a piston ring casting requires consideration of all of them, and an engineer may be willing to make concessions, for example, in wear, to get better results in strength.

The unalloyed irons of Table I-B offer an excellent opportunity to illustrate this principle. The low modulus and strength of the soft iron No. 250 is accompanied by an excellent resistance to wear in all except the important  $\frac{1}{4}$ -in. section. Specimens of heat 253 wear well in the thicker sections, but wear away rapidly in the thin sections. The microstructures seem to provide the explanation. Unfortunately, thin specimens of the low-silicon heat 383 were unmachinable, but their influence upon wearing rate might be predicted from the comments drawn up above. The generally favorable properties in the  $\frac{3}{4}$ -in. sections illustrate the common plight of the foundryman in reproducing  $\frac{3}{4}$ -in. section properties in  $\frac{1}{4}$ -in. section castings. This is where alloys, inoculants and heat-treating procedures are applied by the skilled foundryman and metallurgist to achieve the results the author reports in his conclusions Nos. 5, 6 and 8.

Similarly, conclusion No. 4d provides a splendid test of an engineer's skill in selecting suitable materials. Pearlitic, martensitic and acicular structures of the same hardness and graphite patterns may have approximately similar wearing rates for the type of wear involved, but it takes conclusion No. 6 to emphasize the point that the toughness of these three microstructures is remarkably different. Incidentally, some work at the International Nickel Co. laboratories indicates that the acicular irons wear better. These and many other useful data are recorded in author Shuck's work, and American foundrymen owe him and his company a vote of appreciation for his work.

CHAIRMAN SCHUH: I should like to congratulate the author for a very fine paper. The Koppers Co. is to be highly commended for the release of this information.

W. S. LOEB:<sup>2</sup> I should like to ask the author how thin a section he can get with the alloy irons. He used in his tests  $\frac{1}{4}$ -in. section test bars. Piston rings are down to  $\frac{3}{32}$  and  $\frac{1}{8}$ -in. cross section.

MR. SHUCK: Cross sections smaller than  $\frac{1}{4}$  in. sq. were not reported because most of the alloys evaluated were unsuitable for  $\frac{1}{4}$  and  $\frac{3}{32}$ -in. cross sections, and because the investigation was for the purpose of evaluating the larger sections. However, analyses No. 250, 341, 310 and 343 might have been poured in  $\frac{1}{8}$ -in. or  $\frac{3}{32}$ -in. sections.

H. O. MOFFETT:<sup>3</sup> I want to compliment the author on presenting an excellent paper. I have two questions to ask of the author. The first deals with alloys. Were the alloys cupola or ladle additions? The second question deals with the modulus

of elasticity of the irons. The paper in general correlates cross section of the rings to the characteristics of the iron. The elastic modulus, I assume, was measured on six sizes for all different heats. As a result there seems to be a small discrepancy in the results based on the modulus. I refer primarily to Table 2G in which the author compares modulus of an unalloyed iron with that of an alloyed iron in the form of A/B. This value ranges from less than 1 to over 1. I was wondering whether or not the method of measuring that characteristic influenced the results obtained.

MR. SHUCK: The methods of determining elastic modulus, in cast iron are somewhat open to question. The values obtained on tensile bars often have no practical use when a cast iron member is loaded transversely. We have had little practical success with the apparent modulus values obtained from large transverse bars nearly the full casting cross section. This may be the result of the disproportionate effect of the closer structured, more resistant surface material. The small transverse bars were thought to overcome this disproportionate surface effect although admittedly the results obtained were not entirely uniform.

In answer to the question on the method of alloy addition, part of the alloy was melted in the cupola as return alloy cast iron scrap, part was added as a ladle addition.

T. J. SNOBCASS:<sup>4</sup> Do you care to comment on the correlation of the modulus as you obtained it with a modulus that you would obtain on a regular tensile test bar? You used two sizes, I believe.

MR. SHUCK: The modulus values obtained on tensile bars were higher than those obtained on transverse bars. The modulus was determined at relatively high stress values, 50 per cent of the breaking load. This produced a value which was usable in the calculation of piston rings which often show a tension equal to 75 per cent of the breaking load.

J. T. MACKENZIE:<sup>4</sup> Do you actually mean that the piston ring is working three times its fatigue limit?

MR. SHUCK: Computation of the surface stress of a gray iron piston ring, closed to nominal diameter in a cylinder, shows stress values at 180 degrees from the gap in excess of 40,000 psi. By loading the same ring to the breaking point, the computed breaking stress will be between 55,000 and 65,000 psi. (Automotive cross section).

In some cases where high tensions must be obtained, stress values may be as great as 65,000 to 75,000 psi. In these cases, ordinary gray iron is not satisfactory.

DR. MACKENZIE: That is in compression though, is it not?

MR. SHUCK: Tension and compression values computed from the bending moment.

CHAIRMAN SCHUH: I think the biggest value of this paper is that it has given us many quantitative relationships over a very wide freezing range, the sort of information we have all been looking for. Here we have it quantitatively. It deserves considerable study.

<sup>2</sup> Wilkening Mfg. Co., Philadelphia

<sup>3</sup> General Electric Co., Trenton, N. J.

<sup>4</sup> American Cast Iron Pipe Co., Birmingham



# STATISTICAL QUALITY CONTROL

## A NEW TOOL FOR THE FOUNDRYMAN

By

H. H. Johnson and G. A. Fisher\*

### ABSTRACT

*The application of statistical methods to the measurement and control of the quality of castings as they are being produced in the foundry is a comparatively recent development and one that has been found to be quite useful. The principles underlying such methods are outlined and examples are given of their application in evaluating process control in a foundry that is producing steel castings.*

### Background of Quality Control

APPLICATION OF STATISTICAL METHODS to the measurement and control of quality of the castings being produced in the foundry is a comparatively recent development. The mathematical principles on which the scheme is based are not new and the principles of their application were set forth by Shewhart<sup>1</sup> about 1925. However, comparatively little attention was paid to the application of these principles and to product quality audits until the advent of World War II.

One of the wartime problems with which the foundries soon began to struggle was that of the production of cast armor plate, with its attendant ballistic test requirements. It soon became evident that these test requirements would exact a large quantity of armor plate that was needed for service and that the testing itself would require a tremendous amount of work. The Ordnance Department thereupon developed a Quality Control Plan whereby they were assured of the quality of the product and at the same time, the amount of testing required was considerably reduced. Following the same principles, a similar plan was set up for reduced tensile testing requirements for other steel castings (under the provisions of "Steel Castings Quality Control Plan," QCP-3, of the Ordnance Department).

Other industries also began to introduce programs for reduced testing, reduced sampling, etc., based on the same principles. Courses of instruction were sponsored by the government to train representatives of industry in the application of these principles. Typical of the related publications were the "American

War Standards" series.<sup>2,3</sup> Largely out of this interest in quality control has grown the new technical society known as the "American Society for Quality Control," the members of which are, for the most part, engaged in applying the statistical methods of quality control to many of the phases of manufacturing processes for a great variety of products.

Because this new tool of industry is meeting with such success in so many cases, where it has been applied, it seemed reasonable to attempt to apply it to some of the steps in our foundry practice. Accordingly, we started several months ago to see what we could develop and this is intended to be a progress report in which we set forth some of the principles on which such a quality control program can be built and outline somewhat the results that we have obtained.

### Principles Underlying Quality Control

Two of the principles which underlie the theory of quality control are *variation* and *measurement*. No two things are exactly the same and although the variations may be small they do exist. Further, it is known that a large number of causes may contribute to the variations in quality from one unit of the product to another. If the cause contributing to the variation can be identified, it is called an "assignable cause" and one of the problems in quality control work is to learn to find the assignable causes so that they can be eliminated or allowance made for them.

The statistical measure of the variations in a process is frequently set forth in the form of a quality control chart by means of which we are able to predict whether the property under consideration will maintain a predetermined average and will also stay within certain limits.

The idea is relatively simple. Fay<sup>4</sup> has likened a control chart to a highway (Fig. 1). The highway must be maintained at a certain width to accommodate the necessities of the vehicles using it. In our case, the width of the chart is determined by the process itself. As long as the driver drives his car down the exact center of the highway he is in perfect control. Since this is usually impossible to do, he may

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vary his course on either side of the central line but as long as he remains on the highway his car is "in control" and his driving is probably safe. Should a wheel of the car get off the pavement and on the "shoulder" (which would put him outside of a control limit) then action should be taken in order to

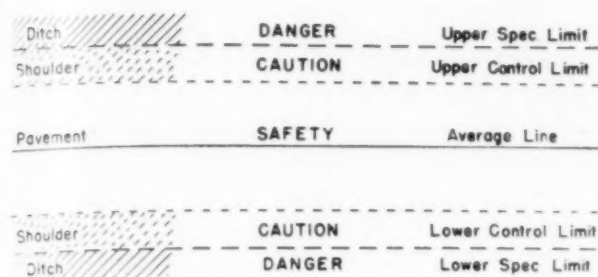


Fig. 1—An analogy of a quality control chart.

bring it back into control or the result may be disastrous. Should the car get entirely out of hand, it will go in the ditch which in our case means that individual parts will go out of specification limits and therefore be rejected. The "width of the shoulder" in our analogy, is the margin between what our process is required to produce.

#### The Control Chart

The control chart is thus a picture of what has happened in the process but it has the added function that it gives us immediate warning when something about the conditions of the process has changed erratically. (Shewhart<sup>1</sup> calls this the presence of an assignable cause of error.)

The mathematical basis for the construction of the control chart lies in the relationships derived from the probability or frequency distribution curve. In Fig. 2 we show a curve typical of the expected distribution of a series of measurements if the process that is being measured is in control. The majority

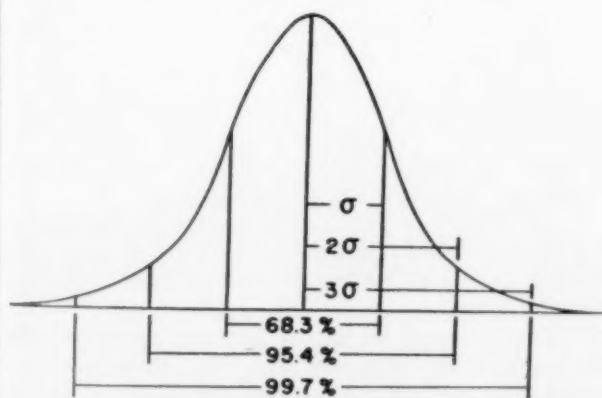


Fig. 2—Normal distribution curve.

of the measurements will fall near the centerline or the desired value, while the number varying from the centerline will fade out toward the high and low limits. The amount of variation, or the dispersion

of results from the average, is expressed as the standard deviation (and denoted by the Greek letter sigma,  $\sigma$ ). It is equal to about  $1/6$  of the range, as shown in Fig. 2. It will be further noted that if a distance equal to  $3\sigma$  is measured off on the X-axis on both sides of the arithmetic mean, 99.7 per cent of the values will be included within the limits indicated.

In shop application, the quality control chart replaces the frequency distribution curve because it is somewhat simple to construct and interpret and also because it is a more effective guide for controlling the process. It may be thought of as another form of the probability curve, in which (for our purposes) the observations are regrouped in lots of four successive observations (Fig. 3). The lots of four observa-

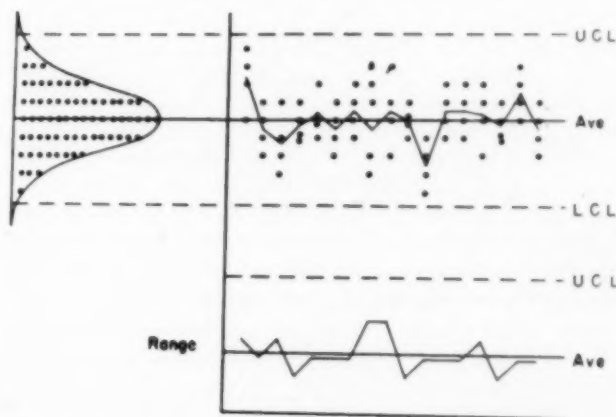


Fig. 3—Relationship between probability curve and quality control chart.

tions were chosen because it is a generally accepted fact that it is more satisfactory to look for causes of error in groups rather than in single determinations. The  $3\sigma$  limits are computed and plotted to indicate the upper and lower control limits for the process, for the conditions under which it is operating.

One other factor which must not be overlooked is the scatter or range of the observations, for each group of four observations, that are plotted on the chart, the spread of values between the greatest and the least is noted and plotted as shown in Fig. 3. For each lot of ten (10) groups, the  $3\sigma$  limits are computed to determine the control limits for this step in the evaluation of the process.

#### Calculation of Control Limits

The calculation of control limits is accomplished by the application of some relatively simple formulae, although the theory behind them is not too elementary. Reference to the American War Standard Z1.3-1942<sup>2</sup> provides the following formulae:

For our calculations, individual values are considered in groups of four and at least ten such groups (or 40 observations) are considered as a unit for calculating control limits.

Usually the groups are considered in multiples of ten groups for such calculations up to 50 groups. As

additional multiples of ten are added, previous groups of ten are dropped.

$\bar{X}$  = Average value for each group of four observations.

$R$  = Range of values in each group of four and is the difference between the smallest and the largest.

$\bar{\bar{X}}$  = Grand average of the averages.

$\bar{R}$  = Average range.

$\bar{X} + A_2 \bar{R}$  = Upper Control Limit (U.C.L.) for the average values.

$\bar{X} - A_2 \bar{R}$  = Lower Control Limit (L.C.L.) and, for the conditions of grouping indicated,

$$A_2 = .729$$

$D_4 \bar{R}$  = Upper Control Limit for the range values, and for the same grouping

$$D_4 = 2.282$$

$D_3 \bar{R}$  = Lower Control Limit for the range values and for our case, since

$$D_3 = 0, \text{ the L.C.L.} = 0.$$

Other sizes of groups (such as 5 values) may be used just as satisfactorily, with values of  $A_2$ ,  $D_3$ , and  $D_4$  corresponding to the size of group chosen.

#### Construction of Average and Range Charts

As an example of the Control Chart technique let us examine some typical production charts. For example, let us consider a chart which shows the carbon and manganese control for Grade "B" steel made in 35-ton basic open hearth furnaces at the authors' plant (Fig. 4). The carbon and manganese

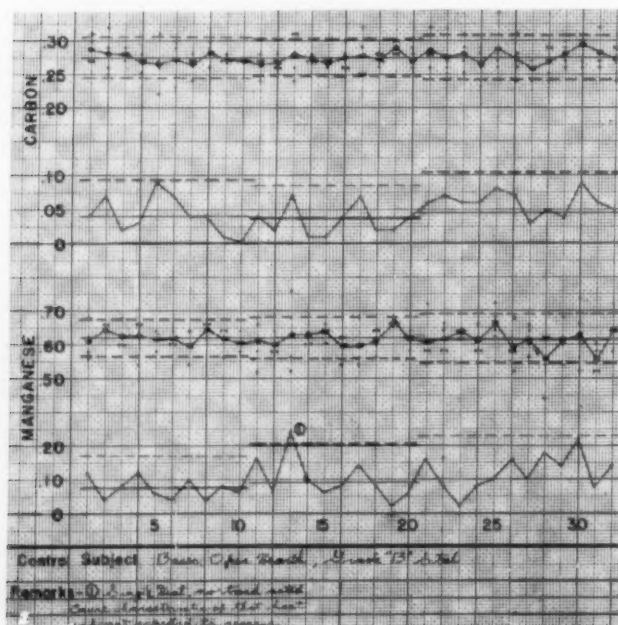


Fig. 4—Carbon and manganese control chart for Grade "B" 35-ton O.H. furnace heats.

value for each heat is indicated as a point on the chart. Values for four consecutive heats are plotted on the same vertical line to form a group. The average for each group is indicated with a circle and the circles are connected with the solid line so that the fluctuation of the average values may be more easily followed. The difference between the greatest and the smallest value for each group is plotted as the range value for the group and these values are also connected to show the fluctuation more readily.

For calculating the control limits it is necessary to calculate the grand average ( $\bar{\bar{X}}$ ) of the average values ( $\bar{X}$ ) and to also calculate the average ( $\bar{R}$ ) of the range values ( $R$ ) for lots of ten groups. Then, applying the formula  $\bar{\bar{X}} + .729 \bar{R}$ , we find the values for the Upper Control Limit and the Lower Control Limit for the percent of carbon and manganese that the process is producing. Similarly  $2.282 \bar{R}$  gives the U.C.L. for the range values.

#### Control of Process

Having calculated and plotted these limits the state of control of the process is readily determined. In the present example, it is noted that, for the conditions under which the process is operating, it is in control so far as carbon and manganese fluctuations are concerned except for the manganese range of the thirteenth group. No average value is out of its control limit on the percentage chart, no other range value is out of its control limit, and the average values are evenly distributed about their central line.

If we wish to also determine whether or not individual values are in control, we may plot the  $3\sigma$  limits for the individual values, which will be twice as wide as the limits plotted for the average values.

#### What Charts Reveal

It should be noted again that such charts as this do not tell us whether the process is operating at a level and with sufficient control to give us results that are satisfactory for our purposes, nor do they tell us what is wrong with our process. The charts do tell us whether our process is doing what it is capable of doing under the operating conditions that are imposed upon it, that is whether or not the measured variables are "in control." They are a picture of what the process is producing and of what the expected variation might be. When an important variation occurs it can then be readily spotted and steps taken to correct the conditions which produced it. In fact, in many cases such a variation can be anticipated from a study of the charts and corrective steps taken.

As an example, let us consider the control of FeO in our basic open hearth slag. For our measure, we determine the FeO in the slag taken at the time of blocking the heat. In Fig. 5 we show the accumulated results from 120 heats. Here, as in the previous chart, individual determinations are plotted as points, the average of each group of four points is shown as a circle, and the range is shown which measures the spread between values in each group. The control limits for the average values are indicated, these limits



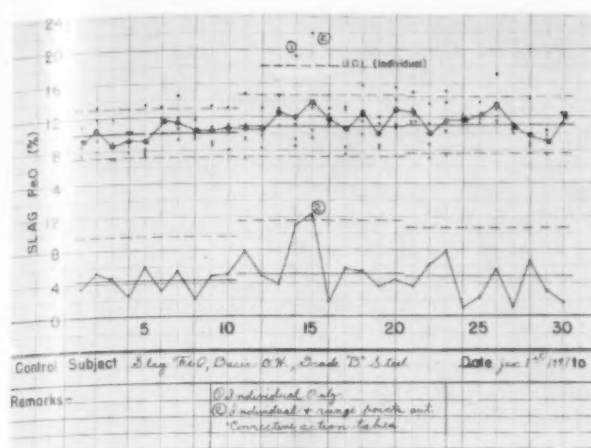


Fig. 5—Control chart for FeO in basic open hearth slag.

being recomputed for each successive lot of ten groups.

The average value of FeO for the first ten groups is about 10 per cent and the average range of FeO about 4.2 per cent and all points are well within the control limits. However, there is a trend toward an increasing FeO content, starting with about the fourth group, and this trend continues through the fifteenth group. In each of the fourteenth and fifteenth groups we have an abnormally high value which increases the range values so greatly that the process is almost out of control in the fourteenth group and is entirely out of control in the fifteenth group. When we plot the control limits for individual points we see that these two points are also outside of these limits. Obviously something has been happening in the practice that has been causing a steady increase in the FeO content and finally the process has gotten entirely out of control. Apparently corrective steps were taken because the range values assume that the same level from the sixteenth group on as they had for the first ten groups. However the values of the FeO content assume a level of about 11 per cent instead of dropping back to 10 per cent, and from the sixteenth group on they seem to fluctuate somewhat regularly about that average line.

This regular fluctuation about the average is of importance because a lack of control is also indicated if the values for ten consecutive averages are all on one side of the average line, although they may be within the control limits. In other words, something has happened to the process so that it is not functioning exactly as it had been and therefore this is indicated on the control chart.

#### Construction of Fraction Defective Charts

A second type of chart which has a useful application in our problem of quality control in the foundry is the Fraction Defective Chart, often referred to as the p-chart. It finds its principal use in the display of the amount of defective material as related to some variable.

The proportion defective,  $p$ , is the ratio of the number of defective units to the total number of units

under consideration. The values of  $p$  for each group (perhaps using heats or shifts as the unit of grouping) are plotted and the average determined for each lot of 25 values (in our practice). This average line ( $\bar{p}$ ) is drawn on the p-chart, usually as a red line. The limits of chance variability (the control limits) are given by the formula  $\bar{p} \pm 3 \sqrt{\frac{\bar{p}(1-\bar{p})}{n}}$  where  $n$

represents the number of determinations in the sample (25 in our case). If the calculated Lower Control Limit is a negative value, it is taken as zero, since it is impossible to have less than zero per cent defective. The Upper Control Limit is drawn in as a broken line.

Such a chart is illustrated in Fig. 6, in which is plotted the percentage of a certain type of castings that were scrapped because of entrapped dirt, as related to time (which is measured in this case by successive heat numbers). The average for each lot of 25 consecutive heats is shown by the central line ( $\bar{p}$ ) and the U.C.L. is calculated from this value and drawn in as shown. From these two lines one can observe the spread or range over which the process is working and also the trend or shift in level from time to time. While such a chart does nothing more than give us a picture of what is being produced by the process as it is now operating, it gives us this picture far better than if one relies on his memory or on a mere examination of the figures on the record sheet. Furthermore, the central line and the upper control limit provide a simple and reliable way of knowing when and if action should be taken. If a plotting falls outside the control limits for no apparent reason, it is desirable to look for an assignable cause and possibly eliminate it. If the plottings indicate the process is in control but at an unsatisfactory level, it is probably desirable to focus attention on the process with the expectation of making a fundamental change.

One other consideration that has to do with the level at which the process is operating is to decide whether or not the process is operating at the lowest level that is economically feasible (where we are considering defectives as the variable). If the process is operating at this desirable level, the function of the chart is to indicate the variations from this level. If the process is not operating at the desired level, the

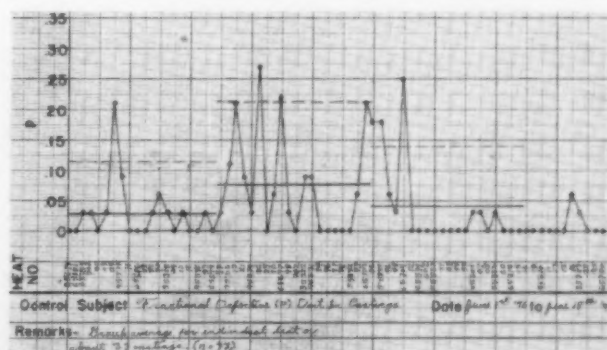


Fig. 6—Example of a p-chart showing percentage of castings scrapped due to entrapped dirt.

chart serves as an indicator of what is actually happening and as a yardstick for measuring progress towards the goal.

### Scatter Diagrams

A third type of Quality Control Chart—and the type that is probably the most familiar to us—is the scatter diagram or scatter chart. This is the type of chart that is used when we wish to determine the correlation between two variables that are related as cause and effect. For example, in Fig. 7 are plotted data representing the percentage of  $\text{SiO}_2$  in acid electric furnace slags, against the viscosity or length of run of the corresponding slags as measured in the viscosimeter. It is evident from an inspection of the scatter

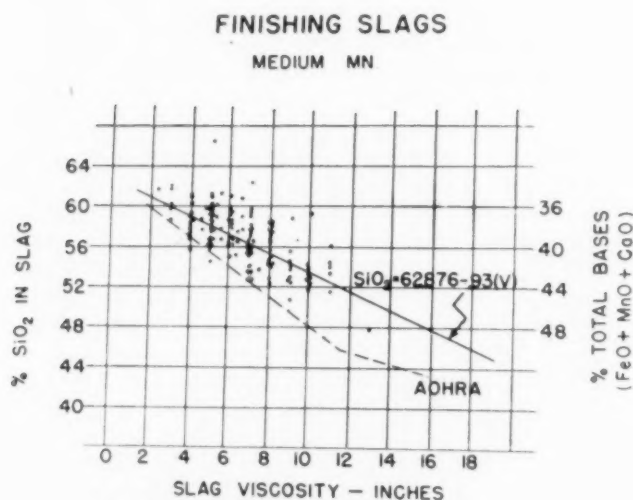


Fig. 7—Example of a scatter diagram.

diagram that there is a relationship between the two variables because, in general, the shorter the length of run of the slag the larger the amount of  $\text{SiO}_2$  present. A negative correlation is said to exist because as the slag length increases, the  $\text{SiO}_2$  content decreases. If the relationship is such that an increasing cause yields an increasing effect, the correlation is said to be positive.

Because of the scatter of values of  $\text{SiO}_2$  for each slag viscosity, it would be difficult to draw by inspection, the line that would best represent the data. The equation for this line can be calculated however, and this is known as the Estimating Equation or the Regression Equation (a brief summary for the method of calculation of which is given in the Appendix). Such an equation enables one, then, to estimate the value of one series for a given value of the other series.

For systems where the relation is not that of a straight line but of a curved line, a more complex relationship exists which can be reduced to an equation by means of a curvilinear correlation. And if we have one dependent variable and two or more independent variables, multiple correlation is applied.

The scatter of data around the regression line is measured as the coefficient of correlation and is a factor which expresses this scatter. This factor is a

numerical one and is independent of the units of the original data. If  $r$  (the coefficient of correlation) =  $+1$ , the correlation between the two variables will be perfect with both of them increasing. If  $r = -1$ , the correlation will be perfect with both of them decreasing. If  $r = 0$ , the scatter will be a random one. (The method is outlined in the appendix for calculating the coefficient of correlation and for determining whether or not it is significant.)

It should be noted that the measure of correlation merely indicates that the relationship may exist. The classical example is that there was a high degree of positive correlation over a period of years between teachers' income and the amount of money spent for liquor. The conclusion that could be drawn was that the increase in liquor sales was caused by the increase in the income of teachers when the underlying variation really was national income, fluctuations in which caused variation in both liquor sales and teachers' income.<sup>5</sup>

### Application of Quality Control to Foundry Practices

In the application of the principles of Quality Control to the foundry processes it must be recognized that the technique of a skilled melter, molder, or core maker cannot be recorded on a control chart and any attempt to do so usually arouses opposition. What can be done, however, is to control the quality of the raw materials and the properties of the sand mixes to measure the process operation, and to provide both workman and management with a factual measure of performance by studying the quality of the castings produced and the detailed causes of rejection.

The control charts in the authors' plant have, therefore, been set up on three general bases:

- (1) Control of Materials—such as the establishment of Quality Control Charts for the physical properties of the several grades of sand used.
- (2) Control of Processes to include such items as:
  - (a) Charts showing carbon and manganese control for each of the several grades of steel for each of the several units of furnaces;
  - (b) Charts that measure such variables as amount of lime and spar additions in the basic electric furnace and the corresponding sulphur and phosphorus reduction that may have been effected;
  - (c) Charts to show power consumption for the several electric furnace units;
  - (d) Charts to show other items of chemical control;
  - (e) Charts to relate slag viscosity from the electric furnace units with manganese recovery or with other variables, such as operator performance.
- (3) Measure of Process results—to include such items as:
  - (a) Charts recording physical properties of the steel being produced;
  - (b) Charts recording defectives, as related to possible variables.

In addition, when we desire to compare the performance of two units or two sets of units, the Quality

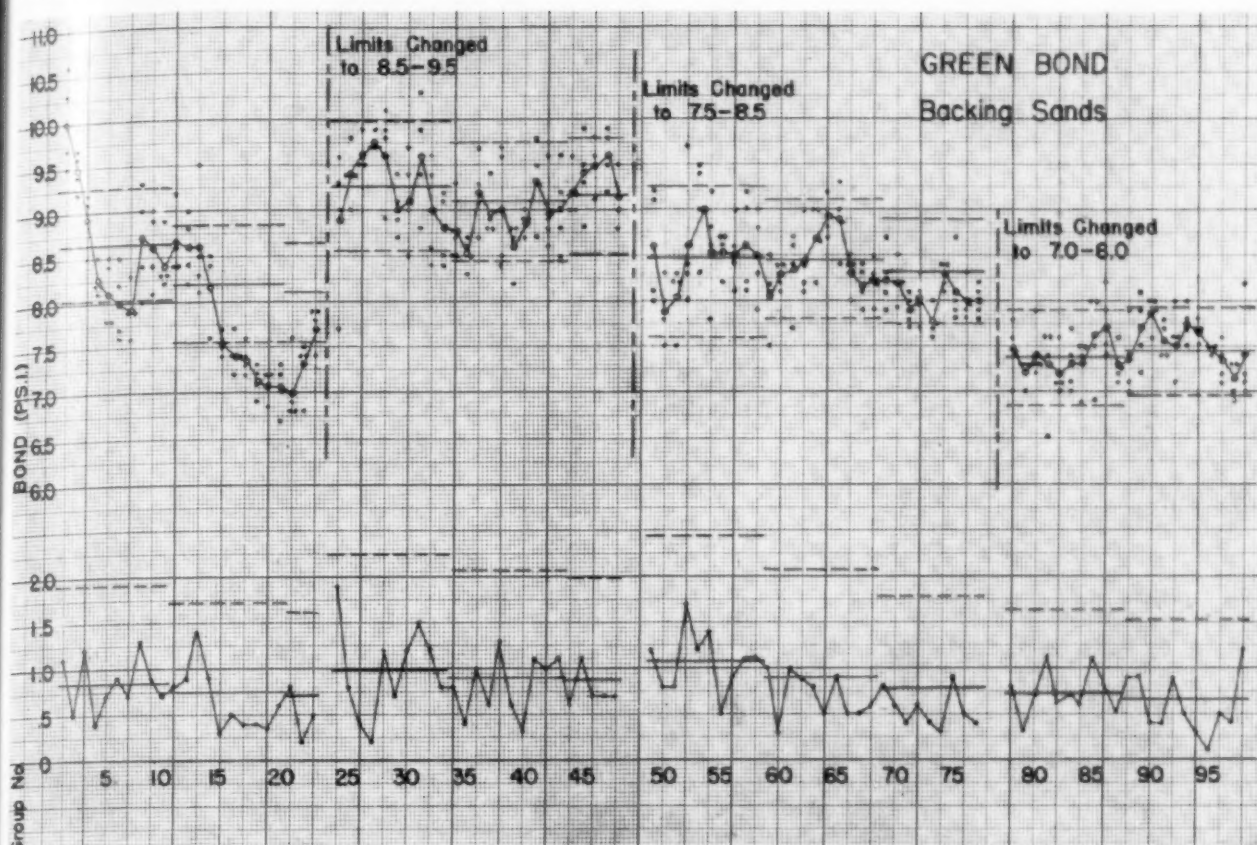
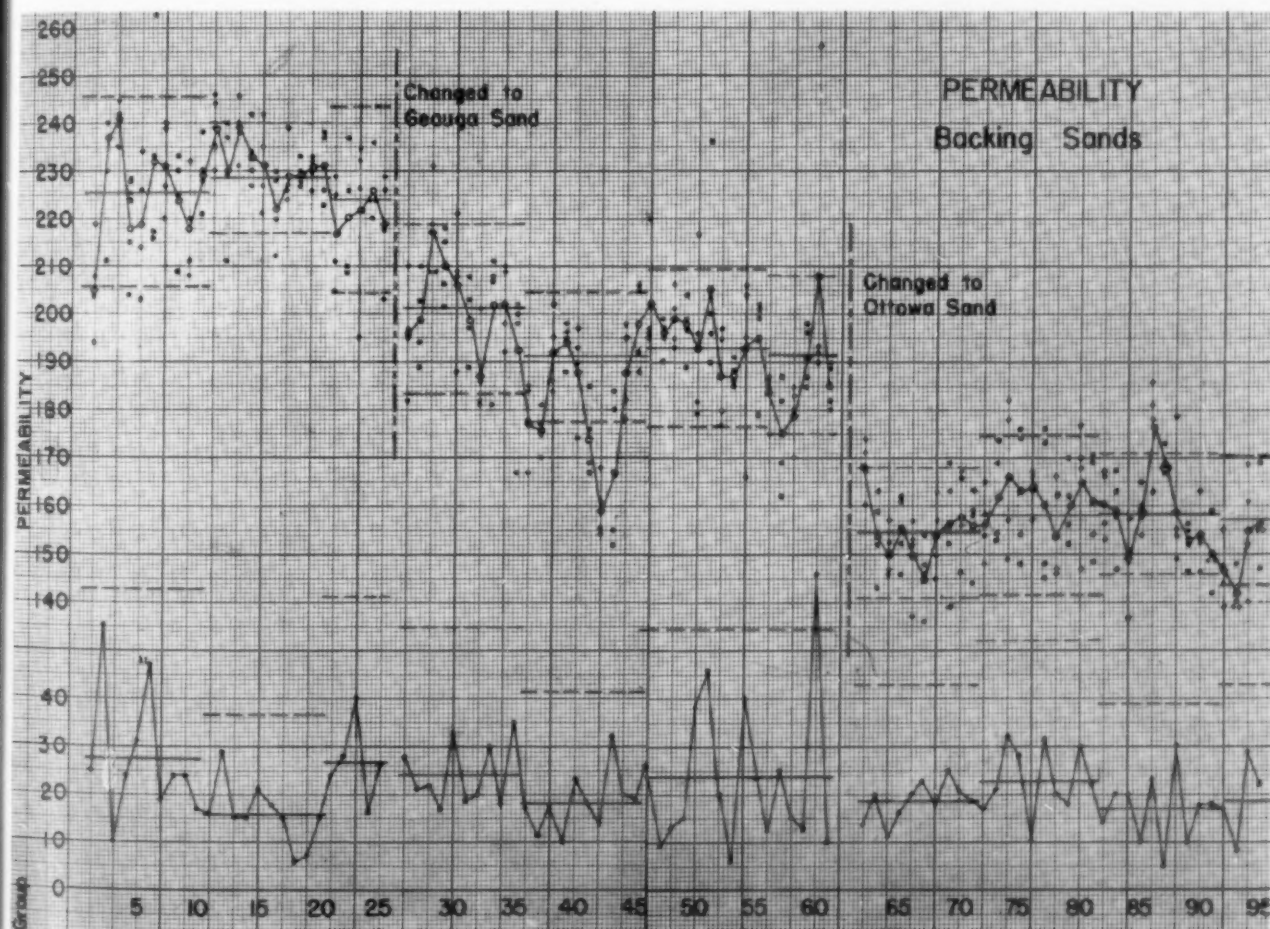


Fig. 8—Control chart for green bond of backing sands.

Fig. 9—Chart for permeability of backing sands.





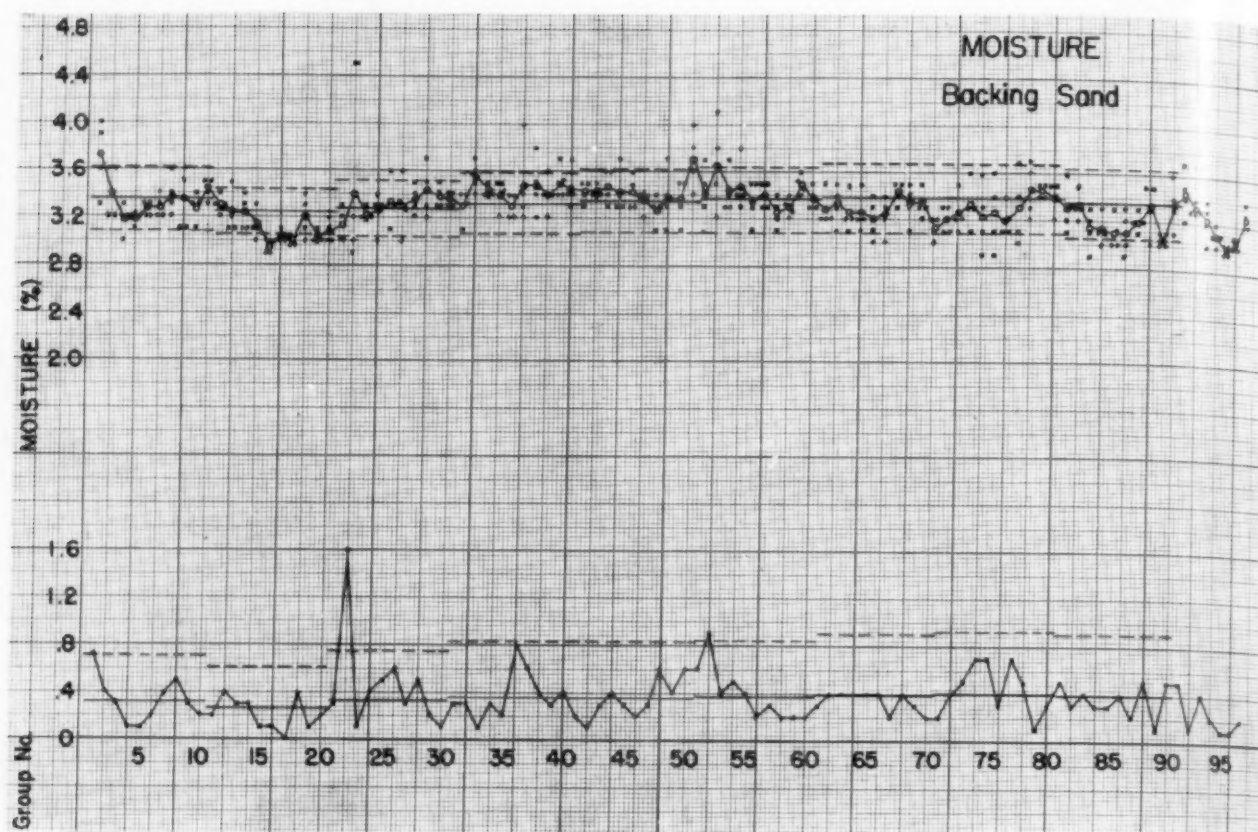


Fig. 10—Control chart for moisture content in backing sands.

Control Chart gives us a picture of the relative performance (as measured by a given property). Alternatively, scatter charts may be constructed, the equation of the most probable curve calculated, and then tested by calculating the coefficient of correlation and its significance. In this way a statistical measure of the performance with which we are concerned is secured.

#### Type of Charts

For our purposes, the charts are in the form of long horizontal wall charts, which we keep posted on our "Quality Control Board." When we first started, our Quality Control work, we tried to plot the data on long vertical wall charts, where all the data for one heat would be shown on the same vertical line. However, it is the authors' opinion that the number of independent variables is too large for any but the most obvious inter-relationships to be detected in this way.

The main purpose of the control chart is to provide a measure of the variability of the several steps in the process and it has been found that in many cases the chart is a quite sensitive indicator of process control.

#### Control Chart for Raw Materials

To illustrate these principles, some charts covering specific applications are presented.

In Fig. 8, 9, and 10, are presented typical charts

such as are found useful in the sand control. For example, the physical properties of the heap (or backing) sand over a considerable period of time are plotted. Each dot represents the value obtained by averaging the individual values for the 20 or 25 mills of sand made during an 8-hour shift. It will be noted that the permeability of the sand is the property that has shown considerable change. This is because a finer grade of sand was introduced into our system about the time represented by the 25th group and still another grade of sand about the time represented by the sixtieth group. The range or spread of the permeability values varies as shown, indicating that the degree of control was at one level for the first twenty groups, then was poorer during the time represented by the twentieth to the thirtieth groups, and finally it returned to about the original level or degree of control for the rest of the time under consideration. Apparently the range or the degree of control is not correlated with the changes in the kind of sand used. However the level of the permeability values certainly changes significantly with the introduction of each type of sand.

The moisture values were held at a constant level and show a uniform range during the period under consideration. The level of the green bond values was twice adjusted, as noted on the chart. Otherwise the results were quite uniform.

From the foundry viewpoint, it seems of interest that there is little correlation between casting defects, that might be attributed to variations in sand properties, and the permeability of the backing sand which would suggest that perhaps the permeability is not a

critical control factor. In the interest of brevity the charts representing this correlation are not presented.

### Control Charts for Processes

The need for chemical control of the composition of the metal is well recognized by most foundries and carbon and manganese determinations are considered essential.

For our purposes, we established Quality Control Charts for carbon and for manganese for each grade of steel that we were producing. An example of such a chart is shown in Fig. 11. On this chart we had plotted both carbon and manganese values for our

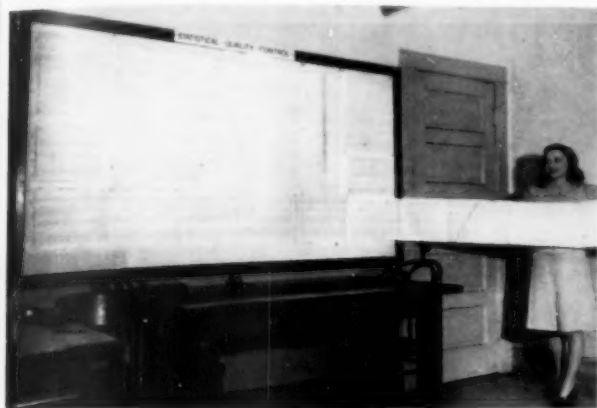
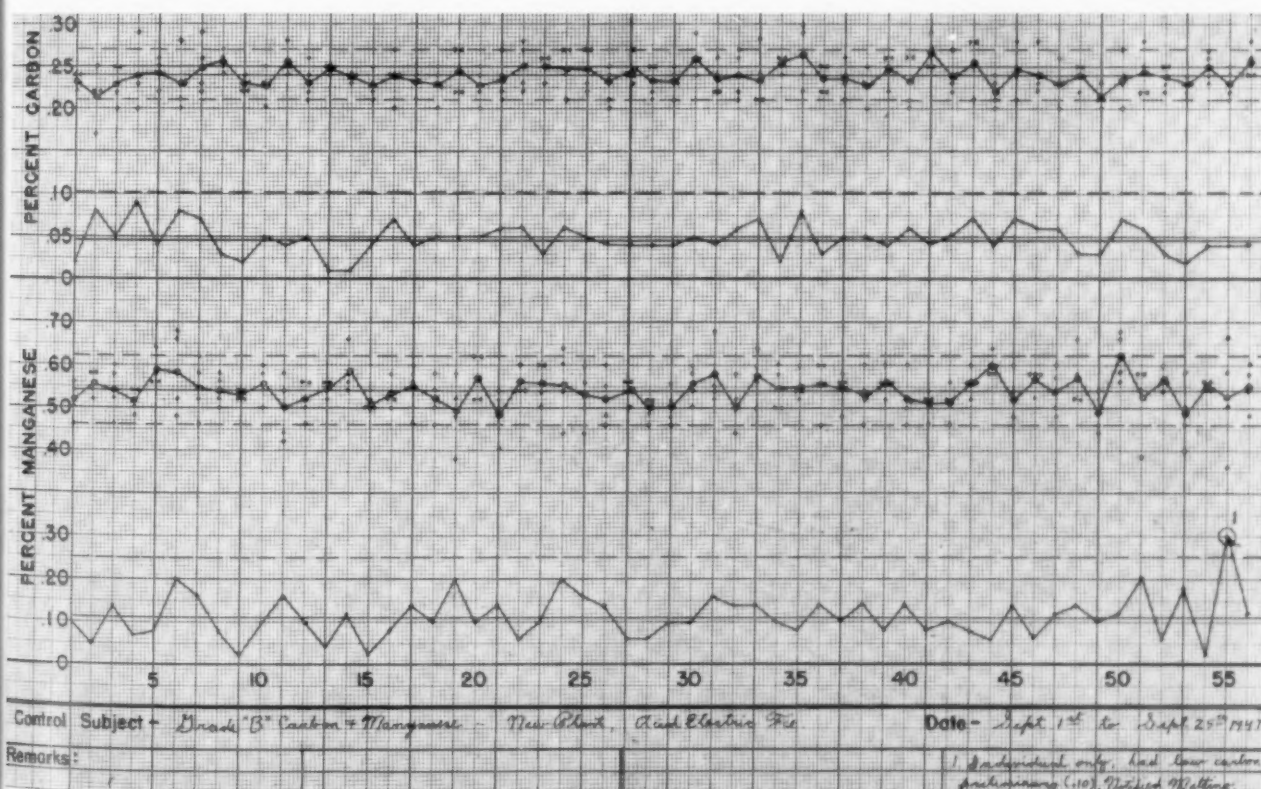


Fig. 11—Quality control chart room at authors' plant.

Fig. 12—Example of quality control chart with projected limits.



Grade "B" steel, produced by the acid electric process for heats made from March 15 to September 30, 1947. Some 434 groups of four observations have been accumulated. Control limits have been calculated to include each successive lot of ten groups (taking 50 groups at a time and dropping a lot of ten groups each time a new lot of ten is added). Because both the level of values and the range have been so uniform over such a long period we are justified in projecting these limits indefinitely and in trying to determine the cause when any future determinations fall outside of their established limits (that is, until a set of conditions are imposed on the process that will establish another universe in which the system is operating). An example of this type of chart with the projected limits is shown in Fig. 12.

Similar charts may be found useful for such variables as the control of alloy content where a specification must be met. In such cases one may wish, for the sake of economy, to stay on the low side of the range and the control chart will enable the melter to determine the safe value at which to aim so as to both stay inside the prescribed range and yet not use any more alloying material than is necessary. In other words, a control chart acts as a guide so that the melter can establish a closer range than that given in the specifications if he wishes to effect economy in the use of alloys at the expense of more rigorous operational control.

Another type of control chart that may be found useful from the managerial viewpoint may be the power consumption chart for the electric furnace operation. Here, again, such charts can be broken down so that we compare the power consumption of several first helpers, both as to the level at which they

operate and as to the fluctuation in their operation over a period of time. As an example we present, in Fig. 13, the performance of three operators, working on the same battery of furnaces over a 6-week interval.

Operators "A" and "C" are apparently operating about the same for their average power consumption is almost identical as to their range of values. However "C" is a little more erratic than "A" for he has two average points out of the control limits. Operator "B" evidently uses more power and also has a somewhat greater spread of values than the other two operators and this confirms our opinion of his work.

Such comparisons of other characteristics may also be readily made between the operation of batteries of furnaces or for comparison of the performance of such units as acid and basic-lined furnaces in the same battery.

From such charts management can readily measure certain characteristics of the performance of individuals and, on that basis, take any necessary steps to improve such performance. Changes in the processes can also be evaluated on a fairly sound basis rather than by estimating what has happened by looking at a sheet of figures.

#### Control Charts for Measurement of Process Results

Process results may also be evaluated in a manner exactly similar to that described.

The tensile test results are the commonly accepted measure of the quality of the steel being produced by a given process. Therefore quality control charts for each of the measured properties (yield strength,

ultimate strength, elongation and reduction of area) for each production unit and type of steel form an adequate measure of what that process is producing and also act as an indicator of what it may be expected to produce.

As an example consider the chart shown in Fig. 14 in which we are measuring the ultimate strength of Grade "B" steel produced by the acid electric process, compared with that produced under similar conditions by the basic electric process. It will be noted that

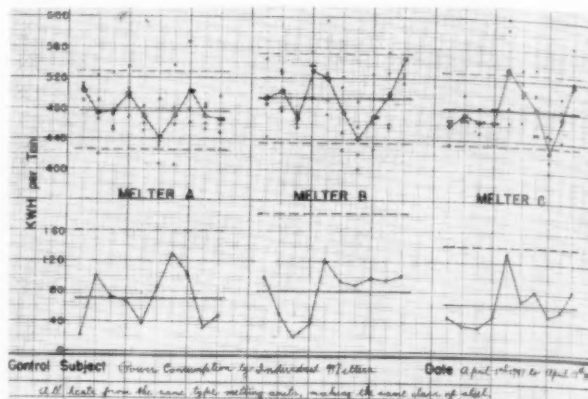
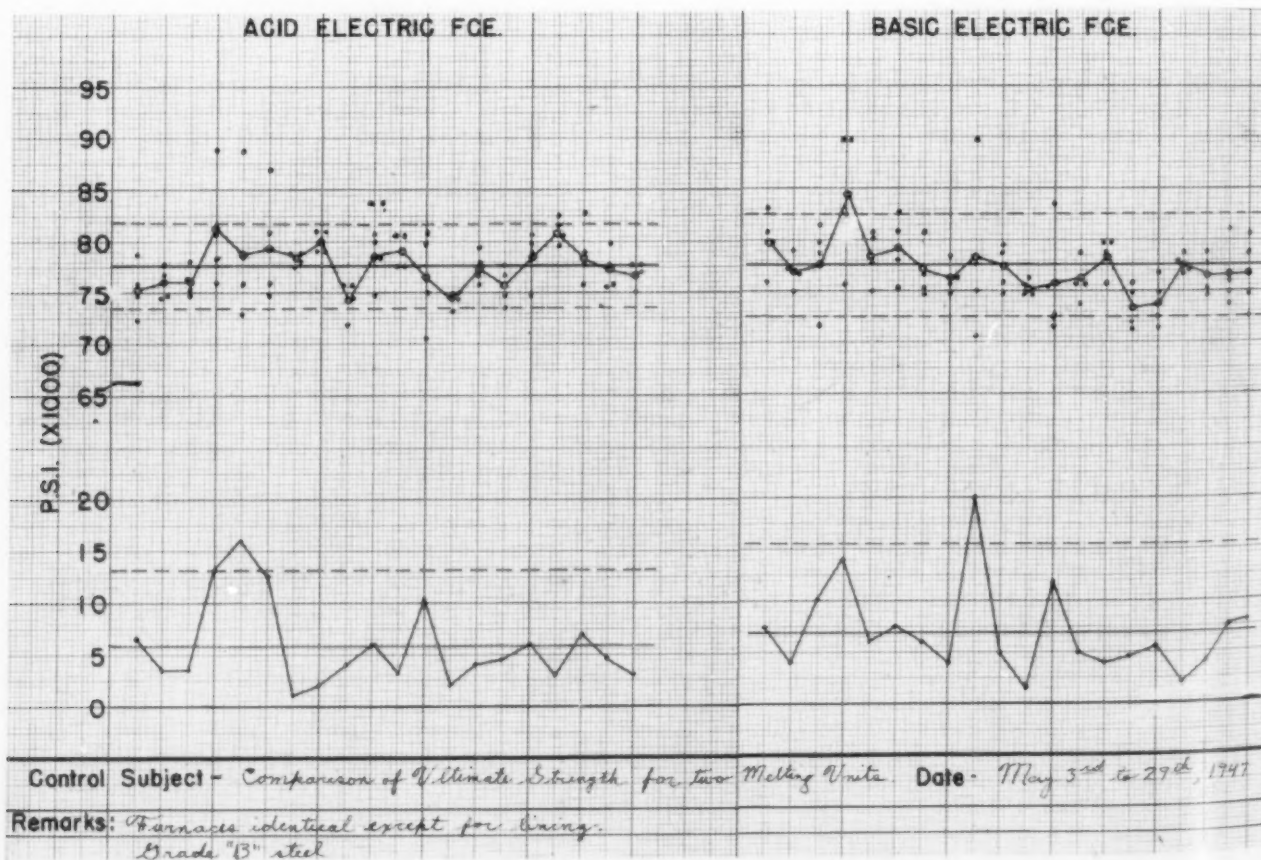


Fig. 13—Control chart showing power consumption of battery of furnaces under three operators.

Fig. 14—Control chart showing comparison of ultimate strength of Grade "B" steel produced in acid electric and basic-electric furnaces.





the level of strength of the steel produced by both processes is practically the same, but that the spread of results is a little greater for the test results of the basic-furnace material. On two occasions, the test results of the basic furnace material went out of control (the fourth and ninth groups) while only one group of results from the acid furnace material (the fifth group) was out of control.

In this connection, such a means of recording test data may be used as a basis for permitting reduction of testing from a requirement of one test for each heat to say one test for every third heat, as long as certain conditions of control are maintained. Such premises formed the basis for the reduced testing permitted by the quality control plans in operation during the war and since that time. The control charts measure the fluctuations in the process of manufacture and, in the words of one such plan, "As long as the differences measured are fluctuations that can be ascribed to chance causes, the customer will be satisfied to test less of the product and will have increased confidence that all of the product is reasonably uniform."

"When fluctuations in successive test results are too great to dismiss as chance variations, we will have lost our assurance that the product is reasonably uniform and we will want to increase the amount of testing that is done. We will now assume that some undesirable cause of variation that can be identified and removed, has crept into the production process. The increased testing will continue until successive test results once more indicate that the process is under control."

Such a procedure in no way minimizes the maintenance of a high standard of quality. On the contrary, a high standard must be maintained for the system to stay in control and thus benefit by the reduced testing that is permitted but when such a standard is maintained the benefits of decreased testing cost, flow of production, and related items are considerable.

A chart showing the yield strength and the elongation for some 160 tensile tests of steel made to meet Federal Specifications QQS-681-4C2 is presented in Fig. 15. On this chart we would call attention to the elongation values in group 144, where one value is an unusually high one. Although the value is excellent from the viewpoint that good ductility is desirable, it still throws the system out of control, bringing the average for the group outside the Upper Control Limit and also bringing the range for this group outside its Upper Control Limit.

The elongation values for groups 145 to 152 are also out of control because the Lower Control Limit for the average values falls below the minimum value of the specification. This is true in spite of the fact that all average values are within control limits, but the spread of values is great enough that there is a possibility that some values will not meet the minimum specifications.

Numerous other details could be recited pertinent to the calculation of control limits and to the steps necessary in order to determine whether or not the

system is in control and that therefore reduced testing is possible but space prohibits.

Two further examples may be given to show the application of quality control to process control where the concern is with the classification of defective castings.

In one case we are concerned with determining the kind and the amount of scrap that is encountered in producing a certain type of casting over a considerable period of time. For example, let us consider the production of a fairly large number of a certain type of journal box. In this particular case we had not previously made this type of journal box, so that we were concerned with determining quickly and accurately the amount and type of scrap castings that were being produced and with measuring the effect of changes in the pattern in the molding practice, etc., on the scrap produced.

In order to secure the correct data regarding the scrap being produced, the inspectors were furnished data sheets (such as is shown in Fig. 16) on which they easily recorded the details concerning the defective castings as the castings were inspected. These data were then tabulated and plotted in the form of a histogram, as shown in Fig. 17. The important types of defects were plotted in different colors on the original chart so that, at a glance, one could determine the kind and the relative amount of the prevailing defects. Whenever a change was made on the pattern

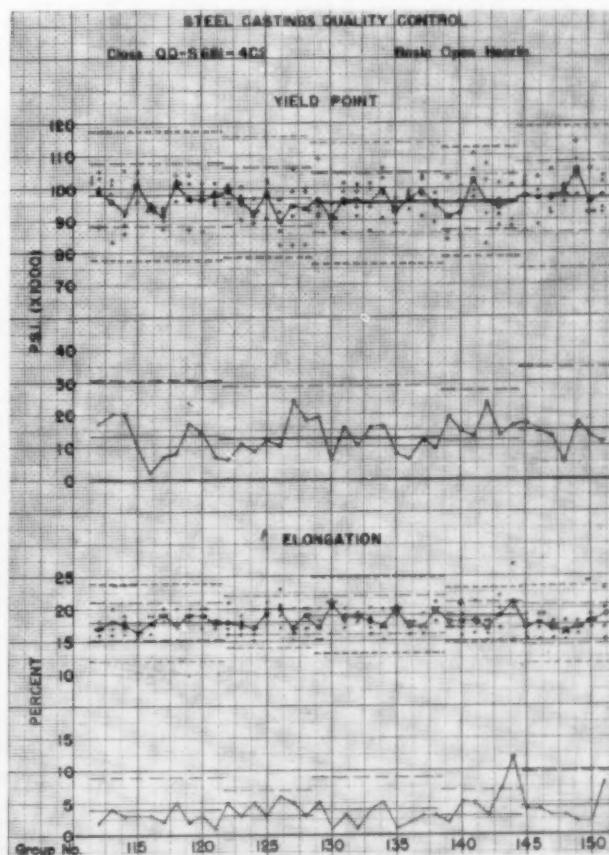


Fig. 15—Chart showing yield strength and elongation for 160 heats of steel.



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and on the left hand side are plotted the per cent of castings from each heat that are scrapped because of pinhole porosity. The average percentage for each group of four heats is plotted on the right-hand side of the chart. Average values and control limits are computed and plotted for each group of ten values.

From such a chart we note the frequency of occurrence and the distribution by heats of the defect and we also note the expected occurrence of it for the conditions under which we are operating. Again, it should be said that maybe this level of occurrence is not as low as could be desired but that the Quality Control Chart can do nothing more than draw a picture of what is happening. It cannot correct the conditions that exist but it can point to these conditions.

Armstrong,<sup>6</sup> in an article in *Industrial Quality Control* magazine, presents a p-chart in which the variable is the amount of rejected side-frame castings produced in a certain steel foundry. The chart is reproduced in Fig. 19 and illustrates how the level of rejections has changed as well as the level of expected rejection. Whenever the rejections are at a level that

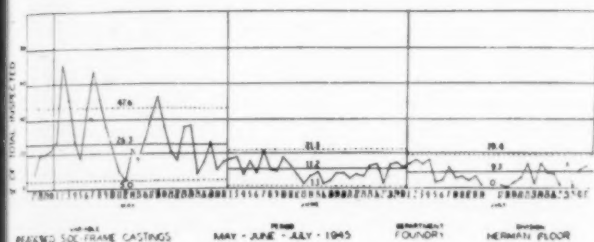


Fig. 19—A p-chart showing occurrence of rejections of side-frame castings.

falls outside of the control limits, it will be recognized that unusual conditions have been present. Otherwise the chart shows the level at which the process is operating and how this changes from month to month. It also shows very distinctly when a change in operating conditions has effected a change in the quality level of the product.

Attention is directed to the paper published by the Electrochemical Society: "Acid Electric Steel for Castings" by S. F. Carter and C. K. Donoho. The paper was preprinted for the general meeting of the Society, held April 9-12, 1947, and includes an excellent application of statistical analysis to determine whether or not the observed differences were significant.

### Summary

It has been the purpose of the paper to set forth some of the fundamental concepts of Statistical Quality Control with special reference to their application in the operation of a steel foundry. By means of this tool, process measurements become more than a tabulation of figures but they become a picture of the process and an indicator of expected future performance. The control charts do not tell what changes should be made to provide the desired quality levels but they do give immediate warning when something

about the conditions of the process has changed erratically.

In order to be effective such charts must be maintained up to date and action taken when a warning indication is given that some erratic change is taking place in the system. Otherwise the effort required to maintain such charts is wasted.

One other application that has been indicated is the use of such mathematical tools to measure the significance of differences between two processes, for example, without resorting to the more conventional idea that "we will take the average and let it go at that."

Such charts further serve to give management an accurate and clear cut picture of the performance of the process under consideration.

### Acknowledgment

The authors would express their thanks to Dr. H. A. Schwartz for his encouragement and his advice concerning the preparation of this paper; to Mr. W. M. Ewing for permission to use the data and for his encouragement to publish it; and to Mr. A. G. Ellis for his fine assistance in the preparation of the paper.

## APPENDIX

### Method for Calculating the Estimating Equation or the Regression Equation

The Estimating Equation is an algebraic expression of the relationship between two variables, the general form of which is  $Y = a + bX$ , where  $Y$  is the dependent variable, and  $X$  the independent variable. The equation represents the straight line which best fits the data. The values for the coefficients are determined by the method of least squares.

Let us assume that we have  $n$  pairs of observations of  $X$  and  $Y$ . By tabulating these and performing the necessary addition and multiplication, values for  $\Sigma X$ ,  $\Sigma Y$ ,  $\Sigma X^2$ , and  $\Sigma XY$  can be readily obtained. ( $\Sigma$  is the symbol for the process of summation.) The values of the coefficients  $a$  and  $b$  can then be determined by solving the two normal equations:

$$a = \frac{\Sigma X^2 \Sigma Y - \Sigma X \Sigma XY}{n \Sigma X^2 - (\Sigma X)^2}$$

$$b = \frac{n \Sigma XY - \Sigma X \Sigma Y}{n \Sigma X^2 - (\Sigma X)^2}$$

Knowing the coefficients, the equation of the line is readily written.

If, on the other hand,  $X$  is the dependent variable and  $Y$  the independent variable, the normal equations become

$$a = \frac{\Sigma Y \Sigma XY - \Sigma X \Sigma Y^2}{n \Sigma XY - \Sigma X \Sigma Y}$$

$$b = \frac{n \Sigma Y^2 - (\Sigma Y)^2}{n \Sigma XY - \Sigma X \Sigma Y}$$

from which the equation of the line is readily determined.



### Method for Calculating the Coefficient of Correlation

The Coefficient of Correlation,  $r$ , is the measure of the scatter of results about the line of regression.

It expresses the amount of variation in the dependent variable for which the regression equation has failed to account.

The values of the coefficient are expressed in numerical terms which are independent of the units of the original data.

The coefficient of correlation is a number varying from  $+1$ , through zero, to  $-1$ . The sign indicates the slope of the regression line. When the correlation is perfect,  $r = \pm 1$ . When no correlation exists,  $r = 0$ .

One of the most frequently applied formulae for calculating  $r$  is

$$r = \frac{\sum(X - \bar{X})(Y - \bar{Y})}{\sqrt{\sum(X - \bar{X})^2 \sum(Y - \bar{Y})^2}}$$

When  $\bar{X} = \frac{\sum X}{n}$  and  $\bar{Y} = \frac{\sum Y}{n}$

To determine whether or not the coefficient of correlation is of real significance, the formula

$$\sigma_r = \sqrt{\frac{1 - r^2}{n - 2}}$$

is used to calculate the standard deviation.

If the value of  $r$  is within the  $3\sigma_r$  limits the correlation is undoubtedly real.

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### DISCUSSION

Chairman: J. F. RANDALL, Ford Motor Co., Detroit

Co-Chairman: D. C. ZUEGE, Sivy Steel Casting Co., Milwaukee

R. G. STORMS (Written Discussion):<sup>1</sup> Messrs. Johnson and Fisher are to be commended for their presentation of the fundamental concepts of quality control as related to steel foundry operation. This is a paper which has been written in a comprehensive manner for the man on the job and, as such, is of great value.

We at Hamilton, Ontario, have been using control charts and histograms for the last few years in an effort to bring the many variables in the foundry under some semblance of control. In some instances we have had very satisfactory results; however, in others we have found that the simple control chart is inadequate to obtain the significance of the many variable factors involved in producing steel castings. With this latter in mind I am not inclined to agree with Messrs. Johnson and Fisher in their tentative assumption on "That permeability of the backing sand is not a critical control factor" in the quality of castings as concluded from a simple correlation between casting defects and permeability.

<sup>1</sup> Dominion Foundries and Steel, Ltd., Hamilton, Ontario, Canada

The writer feels sure that the two variables, permeability and sieve analysis, must be studied in correlating rejections together since a sand with a uniform grain size may give the same permeability as a sand with larger grains mixed with smaller grains and the two sands may have widely different effects on the quality of the finished casting.

The writer brings this matter up for he judges that persons starting a system of statistical control of quality should realize that many varying factors must be reviewed in studying the effects of process conditions on the quality of the finished castings, and that the obvious answer in a study may often be clouded by the action of allied variables.

MR. JOHNSON: We appreciate the comments made by Mr. Storms. We know what he is talking about when he says that you cannot always make statistical control as simple as you would like to. His point is well taken, that in any process where many variables are present you have to pick out those which seem significant and see if any correlation exists.

R. C. WAYNE:<sup>2</sup> What is the basis for selection of the groups when you plot a number of groups of points for the averages? I noticed that Mr. Johnson used four in one illustration. Why would you use four rather than two, three, or five, and how would you arrive at that selected figure?

MR. JOHNSON: We used four because Army Ordnance used four. The idea is that if you use several points, the chance of error due to variability is less than if you just use one or two or a small number. Many people use five points but we used four because that was the number used by the Army Ordnance. If you refer to the tables which give the factors for computing 3-sigma control limits, such as the American War Standards Tables you will find that they use, three, four, five. I believe up to 25 observations in the sample. It just happened that four was the number that the Ordnance people used and we thought they knew what they were talking about.

W. K. BOCK:<sup>3</sup> There are cases, for example, in reduction of area, where the distribution is not this normal distribution that Mr. Johnson showed for the individual tonnage. However, this system works on a normal distribution. The average of groups comes out to be a normal distribution. The reason for using four, it is just a convenient number to handle mathematically. I think that may be the origin of its use. There are people who have charts where their data comes to them in such form that they cannot conveniently handle a group of four, in which case they set up charts for a group of three or five or whatever is convenient. The prominence of four comes down to the fact that four is the handiest number to use, but you can use any other grouping you want to use.

CHAIRMAN ZUEGE: How much time is involved in keeping up a system of this kind?

MR. JOHNSON: At our plant we have one man who gives perhaps 75 per cent of his time to this work. His time is not spent entirely in plotting the points on the chart. Part of his time is spent in doing something about it. He contacts the melting superintendent or the sand technician or whoever should be contacted. He does quite a bit of work interpreting this data. We feel that we save money by having that man carrying out this work. Other folks may not want to spend quite that much time on it. The amount of effort spent depends on how much value you place on it.

CHAIRMAN ZUEGE: How much additional time is taken in keeping up the graphs?

MR. JOHNSON: That is this one man's job. He keeps the graphs up. He does the whole thing.

J. A. RASSENFOS:<sup>4</sup> In the paper, one graph shows the relation between the control of permeability and casting quality. Have the authors tried to evaluate the effect of properties other than the permeability of the sand?

MR. JOHNSON: We showed the green bond, the permeability, and the moisture content for those particular sands. The moisture content was kept practically constant and we controlled the changes in green bond. The only uncontrolled variable therefore was the permeability and this changed with

<sup>2</sup> Steel Founders' Society, Cleveland

<sup>3</sup> National Malleable & Steel Castings Co., Cleveland

<sup>4</sup> American Steel Foundries, East Chicago, Ind.

the type of sand that was used. We have done some work in which we correlated permeability with the defects which were produced and found very poor correlation.

ERIC WELANDER:<sup>2</sup> In Fig. 4 you show carbon analyses plotted. Do you take a number of carbon determinations and make one plotting for the average of those determinations?

MR. JOHNSON: Each morning the man in charge of this statistical control gets the analyses from the day before and spots the individual points on the charts. Then he takes each group of four and gets his average for the groups. If we have made 24 determinations during the day, he has six groups of four and draws a line connecting the average points to get the trend. We have our limits projected on the charts and as long as the average points are within those limits, we are satisfied that the process is working satisfactorily.

MR. WELANDER: You are satisfied that it was working satisfactorily yesterday. You are plotting past history.

MR. JOHNSON: It is true that we are plotting a day late. We can extend that and take the chart up and have the chemist put in the points if we want to, as he gets them.

MR. WELANDER: Even in that case you would have to wait for four determinations would you not?

MR. JOHNSON: In that particular case we have extended our limits because we have this background of 1600 heats as a basis for establishing the limits. As long as the new points come within those limits, we are satisfied that we are doing satisfactorily. Now, if the process is changing, then we will still have to wait until we get a group of four determinations (in our case) before we can determine whether or not the average is within control limits.

<sup>2</sup> Union Malleable Iron Works, East Moline, Ill.

MR. WELANDER: The point is that this control does not work quite as well in the foundry as it does on a screw machine, for example, where you can take five measurements immediately and then set your machine from them. You are plotting past history for the most part.

MR. JOHNSON: That is right. You can plot it, however, as fast as you get the data, which is what you do on a screw machine.

MR. WELANDER: You do not get the data fast enough.

MR. JOHNSON: That may be so.

H. H. FAIRFIELD:<sup>3</sup> Naturally, the operating man would rather plot one point at a time, and there are statistical laws which allow him to do that by making use of the laws of runs and cycles. There is one inference in the paper that, fortunately for the authors, who do good work, no doubt, their process is well under control. I have made quality control charts in a number of steel plants and for the most part they are actually out of control.

There is one point I might mention that we use on Grade "B" steel castings. We use correlations to determine the relationship between tensile strength and reduction of area; that is, for each tensile strength we know that we should get a certain reduction of area. We make a control chart of the deviation from that table. In that way, we can detect the tendency for flake or aluminum nitride inclusions quickly, and since we applied that method, we have been able to increase our average reduction of area about 5 or 10 per cent.

MR. JOHNSON: That is a most interesting application and we are glad to know about it.

<sup>3</sup> The Wm. Kennedy & Sons, Ltd., Owen Sound, Ontario, Canada

# TESTING REFRACTORIES FOR THE FOUNDRY

By

S. M. Swain\*

## ABSTRACT

*The careful planning, selection of objectives, justification of costs and safety factors for a testing program are discussed. Detailed recommendations are made for sampling brick and other refractory materials. Critical discussions and brief descriptions of testing are given for refractoriness, reheat change, strength, load bearing ability, spalling, porosity, water absorption, specific gravity, bulk density, chemical analysis, size, thermal conductivity and slagging of fire brick; for refractoriness, screen analysis, bonding strength and consistency of mortars; for refractoriness, spalling, workability index and firing shrinkage of plastic refractories; and reheat strength and refractoriness of castable refractories. Three general types of testing programs are discussed and recommendations are made for small and large foundries.*

## I. Introduction

THE PROPERTIES OF REFRACTORIES and the methods of testing them follow similar properties and methods for bonded foundry sands. In both types of products there is a softening range involved rather than a sharp melting point. Near the upper temperature limit of use of these materials, changes occur at an increasingly rapid rate. It is unusually difficult to isolate one characteristic of these materials at high temperatures and then to measure it alone with accuracy. High temperature testing of these materials is expensive, time consuming and difficult to conduct.

## II. General Conditions of Testing

Any testing program should be well thought out before it is started. The work should begin with the most simple tests. Definite arrangements should be made for accumulating data and for filing it in a readily usable form. By this process a record is built up on the uniformity of products used and a better understanding of the problems involved is acquired. The information is then a sound foundation for slowly expanding both the extent of the testing and the application of the results.

The collection of test data by itself is a misleading and wasteful procedure. Each testing program should contain definite arrangements for correlating the

properties of refractories with service results obtained and with service conditions to be met. One set of studies will not last indefinitely. Modern progress means continual changes in furnace conditions so that it is necessary to make definite arrangements for reexamination of all of the work involved at set regular intervals, to be sure that changes are not passed by in the rush of conducting the tests.

Considerable thought should be given to the physical arrangement of the entire program not only to reduce the expense of the work but to insure that all of the data will be recorded, filed and used. Arrangements should be made for the easy but accurate selection of the samples. The testing equipment should be located as conveniently as possible so that samples do not have to be transported long distances and possibly be contaminated or lost on the way. It is very important to set up printed forms with proper spaces so that it is easy to record the data required. If possible, space should be provided for the calculations so that the original data, calculations and references are on one sheet. It is very convenient to have the sheets in such form that they can be filed in a loose leaf book in proper chronological order. Full attention to details of this sort will pay big dividends in the completeness and availability of the information obtained.

Throughout the program, safety should always be kept in mind. Careful consideration should be given as each program is planned, to avoid accidents due to improper application of materials. This work should definitely be conservative in this respect. Any new materials should be tried out in locations of moderately severe conditions only until they have been proven satisfactory in these locations before being tried out under conditions of more severe wear.

## III. Sampling

The most important part of any testing procedure is the selection of a true sample. If a compromise must be made, a skilled technician should do the actual selection of the sample and any economizing be done on personnel to run the tests. Errors in the test procedure are easier to locate and correct than are errors in the sampling process.

\* Director of Research, North American Refractories Co., Cleveland.



A good method of sampling clays and similar granular materials is given in the *FOUNDRY SAND TESTING HANDBOOK*,<sup>1</sup> Section 2.

The sampling of fire brick involves similar basic principles but different details and technique. With the normal methods of loading brick into a railroad car or onto a pile in storage one good method is to draw imaginary lines from diagonally opposite corners of the top of the pile and then take sufficient samples at regular intervals along these lines as shown in Fig. 1 to obtain the number required. An extra number may be taken and the total reduced to the desired amount by quartering. The A.S.T.M. Specifications C63,<sup>2</sup> C64,<sup>3</sup> C106<sup>4</sup> and C153<sup>5</sup> provide for the selection of 22 samples for each lot of 50,000 brick or less. This size of sample is sufficient for most testing programs.

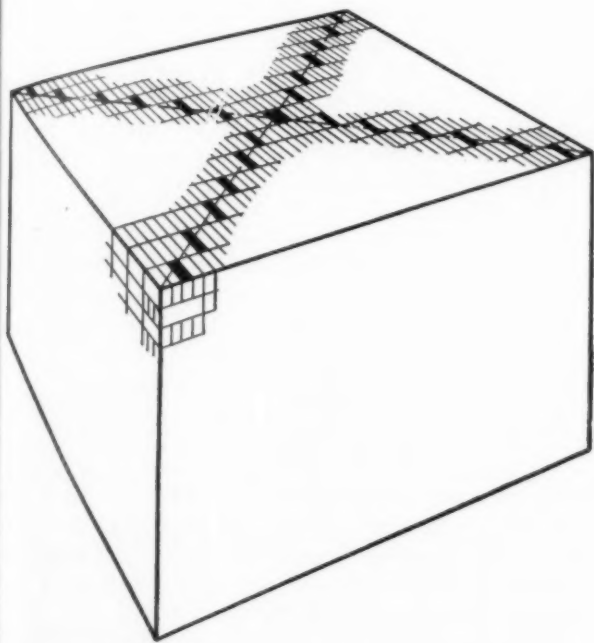


Fig. 1—Method of sampling brick in a pile.

In the case of high temperature mortars, prepared ramming mixes, plastic refractories and similar materials shipped in packages, the provisions of A.S.T.M. specification C178<sup>6</sup> may be used. These provide for the selection at random of one container in any lot of 100 or less, two containers from lots of 100 to 300, three containers from lots of 300 to 600 and four containers from lots of 600 to 1,000.

The selection of the sample for any given test from these larger samples involves the use of judgment.

In the case of granular materials and fire clay the samples should be mixed carefully and then be divided by passing through a sample splitter or by quartering to secure the necessary, representative portions. Special attention must be given to the fine particles since errors in dividing this part of the sample are difficult to avoid.

In the case of fire brick, samples for each individual test should be selected to represent the range in appearance covered by the original sample. In other

words, the five samples that might be selected for determination of water absorption should contain one of the darkest colored brick, one of the lightest colored brick and the remainder should be representative of the balance of the sample. This method of selection is used because of the relatively small number of samples that can be tested economically and makes use of the obvious differences among the samples. If a larger number of samples can be tested, then the selection of the number for any individual determination can be made on the basis of chance.

In the case of plastic materials packed in containers, a representative sample should be taken by selecting small quantities from several locations in each of the containers involved to be representative of the entire lot. This quantity should be mixed together and then reduced by quartering to select the amounts necessary for conducting the various tests.

#### IV. Test Procedures for Fire Brick

In the development of standard test procedures for refractories Committee C-8 of the American Society for Testing Materials has been the outstanding leader. The Committee is so organized as to insure adequate and critical review by both users and producers of all of the matters adopted. The interest and support of the American Refractories Institute in this project has permitted development work to be carried out to supply data for committee consideration. The standard specifications and test methods, together with industrial surveys on requirements or refractories, have been gathered together in one volume known as the *Manual of A.S.T.M. Standards on Refractory Materials*<sup>7</sup> which is similar to the *FOUNDRY SAND TEST HANDBOOK*.<sup>8</sup>

A general description of these and other tests will be given and an interpretation of the test results will be made, but full details will be omitted since they are so readily available.

#### Refractoriness

Refractoriness is the one fundamental property required of materials to be used in furnace structures. Many refractories are used in outer walls right up to the limit in this respect, with the knowledge that the cooler portions of the walls will carry the superimposed load. Softening is encountered in service not only because of excessive temperatures on the hot face but also because conditions of use are changed so that the refractories are heated on more than one side or that they are backed up with too much insulation.

In most cases softening will show up as vitrification of the outer zones of the refractory or the displacement of joints in one portion of a wall. In piers softening can be distinguished only by comparison of measurements in thickness of the brick before and after use with due allowance made for normal variation in size of the original brick. It is necessary, in analyzing possibilities of shrinkage, to give proper allowance for the absorption of fluxes which will change the refractoriness of the original product.

Softening of refractories is not necessarily detrimental to service. Ladle brick prevent penetration of

metal into joints by softening enough to be compressed tightly into the joints. Certain brands of high heat duty brick give as good service as super duty brick under severe spalling conditions because the former soften and bloat on the hot face to form a resilient layer that is effective in absorbing strains due to sudden heating and cooling.

Softening is sometimes erroneously reported to be present when shrinkage occurs. Under many conditions shrinkage may be due to absorption of fluxes or to insufficient firing of the refractory. Both shrinkage and softening may cause opening up of joints. Of course, there is a possibility that what appears to be shrinkage is actually slagging of joints by fluxes.

The standard test for refractoriness is the determination of the pyrometric cone equivalent (P.C.E.). This test<sup>9</sup> determines the softening characteristics in comparison with standard cones, details of the test being covered by A.S.T.M. Procedure C24.<sup>10</sup> The test is cheap and quick to run and therefore is used widely. It divides refractories into groups for classification but at times has limitations in bringing out small differences.

In conducting the test the sample is ground and molded into a cone of standard dimensions. After drying, a number of unknown materials may be mounted in a plaque with the standard pyrometric cones as shown in Fig. 2. The plaque is heated at a

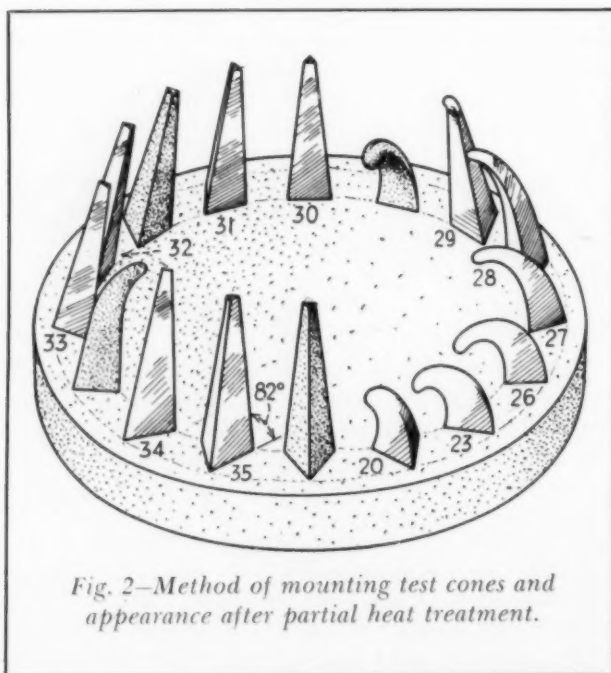


Fig. 2—Method of mounting test cones and appearance after partial heat treatment.

specified rate and a record made of the deformation of the unknown cones in relationship to the standards and are reported in that manner. For example, a material may be reported as having a pyrometric cone equivalent of 31, meaning that the material deformed at the same time as standard cone 31. If the unknown material deformed between two cones the higher and lower cones are reported in a hyphenated manner as, for example, P.C.E. 31-32. It should be noted that

there is a long temperature interval between cones 32 and 33, and, since many high grade clays fall in this range a cone is made with the designation 32 1/2. This is a definite cone in the series and is considered as such. Intermediate values are reported in the hyphenated manner as, for example, 32-32 1/2.

The load test had advantage in measuring refractoriness, particularly of siliceous clay brick, silica brick and basic brick as described later. It is limited in its application to clay brick of average composition since the degree of burn of such brick has an important influence on the deformation under load. Details of the test are covered by A.S.T.M. Standard Method C16.<sup>11</sup>

#### Permanent Linear Change after Reheating

The reheating of refractories was used originally to detect underburning. Some refractories may be checked for this characteristic in the reheat test. This is particularly true in the case of super duty brick. It should be pointed out, however, that a large percentage of satisfactory refractories show expansion instead of shrinkage in the laboratory test. This is known as secondary expansion and is practically without meaning from a service angle.

Details of the testing procedure for standard brick are given in A.S.T.M. Method C113.<sup>12</sup> Three 9-in. straight brick are measured for length to the nearest 0.02 in. and are reheated in a kiln, on edge. Low heat duty brick are heated to 2190 F (1200 C), intermediate heat duty brick to 2460 F (1350 C) and high heat duty brick to 2550 F (1400 C). These temperatures are reached in three hours and held for five hours after which the kiln is allowed to cool. Super duty brick are heated to 2910 F (1600 C) in five hours and held for five hours. The cool specimens are again measured for length and the individual per cent changes in length are calculated on the basis of the original data and then averaged.

Insulating fire brick are tested for permanent linear change in the same manner. The temperatures of reheat are set at 50 F (28 C) below the maximum temperature for service, e.g. 1550 F (845 C) for group 16 brick. The details are covered in A.S.T.M. Method C210.<sup>13</sup>

#### Strength

The strength of refractory products is determined at room temperature and is used as a measure of other characteristics as well as that of ability to withstand strains. Hunt and Bradley<sup>14</sup> have shown that the strength of refractory products increases with temperature up to 1500 F or higher, so that the values obtained at room temperature may be used for a major portion of the wall structure.

In recent years the modulus of rupture determination has gained favor over the use of cold crushing strengths. It is felt that failures which take place in refractories due to low strength are usually a cross breaking phenomenon rather than actual crushing. It should be noted, also, that the modulus of rupture determination falls to a low value with even minor defects in the physical structure of the product and

so is a sensitive indicator of physical defects. In addition, the resistance of refractories to abrasion follows rather closely the value for modulus of rupture. Manufacturers of refractories consider the modulus of rupture to be a reasonably accurate measure of the proper firing of refractories and of the ability to withstand rough handling in shipment.

The cold crushing strength is an equally reliable but somewhat different measure of strength of refractories than the modulus of rupture. The relationship between the two values varies with the process of manufacture and other characteristics of the brick so that there is no reliable way of estimating one value from the other. The cold crushing strength is used also as an indication of the proper burning of refractories.

Details of determining both modulus of rupture and cold crushing strength are covered by A.S.T.M. Method C133.<sup>15</sup> For cold crushing strength, 9-in. straight brick or pieces of that size cut from special shapes or tile are used in the test. (It should be noted that specimens cut from larger tile frequently do not have the same strength as 9-in. straight brick molded to size of the same mix and burn.) The specimens are placed on end in the testing machine after preparation of the load bearing surfaces and the pressure applied until crushing occurs. The crushing strength per square inch is then calculated on the basis of the original cross section. For determination of the modulus of rupture, the specimens are placed flat on knife edges spaced exactly 7 in. apart. Pressure is applied at the center of the upper face through a similar knife edge and the load applied until fracture occurs. The modulus of rupture is calculated according to the formula  $R = \frac{3Wl}{2bd^2}$

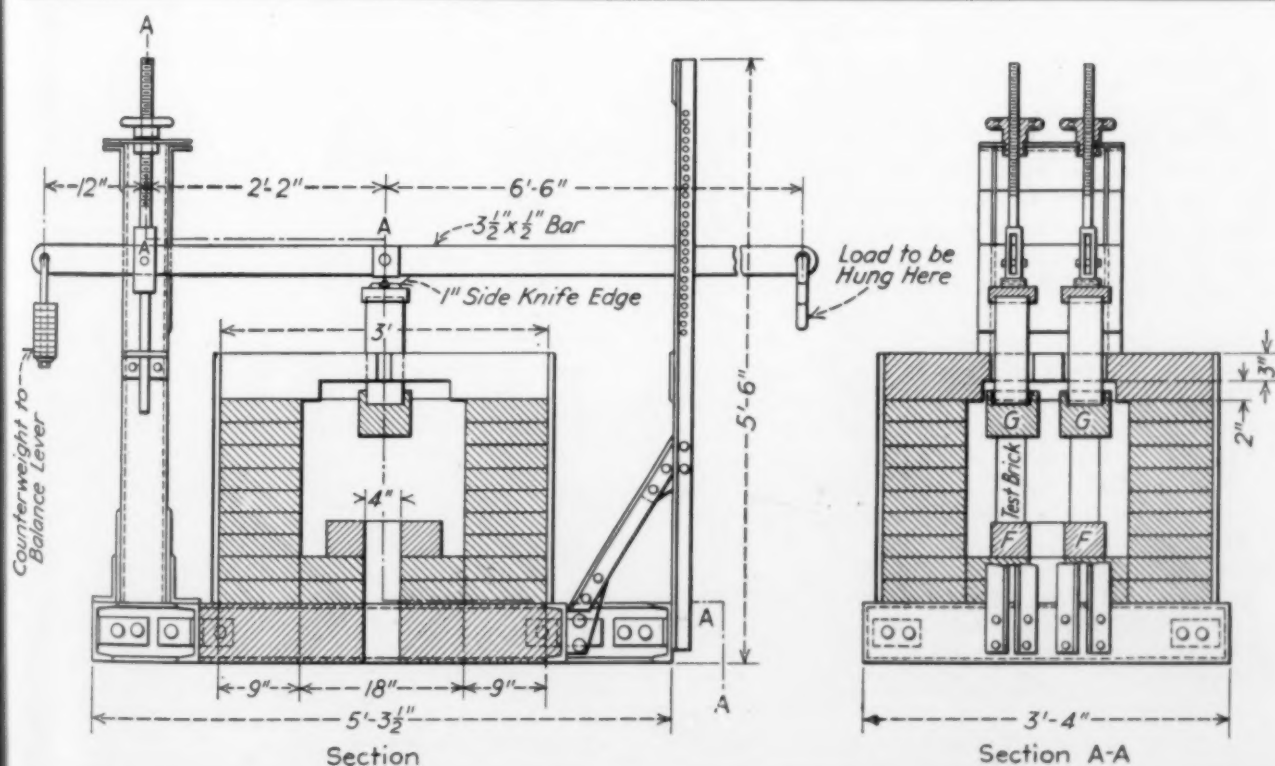
In testing insulating fire brick, special modifications must be made as covered by A.S.T.M. Method C93.<sup>16</sup> In determining cold crushing strength half-brick specimens of approximately  $4\frac{1}{2} \times 4\frac{1}{2}$  in. size are used and 10 specimens are preferred to secure reliable data. These specimens are crushed on their  $4\frac{1}{2} \times 4\frac{1}{2}$ -in face and the testing machine applies the load at a slower rate. In the case of the modulus of rupture determinations, the procedure is identical to that for standard refractories except for the slower rate of load application.

#### Load Test

The load test is used to measure the ability of refractories to carry a load when heated on all sides and, in addition, is employed to measure the refractoriness of some products such as siliceous fireclay brick, silica brick and basic brick. It may predict the service to be obtained in piers, division walls between checker chambers or boilers, and similar applications. In outer walls the load is carried by the cooler portion of the wall and so the load test does not predict service.

Details for conducting the load test are given in A.S.T.M. Method C16.<sup>17</sup> Nine-inch straight brick are measured accurately for length, width and thickness and then placed on end in a suitable furnace (see Fig. 3). Pressure is applied through silicon carbide or fused alumina brick to give a value of 25 psi. Intermediate heat duty brick are heated to 2370 F (1300 C), high heat duty brick to 2460 F (1350 C) and super duty brick to 2640 F (1450 C) in  $4\frac{1}{2}$  hours and

Fig. 3—Lever-type load test furnace.





these temperatures maintained for  $1\frac{1}{2}$  hours in addition. During the progress of the test valuable information may be obtained by making a graph of the expansion and deformation of the brick assembly to note the temperature at which deformation begins and the rate of deformation during the final stages of the test. When the brick are cold they are again measured for length and the per cent deformation calculated on the basis of the original dimension. Usually two brick are tested and the average is used in the report.

Under a new schedule silica brick are heated to 2700 F (1480 C) in 5 hours and then the temperature is continued at the rate of 100 F (56 C) per hour until failure occurs. Basic and neutral brick are tested on a similar schedule reaching 2460 F (1350 C) in  $4\frac{1}{2}$  hours and continuing at the rate of 100 F (56 C) until failure occurs. These values are measures of refractoriness rather than load bearing ability.

### Spalling

Spalling of refractories is divided into three classes according to the mechanism bringing about the loss of heavy pieces. These three types are thermal, structural and mechanical. A proper analysis of service conditions and the selection of the remedy depends upon the differentiation between the three types.

Thermal spalling is due to temperature shock alone. The sudden increase in temperature of one surface in comparison to the remaining refractory results in marked expansion and the strains which are set up cause rupture of the outer surface from the body of the refractory. Thermal spalling rarely occurs by itself but usually supplements the other two types, or a combination of them, to cause the loss of pieces. It is characterized by a rough fracture usually around the grains. (It should be noted that spalling is considered to be the loss of pieces  $\frac{1}{4}$ -in. or more in thickness and is distinguished thereby from shelling which is the loss of thin layers up to  $\frac{1}{4}$ -in.)

Structural spalling is brought about by a change in structure of the refractory. This change may be due to absorption of fluxes, to vitrification, to mineral changes or to loss of bonding material by heat. These actions either increase the rigidity, reduce the strength or cause shrinkage of the hot face of the refractory so that strains set up by this action and by thermal expansion will cause rupture. This type of spalling is characterized by a decided change in structure of the hot face. If the change is brought about by absorption of fluxes or by high temperature vitrification the fracture will usually be smooth and pass through the grains in this altered zone. However, there are cases where the fracture passes around the grains, just back of the altered zone, in which case the broken surface resembles that of thermal spalling.

Mechanical spalling is the result of mechanical strains. These may be brought about by pinching due to the use of insufficient expansion joints, to settling of foundations or other supporting members, or to similar mechanical causes. The action is identical to crushing at room temperature except that the shattering force may be a combination of mechanical

pressure on which is superimposed strains due to thermal expansion. This type of spalling is generally characterized by a rough, jagged surface and sometimes it is accompanied by the formation of splinters or loose grains of brick.

The A.S.T.M. panel spalling tests are the most reliable indications of the behavior of refractories in service or any of the known test procedures.<sup>18</sup> They involve a combination of reheating the surface of a

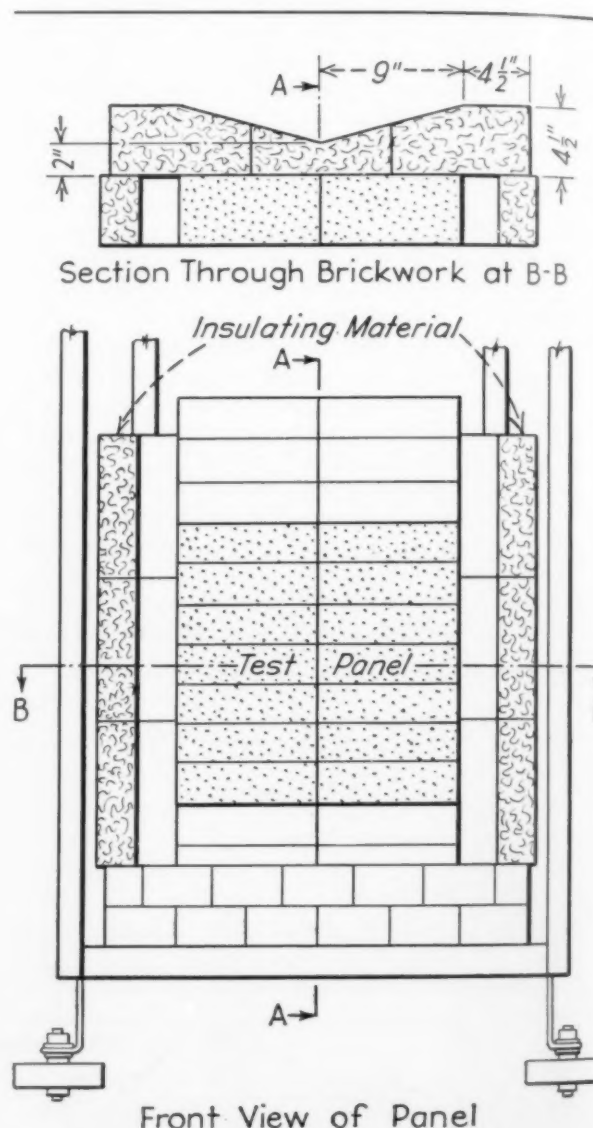


Fig. 4—Spalling panel frame and brickwork.

panel of brick (see Fig. 4) to secure the change in structure that may occur under high temperature service followed after cooling by a severe thermal shock treatment. (See Fig. 5.) After treatment the panel is dismantled, loose pieces are removed and the per cent loss in weight is calculated. The test measures structural and thermal spalling under set high temperature conditions.

Details of the spalling equipment are given in

A.S.T.M. Method C38<sup>19</sup> while the procedures for high heat duty and super duty brick are covered in A.S.T.M. Methods C107<sup>20</sup> and C122<sup>21</sup> respectively.

It should be pointed out that the spalling procedures involve a considerable amount of equipment and are expensive test routines. Because of this fact the spalling test is used by foundries only to investigate the most pressing problems or possibly to secure a yearly check on the character of refractories they are using. It is believed that the spalling tests conducted by the manufacturers of refractories can be relied upon as a check of that individual product and need not be repeated by the foundry as a routine test. In most cases where tests are needed it would appear advisable to have the test conducted by a commercial laboratory skilled in the work.

In those cases where some sort of a check on spalling characteristics is advisable, substitutes of moderate value may be used. These would include the routine determinations of porosity before and after reheat for the class of brick involved together with those of modulus of rupture. Constant values for these properties would suggest that the spalling values were remaining constant. Questions should be raised when there is a marked change.

Fig. 5—Spalling equipment for panel spalling test.

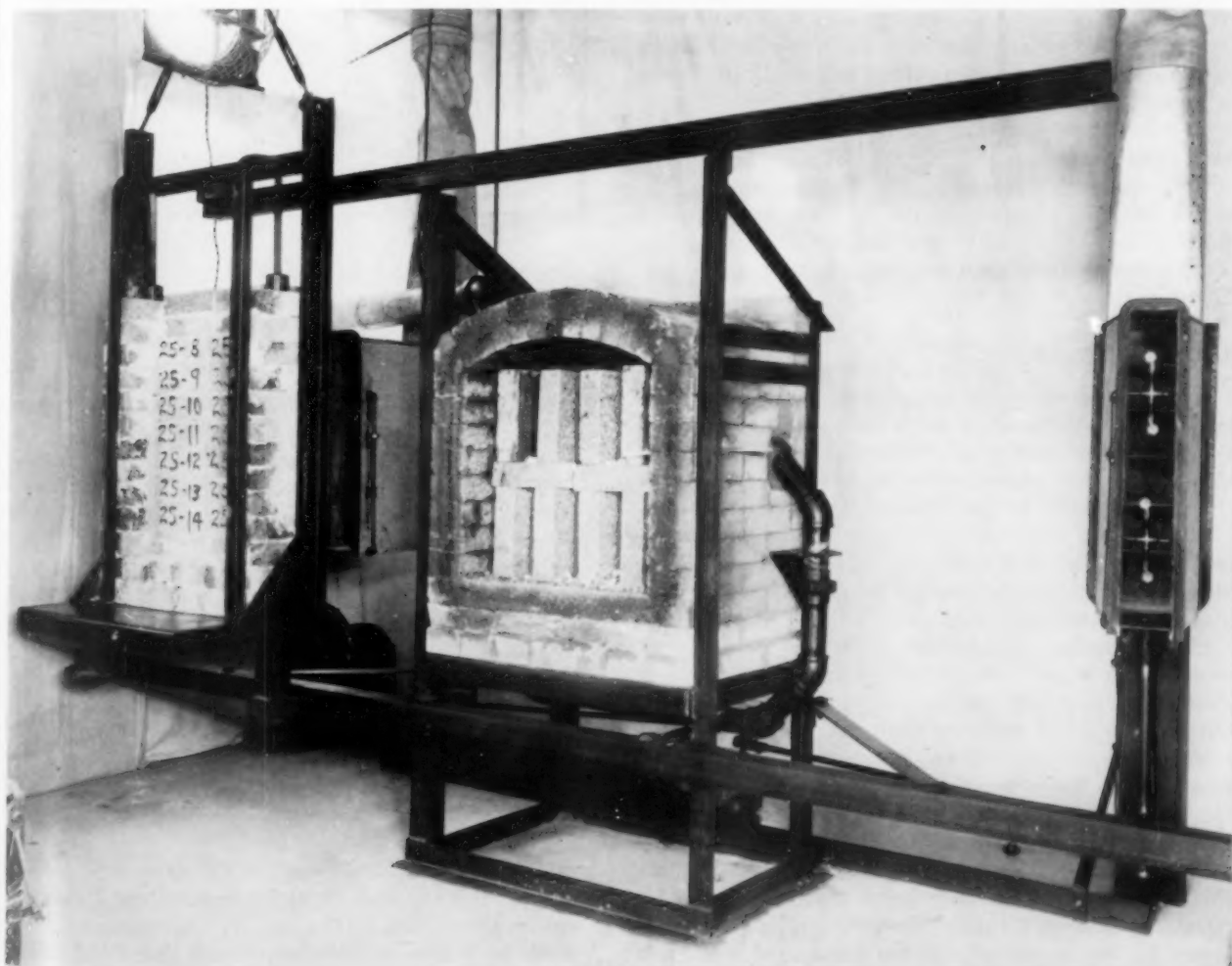
#### *Apparent Porosity, Water Absorption, Apparent Specific Gravity and Bulk Density*

Porosity and water absorption give valuable information concerning the physical structure of refractories. However, they have greatest use when considered in conjunction with other properties. Since porosity and water absorption are usually employed in connection with one type of refractory at a time either may be used for drawing conclusions. When different types of refractories are involved porosity is much to be preferred. This is due to the fact that porosity is a volume percentage of pore space while water absorption is a weight proportion and changes in specific gravity of the refractory alter the later values.

Similar results on porosity determinations are reasonably good indications of uniformity of properties of the refractories involved. Any variation in forming procedure, grain sizing, or degree of burn may show as variations in the porosity of the finished product.

In the case of high heat duty brick, moderately high porosity is usually associated with good spalling resistance while low porosity is an indication of resistance to slag attack. It should be noted, however, that in the case of super duty brick the porosity figures should not be used for such an interpretation.

Bulk density and apparent specific gravity figures have less significance on service behavior of refracto-



ries than does porosity. Of course, bulk density figures may be used in engineering calculations to denote the weight of refractories involved in any installation. In the case of silica brick apparent specific gravity data give an indication of degree of firing and of the thoroughness of conversion to tridymite and cristobalite. Normal products will fall in the range from 2.28 to 2.34.

The apparent porosity, water absorption, apparent specific gravity and bulk density data are determined on burned refractories by filling the pores with boiling water, in accordance with A.S.T.M. Method C20.<sup>22</sup> Five specimens consisting of approximately a quarter of a 9-in. straight each are dried and weighed. They are then placed in water and boiled for two hours. After cooling, the suspended and saturated weights are determined. From the three weight figures the four values of apparent porosity, water absorption, apparent specific gravity and bulk density are calculated and then the values are averaged for the five pieces.

#### Chemical Analysis

The chemical analysis of unused refractories is useful in dividing them into classes but does not predict the service that may be obtained. This is due to the fact that refractories are structural materials and their physical properties are more important than their chemical ones provided they are in the proper classification. Exception to this general statement may be had in the case of silica brick and basic brick since with these products the analysis may indicate types of important impurities.

Chemical analyses of used brick are valuable in indicating the relative importance of absorbed fluxes and the characteristics of slagging constituents. In this case it not only gives the nature of the contaminating material but the relative rate and seriousness of this contamination. Methods for analyzing refractories are found in the A.S.T.M. methods under C18.<sup>23</sup>

#### Size and Warpage

The sizes of brick, tile and shapes are important in permitting their erection with the proper size joints. With the development of modern refractory mortars this point is not as important as formerly. However, the matter should be given proper attention.

Methods for determining the size of refractory brick are covered by A.S.T.M. Procedure C134.<sup>24</sup> This method covers procedures for 9-in. brick and series and for tile and large shapes.

Determination of warpage is usually made only when there is an indication of serious defects in a shipment. The general condition of freedom from warpage is made quite evident by pressing several brick or shapes together and noting the relative tightness of fit. This practical method permits the inspection of large numbers of brick and tile with a minimum of effort. If appreciable warpage is indicated by this inspection the individual pieces may then be measured in accordance with the A.S.T.M. standard method C154.<sup>25</sup> Twenty specimens are selected for measurement and the concave side of each

determined. A straight edge is then placed diagonally from corner to corner as shown in Fig. 6 and the amount of clearance between the edge and the tile

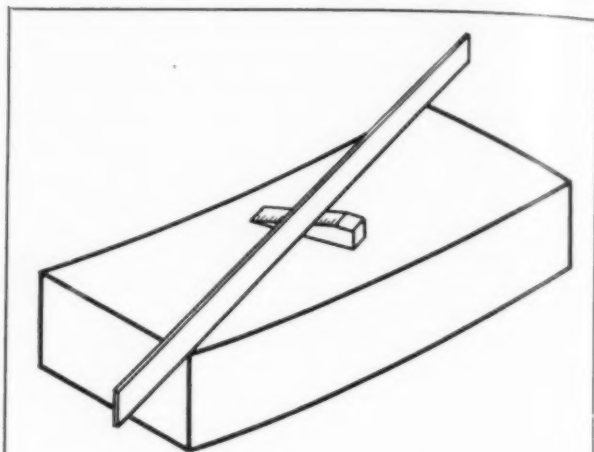


Fig. 6—Method of measuring warpage.

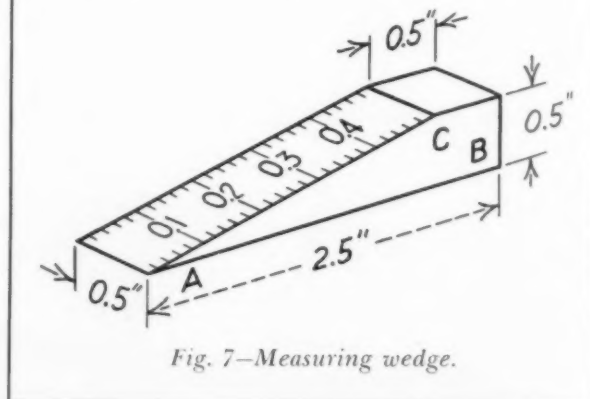


Fig. 7—Measuring wedge.

measured at its greatest value with a calibrated wedge illustrated in Fig. 7.

#### Thermal Conductivity

The thermal conductivity of standard refractories is of practical importance only in checking unknown materials. The value does not vary greatly from one brick to another in a given brand or class. In the case of insulating fire brick this factor is of greater importance. Methods for its determination are covered by A.S.T.M. Procedures C201,<sup>26</sup> C182<sup>27</sup> and C202.<sup>28</sup> A sketch of the apparatus is shown in Fig. 8. In view of the small amount of variation from sample to sample and the expense of the determination, it would appear advisable for most foundries to use data supplied by the manufacturer or to have the tests made by a commercial laboratory possessing the equipment.

#### Slag Test

There is no suitable slag test for refractories because of the large number of variables involved. The difficulties are shown by the surveys of proposed methods made by Ferguson,<sup>29</sup> Simpson,<sup>30</sup> and Hurst and Read.<sup>31</sup>



A qualitative estimate of slag resistance may be made from values of water absorption, the lower the absorption the greater will be the resistance to attack, in any given class of brick.

#### Abrasion

There is no accepted abrasion test. Under general conditions the modulus of rupture of brick may be used to estimate resistance in this respect.

### V. Test Procedures for Fire Clays and Mortars

#### Refractoriness

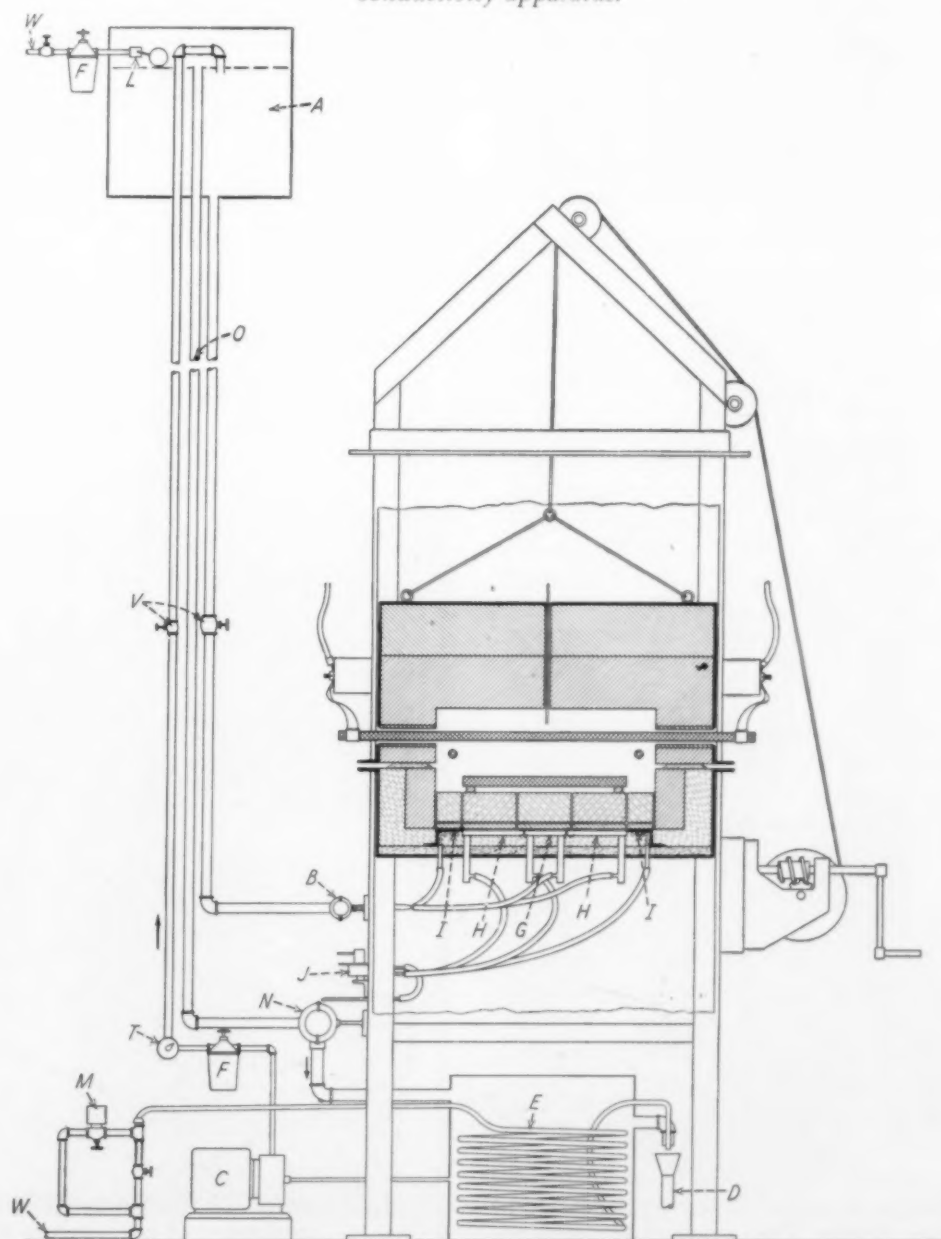
The American Foundrymen's Association has done considerable work on the testing of fire clays as they

relate to use with foundry sands. Some of these procedures can be used to advantage for clays used for mortars. Recent work has shown that distinctly different procedures are required to evaluate correctly the properties of high temperature mortars.

The refractoriness of fire clays may be determined by the P.C.E. test which has been described already.<sup>32</sup> The same type of determination may also be used for heat-setting refractory mortars.

In the case of air-setting refractory mortars the P.C.E. test is unreliable and a different procedure is required involving the building and reheating of a pier of three brick separated by mortar joints. This procedure, which is described in detail in A.S.T.M. Method C-199,<sup>33</sup> provides for construction of the pier

Fig. 8—Diagram showing essential parts of thermal conductivity apparatus.



as shown in Fig. 9 with one vertical and two horizontal joints all  $\frac{3}{32}$  in. thick and resembling a portion of a bonded wall. Each type of mortar is used with brick of the corresponding class with which it

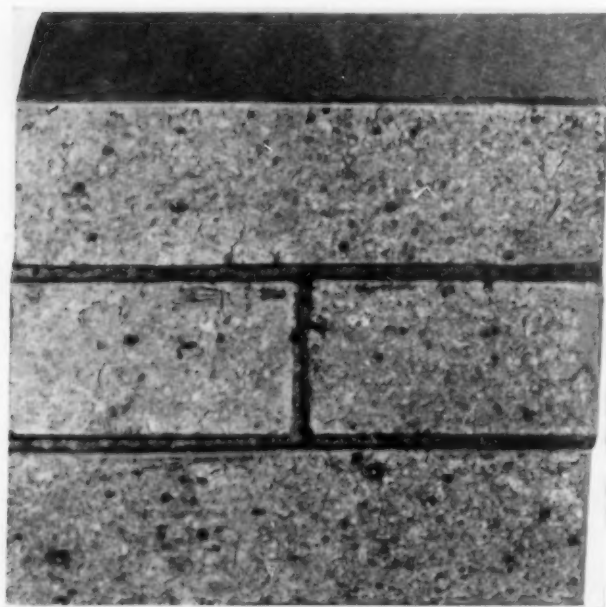


Fig. 9—Panel for refractoriness test of air-setting refractory mortar.

will be used, that is, super duty, high heat duty and intermediate heat duty. The dried specimen is placed in a furnace and fired to a temperature corresponding to its class. Intermediate heat duty materials are heated on Schedule B of A.S.T.M. Method C113<sup>34</sup> to 2550 F (1400 C), high heat duty on Schedule C except the temperature is held at 2730 F (1500 C) and super duty on Schedule C to 2910 F (1600 C). On all schedules the temperatures are held for five hours before cooling. The specimen is then inspected and satisfactory materials must show no flowing of the mortar from the joints.

#### Screen Analysis

Screen analyses may be conducted by two basic methods, either wet or dry. The wet method permits clay to slake and thus simulates conditions under which clay would be used for mortar. On non-slaking ingredients the wet method gives a more thorough analysis. On the other hand, materials which are to be used as packing or ramming mixes in a dry condition are tested under the dry procedure in order to give a true indication of the condition as ground and to avoid possibilities of breakdown of grains which might occur with the wet method. Details of these procedures are given in A.S.T.M. Method C92.<sup>35</sup>

#### Bonding Strength

One of the valuable characteristics of air-setting refractory mortars is the ability to develop high bonding strength upon drying. This property is measured by forming a joint  $\frac{1}{16}$  in. thick between the original

ends of 9-in. straight brick cut in half. A.S.T.M. Method C198<sup>36</sup> gives details for forming five such specimens which are dried carefully and, when cool, tested for modulus of rupture by breaking at the joint. Calculations are made in accordance with the same procedure as used in C133<sup>37</sup> and the average of five determinations are reported.

#### Consistency

Since the properties of mortars, particularly bonding characteristics, vary with the consistency of the mortar, it has been recognized for some time that it would be advisable to bring all mortars to a standard value before testing. Attempts to produce equipment for making such measurements have not been entirely successful because of difficulties of handling all types of mortar. However, one instrument may be used on a tentative basis until a standard is adopted.

Petrie and Kocher<sup>38</sup> have suggested the use of a modified Vicat needle 19.0 mm. in diameter, 44.4 mm. long, weighing 52.3 grams. On the basis of their work a penetration of 35 mm. could be considered standard for testing air-setting refractory mortars. Figure 10 shows one form of this apparatus. Mortars would be

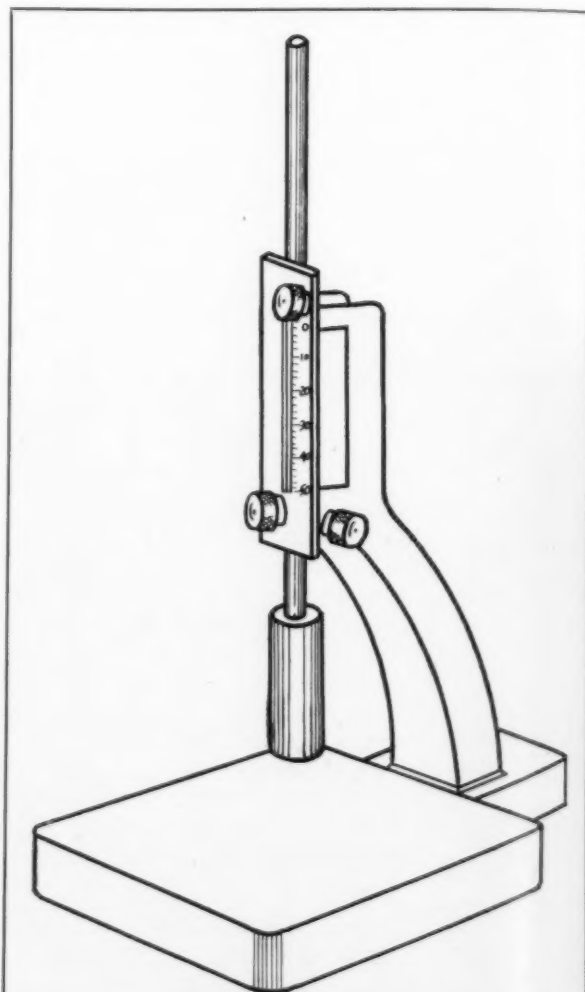


Fig. 10—Modified Vicat apparatus for measuring consistency of refractory mortars.

A.S.T.M. ve such en cool, ne joint, e same of five

brought to this consistency by addition of water before testing for other properties.

### Water Retention

Work on high temperature mortars has shown a need for a measure of the property of remaining workable in contact with fire brick. The Refractories Fellowship at Mellon Institute has suggested a water retention test<sup>30</sup> which is in process of being perfected. Even though it has limitations at present, it may be used to advantage in foundry testing.

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Early work indicated that silica brick were more uniform in their characteristic of absorbing moisture from mortar than was the case with fireclay brick. Accordingly, silica brick are cut in half forming  $4\frac{1}{2}$  by  $4\frac{1}{2}$ -in. pieces. A generous amount of mortar is placed on the end of a half-brick. A second half is placed on top of the mortar, squeezing it down to  $\frac{1}{4}$ -in. in thickness by use of spacer rods and the excess mortar is removed. At a time estimated to be the proper one for the particular mortar, the spacer rods are removed and a weight to give a pressure of 2 psi is placed on top of the specimen. (See Fig. 11.) If the mortar squeezes out of the joint the test is repeated after a longer holding period. If the mortar does not squeeze out it is repeated for a shorter period. The procedure is continued until a time is found at which the mortar will just squeeze out of the joints. Poor mortars show values as low as 10 sec. while satisfactory ones give values of 2 to 10 min. or more.

## VI. Test Procedures for Plastic Refractories

### Refractoriness

The refractoriness of plastic refractories may be determined by the P.C.E. test.

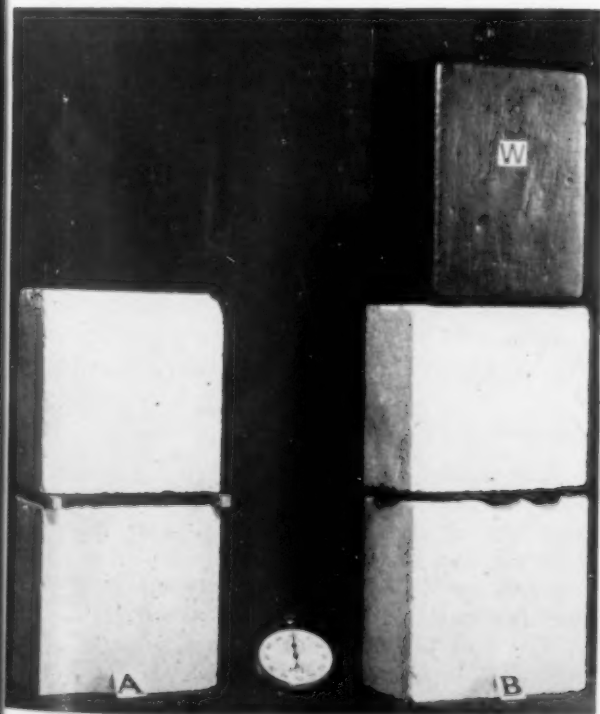


Fig. 11—Arrangement for measuring water retention.

### Spalling

The spalling of plastic refractories is carried out on 9-in. straight brick molded from the plastic in a press as described in A.S.T.M. Procedure C180.<sup>40</sup> The panel is laid up of dry brick using kaolin mortar and tested exactly as would be the case with fireclay brick. A separate sample of the plastic refractory is tested for loss on ignition and this value used to correct for the loss in weight of test brick during reheating cycle.

### Workability Index

The refractories industry has adopted the sand rammer of the American Foundrymen's Association for measuring the workability index of plastic refractories. The instrument is modified only by attaching a scale alongside the top of the plunger rod to measure distances and installing hooks for holding the weight during part of the operation as shown in Fig. 12.

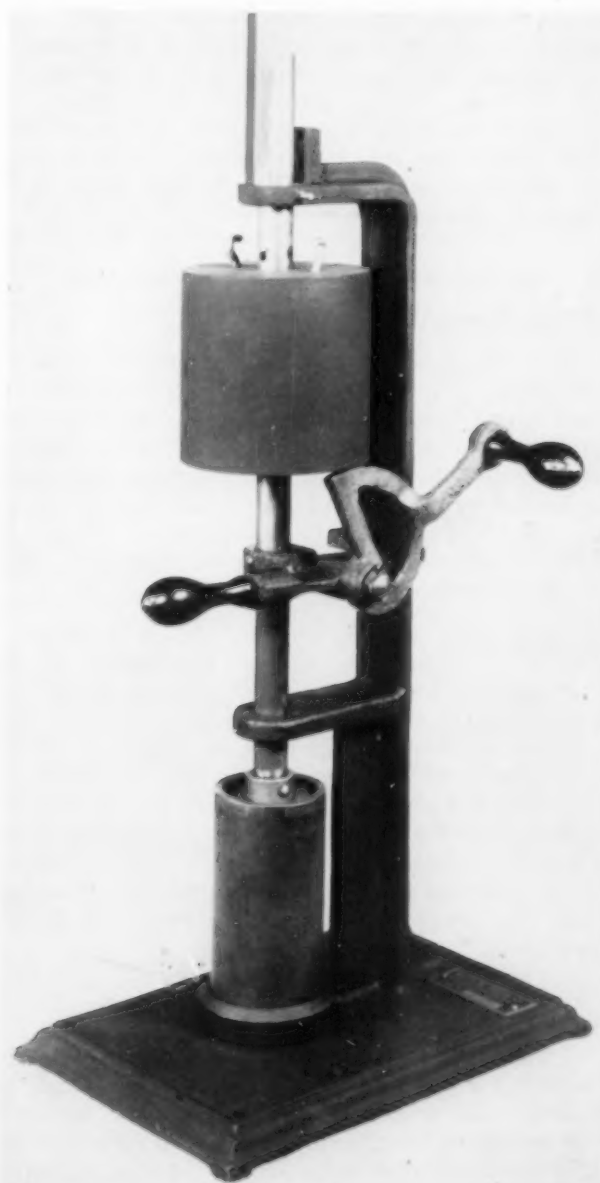


Fig. 12—Apparatus for workability index test.



A.S.T.M. Method C181<sup>11</sup> provides that specimens be formed from 300-gram samples by tamping for 10 impacts each from top and bottom in the standard holder. The plastic specimens are then removed from the holder and deformed by three impacts of the equipment. Measurements before and after deformation give a percentage which is averaged for four specimens in the report. Satisfactory workability is considered to fall between 15 and 35 per cent subsidence. If the test specimens crack during the deformation the plastic is considered to have failed in the test.

#### *Drying and Firing Shrinkage*

The combined drying and firing shrinkage of plastic refractories should show a low value. A.S.T.M. Method C179<sup>12</sup> provides for the forming of 9-in. straight brick from the plastic under a pressure of 1000 psi. After the lengths are measured with a hooked rule three specimens are dried and fired in a kiln to 2550 F (1400 C) in the case of high heat duty plastic and 2910 F (1600 C) in the case of super duty plastic. The temperatures are held for 5 hours before the kiln is cooled. The measurement of the length of the cooled specimens permits calculation of the combined drying and firing shrinkage expressed on the basis of the original plastic length. The average of the three values is reported.

### VII. Test Procedures for Castable Refractories

#### *Reheat Strength*

The new A.S.T.M. specification C213<sup>13</sup> for castable refractories provides that test specimens shall show a modulus of rupture of at least 100 psi. both after oven drying and after firing to 2000 F (1095 C). Details are given for the casting of 9-in. straight brick, four of which are oven dried and then tested for modulus of rupture while four additional samples are reheated to 2000 F (1095 C) in 3 hours, held for 5 hours and then tested for modulus of rupture after cooling.

#### *Refractoriness*

It has been found that the refractoriness of castable refractories may be determined by measuring the permanent linear change after reheating. A.S.T.M. Specification C213<sup>13</sup> provides that 9-in. straight brick made from Class 24 castables shall be reheated to 2250 F (1230 C) for 5 hours and that specimens of the Class 27 castable shall be reheated to 2550 F (1400 C) for a similar period. Satisfactory materials show less than 1½ per cent expansion or contraction in this procedure.

### VIII. Testing Programs

Three types of testing programs may be considered suitable for use in foundries. The first type might be called "rule of thumb" observation. The second is the testing of basic properties without regard to use. The third and most valuable is the testing of properties required by the service conditions and so is based on a critical survey of conditions to be met.

#### *Rule of Thumb Observation*

In general, "rule of thumb" testing is not even considered in a technical presentation. However, it has the merits of rapidity and cheapness and should not be passed by lightly.

In the case of brick, this testing would start with striking them together or with a small hammer to note the ring. A clear note indicates a satisfactory bond and soundness of structure. At the same time, the edges of the brick can be rubbed together and the relative resistance to abrasion observed. Several brick may be broken and a note made of the grain sizing, the bond and the soundness of structure. The brick may be measured with a rule, if care is taken to place one end exactly at the end of a brick. In a pile the brick may be pushed together so that their relative straightness is evident.

In the case of fire clay, dry mortars or castable refractories, the material may be worked between the fingers to observe grain size. This may be carried farther by mixing a small amount with water in the hands and observing the workability and feel.

In the case of wet mortar, the material can be worked under a trowel to note the consistency, the smoothness and possibly the grain sizing.

In the case of plastic refractories a piece can be broken and kneaded in the hands to note workability and freedom from tendencies to crack as well as maximum grain size.

Since every testing program must be limited in some degree, these "rule of thumb" methods can be relied upon for checking some lots while others are tested in the laboratory. In such a case it would be advisable to make out a simple form with spaces for the proper comments and make a general rating of the properties observed, the record of the shipment and an approximate rating of the material on the basis of these data. A file of such information would be valuable to indicate the relative uniformity of the products received.

This type of testing has the advantages of being very quick and inexpensive to conduct since it requires a minimum of apparatus and time. Due to this simplicity a large number of samples can be observed. The results may indicate uniformity in general properties of the products.

The process has the disadvantages that it gives no numerical values and that the ratings will vary decidedly not only with individuals but also with different times of testing by the same individuals. The results can be connected only vaguely to service needs.

#### *Laboratory Testing for Basic Properties*

The testing of basic properties of refractories is the first step forward in a true testing program. Only a very few of these basic properties are determined, however, and the plan is considered a stop-gap until the next steps can be taken, or a fill-in between true specification tests.

The tests to be conducted vary with the class of material. In the case of fire brick the P.C.E. would be determined to check refractoriness, the modulus of rupture would be measured to indicate adequate strength, soundness and burn, and the water absorp-

tion would be checked to reflect the general structure of the product. The tests for fire clay dry mortars and plastic refractories would involve screening through one coarse sieve and determination of P.C.E. In the case of wet mortars and castable refractories, a limited screen analysis would be the sole test.

The data should be recorded and plotted on a graph against time. By this means it is easy to note the uniformity in basic properties of the products received. However, since the properties are not yet definitely correlated with service the average values may have a large factor of safety and low results should not be used as a basis for rejection of shipments. In observing the graph consideration should be given to the findings of Phelps<sup>44</sup> on the reproducibility of test data.

This method has several advantages, one of which is the obtaining of definitely reproducible data. The results will show uniformity or lack of uniformity in the material. Since the tests are restricted they can be conducted in a reasonable time and at only a moderate expense.

The chief disadvantage of this type of testing is the lack of correlation with service. The importance of this limitation is very often overlooked. The fault

lies in a mistake in simple logic. In order to obtain satisfactory results it is not necessary to duplicate a good product in all respects, but rather only in the necessary properties. The fault is so common that the following example is given.

A cupola block that had been found to be satisfactory on the basis of actual use in a gray iron foundry was tested in all the standard procedures except spalling and the complete results were set up as those required for satisfactory service. All future shipments of various brands were tested fully at high cost and if they did not conform to these specifications, were used only in the upper part of the cupola. Usually the reason for rejection was low P.C.E.

At first the setting aside of certain shipments for less severe use was pointed out with pride as justification for the testing program. Later, however, some of the blocks of low P.C.E. were used by mistake and gave good service. Also, some blocks of a rejected brand were used with outstanding success by a neighboring foundry operating on a similar melting rate. As a result, pointed questions were raised, and in view of the high cost and the failure to secure the results boasted about, the entire program was discontinued.

TABLE I—TYPES OF FIRECLAY REFRACTORIES

Tabulation of the properties required of each type and the test procedure involved.

A.S.T.M. type of brick	Class Heat Duty	Type of Service	Refractoriness evaluation		Shrinkage in permanent linear change test Maximum	Panel spalling maximum loss, Temperature of preheat given	Modulus of Rupture minimum	Water Absorp- tion max.
			P.C.E. Min.	Load Test (Cone) Deformation Maximum				
A.S.T.M. test	C 27		C 24	C 16 (2460° F.)	C 113	C 107 (2910° F.) C 122 (3000° F.)	C 133	C 20
Type D	Super	Extremely severe spalling or slagging.....	33	—	—1.0% (2910° F.)	4% (3000° F.)	—	—
Type I	Spec'l	Bungs malleable iron furnaces.....	32	—	—	7.5% (2910° F.)	200 psi.	—
Type A	High	Average conditions, boilers and incinerators.....	31-32 or	1.5%	—1.5% (2550° F.)	15% (2910° F.)	—	—
Type B	High	Moderately severe spalling — boilers.....	31-32 or	1.5%	—1.5% (2550° F.)	10% (2910° F.)	—	—
Type C	High	Moderately severe slagging — boilers.....	31-32 or	— 1.5%	—1.5% (2550° F.) —1.5% (2550° F.)	— —	— —	10% 18%
Type E	High	Division walls — boilers.....	31-32 and no P.C.E. and 1.5%	6.0% or	—	—	—	—
Type J	High	Malleable side walls.....	32	—	—1.5% (2550° F.)	—	—	—
Type K	High	Fire boxes and chimneys malleable iron furnaces.....	31-32 or	1.5%	—1.5% (2550° F.)	—	—	—
Type F	Inter.	Intermediate temperatures— boilers and incinerators.....	29	or 3.0%	—	—	500 psi.	—
Type G	Low	Low temperatures — boilers.....	19	—	—	—	500 psi.	—
Type L	Low	Annealing ovens — malleable castings.....	19	—	—1.5% (2190° F.)	—	—	—

NOTE: Each type of refractory must comply with the following:

SIZE VARIATION—9" brick not more than  $\pm 2\%$ . Tile or shapes  $\pm 2\%$  on dimensions of 4" or over, and  $\pm 3\%$  on dimensions smaller than 4". Measurements must be made by A.S.T.M. method C 134.

WARPAGE—95% of the shapes shall not show warpage greater than 1% of the diagonal used in making the measurements. Test method A.S.T.M. C 154.

In this case the worst error occurred from placing equal value on all properties rather than classifying them into groups as very important, moderately important or not important on the basis of a study of service conditions to be met and then drawing up limits on the basis of these data. It was not recognized that the high P.C.E. value of the blocks tested originally was unnecessary and that a lower value could be used. A study of many more than one product must be made before the service requirements can be rated and a specification drawn up.

#### *Testing for Service Properties*

To obtain maximum benefit from a test procedure the properties determined should be only those that are required for the service conditions to be met by the refractories. This involves a lengthy and detailed study of the service requirements for the type of equipment in general and then a reevaluation on the basis of conditions in a particular plant and furnace. Findings of scientific groups such as Committee C-8 of A.S.T.M. are particularly valuable in starting such a study. Table I lists the types of refractories given in A.S.T.M. specifications and the test limits for each. The suppliers of refractories can often give valuable information and will supply advice and counsel during the progress of testing.

As an example of this type of preliminary evaluation, the problem might be stated as the selection of super duty brick for service in the sidewalls of a malleable iron furnace used in a duplexing process with cupola melting. While the A.S.T.M. Standard Specification C63<sup>45</sup> for this type of service has not been revised to cover such a product, it is pointed out that the particular type of brick is covered in the Specification for Heavy Duty Boiler Service C64.<sup>46</sup> In this specification type D brick shall show a P.C.E. not lower than Cone 33, a change in the super duty reheat test of not more than 1 per cent shrinkage and a loss in the super duty spalling test of not more than 4 per cent. This could be used on a tentative basis and various brands of brick could be tested. Those that met these requirements could be considered as satisfactory for further study. Carefully conducted panel tests in the production furnaces might then be carried out and calculations made to determine correlation between the service results and the test data. Use could be made of statistical formulas in the A.S.T.M. *Manual on Presentation of Data*<sup>47</sup> to assist in deciding if there is a correlation. The problem could be carried to a tentative conclusion in this manner.

After setting up this tentative specification tests would be conducted on regular shipments for the properties deemed important. The results would be recorded, filed and plotted on graphs as obtained. The tentative limits set by the specification would be placed on the graph in distinctive colored ink to be readily apparent.

Whenever a lot is tested and shows exceptionally good results in one or more properties the product should be watched in service to justify the selection of that property as important. If it gives better than

the usual service that fact is more evidence that the previous conclusions were correct.

The data will be more valuable if two panels can be installed, one from a lot of refractories with average properties and the other from a lot with superior properties. Inspection during and after service will show the relative difference in behavior. New brick from the two lots can then be reversed in position on the next campaign and similar observations made. This procedure will eliminate variables in operation and in conditions due to position in the furnace.

Similarly when a lot gives poor test results in any important property it should be given a service trial to assist in locating the lower limit of the specification. Panel tests are necessary in this case since the first trial should be in locations of only moderately severe conditions. If, at the end of the run, both panels (poor and average lots) are in good condition they may be left for additional service. This is advantageous in bringing out the differences gradually and making sound comparisons. In such cases of good results on both, new panels may be installed in the zone of most severe service. If the results are fairly good another run should be made with the refractories reversed in position, in order to secure fully accurate results.

If the poor lot shows very poor service the specification limit has been justified. Service results that are fair would point to a slight lowering in limits to avoid excessive costs in securing satisfactory refractories. Good results would indicate that a definite revision of the earlier conclusions and limits should be made.

By use of these studies for correlating test results with service behavior a sound testing procedure and specification can be developed that will select satisfactory refractories, will eliminate unsatisfactory ones and will point to superior ones.

At intervals of a year or less each specification and procedure should be reviewed to see if changes in service conditions have progressed far enough to alter the conclusions. If there is any question a short study should be made to check the former results or set new values or procedures.

The only disadvantage of this method of testing is the expense involved. Usually adjustments and compromises must be made to use as much of this type as can be justified.

#### **IX. Recommendations**

It is recommended that every foundry lay out a balanced program of testing on the basis of the considerations given above. The type of work must be set to conform to the size of the foundry and the quantities of materials being used. It is strongly recommended that the program start slowly and that expansion be made on a very gradual basis. In order to make greatest use of the testing program it is suggested that in every size foundry extensive testing be done on intermediate shipments. Thus the maximum benefit can be obtained with a minimum expenditure of money.

Several large users of refractories have adopted the program of testing a given brand of product for a



rather extended list of properties at the start of their consideration. After approval of the brand they follow a cycle of testing of approximately ten lots, checking some of these to a very limited extent and then following with testing on the tenth lot in a more extended manner. If test results are satisfactory, the program continues through repeated cycles. If one lot fails, the next lots go through the extended series of tests until three consecutive lots have been satisfactory, after which the program reverts to the cycle of ten tests.

On the basis of this plan it is recommended that a small foundry check a new product by inspecting test data supplied by the manufacturer and testing themselves as many of the service properties as they can afford. Due to the limited number of shipments received they would then follow a cycle of five lots checking four by "rule of thumb" inspection only and submitting the fifth to laboratory tests of as extended nature as their budget and purchases will permit. Records should be kept of even the "rule of thumb" inspection tests so that they can build up a file for reference when unusually good or unusually poor results are obtained. In this way they are not passing up any material without checking but on the other hand are not spending more money than can be afforded by the size of the operation.

For the large foundry, a similar plan can be followed over a cycle of ten lots. In this case again the preliminary information can come from the supplier of the refractories and his results can be checked by laboratory testing by the foundry. In this case the tests on each tenth lot should include all of the service properties that are necessary for the particular use. Each fifth lot between should cover the more important test properties that can be checked for a limited cost and the four lots in an intermediate position can be tested by the "rule of thumb" inspection methods. While this procedure is more expensive than for the small foundry, it can be justified on the basis of the more positive testing of materials.

Both large and small foundries must spend much of their total effort in correlating test results with service results. Since this is the most important part of the program it must not be slighted. The securing of even limited results is worth the entire effort.

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## DISCUSSION

Chairman: R. H. STONE, Vesuvius Crucible Co., Swissvale, Pittsburgh

Co-Chairman: A. S. KLOPF, Western Foundry Co., Chicago  
Dr. V. PASCHKIS:<sup>1</sup> The paper does not present tests for specific heat of refractories.

MR. SWAIN: Since the properties of specific heat and thermal conductivity of any given class of refractories fall in a rather narrow range there is little reason for the foundrymen to check these properties on repeated shipments but rather to test for the other properties that are likely to vary and that have a major influence on the service to be obtained.

For original work the properties of thermal conductivity and specific heat must be given consideration. In such a study the values may be obtained from the manufacturer and used with satisfactory results.

In the preparation of this paper an attempt was made to prepare a table showing the most important properties of re-

<sup>1</sup> Columbia University, New York

fractories that should be measured for various foundry uses. It was soon evident that this was too large an undertaking for this paper and might be the subject of a separate paper or series of papers for future meetings.

One reason why it is such a difficult problem is that service conditions vary from plant to plant and the group of properties that may be important in the furnace in one plant or one district may be quite different from those of another. Each plant has to work out its own specific requirements, try to select refractories to meet those requirements, and make the selection of the tests on that basis.

DR. H. RIES:<sup>2</sup> I wish to ask Mr. Swain how the cones for the refractoriness are made.

MR. SWAIN: The standard method calls for grinding the refractory to pass a 65-mesh sieve, bonding with an organic binder that does not contain fluxes, forming to a standard size, drying and mounting in an inert refractory plastic material for testing.

DR. RIES: The reason I asked that question was because a number of years ago I was testing the refractoriness of some New Jersey fire brick. According to the alumina and silica ratio shown by the chemical analysis, those brick should have fused at about 27. When we tested some samples of the cones cut from the bricks, they fused at cone 31, but when we ground the brick sufficiently fine to pass a 100-mesh sieve, and then formed this into cones the fusing point dropped to cone 27. If you ground your material finer than 65-mesh it might show a lower fusion point.

MR. SWAIN: There are some technical points in preparing cones with plastic clays. Such material should be calcined before the cone is made to avoid bloating of the cone during the final heating. In the case of materials containing fluxes and highly refractory materials, the operator has to be extremely careful to get thorough mixing and to avoid loss of materials or he will secure wide variation in results.

CHAIRMAN STONE: Is it ever the practice or might it be the practice to cut the cone from the refractory itself, or might that not be a supplemental test? I understand that the refractory engineers definitely are determined on the ground and reformed cone as the standard test. I wonder if that might not be a valuable supplemental test.

DR. UNGER who had charge of refractories for many years at the Carnegie-Illinois Steel Corp. held out for the cone cut from the refractory itself. Of course, it is difficult to cut to the size of the standard cone.

MR. SWAIN: Use of a cone cut from the brick and ground to size is the English standard method of determining refractoriness. In this country, we have standardized on grinding the material and forming the cone. The method has been used for so many years and there is such a background of information available on that method that it would be confusing to change. There are arguments in favor of both methods. The difficulties of grinding rather friable, soft material to the size and shape of a test cone are so great that most operators prefer to grind the material to 65 mesh and mold cones thus keeping away from such troubles. The size of the test cone has a material influence on the temperature of deformation, so that it would be necessary to grind any sample to a given size to be sure of consistent, reproducible results.

C. E. BALES:<sup>3</sup> Mr. Swain has given us a very interesting paper on the testing of refractories and I think it might well serve as a chapter in a handbook on foundry refractories, which we should have. Unfortunately, the tests Mr. Swain has described are really comparative tests rather than tests which indicate how a cupola block, a fire brick, a plastic refractory or a refractory cement will hold up in actual service. I say it is unfortunate, but it is the best that the refractories engineers have been able to do over a period of 25 or 30 years.

One of the best tests which the author described is the so-called panel spalling test. In the old days, we would heat indi-

vidual brick to a certain temperature and then dip the heated end in water. That test was in vogue for a good many years until finally somebody found out that the test did not measure the performance of a refractory brick in service.

In the development of the panel spalling test, it was found that it did predict how a bung brick would perform in service and how a boiler furnace brick would perform. The results obtained on those brick in this panel spalling test correlated very closely with the way they performed in service. That test has been widely used in improving the spall resistance of bung brick over the past 10 or 15 years. Now, the question is, could the same spalling test be used in improving cupola blocks? The service conditions in a cupola, as all foundrymen know, call for resistance to abrasion, resistance to high temperatures, resistance to slag erosion. A brick that is to be most resistant to abrasion may have poor resistance to spalling. One problem that refractories people are up against, is to develop a cupola block, that will be well suited for service in the ablation zone and at the same time give good service down in the melting zone. This is true especially in some foundries where they drop the bottom at the end of the day and direct a stream of water up on the white hot brick. I realize that in some foundries it is necessary to do this because they have to get the cupola patched by the time they are ready to operate the next morning. So far as spalling goes, that is very severe on the melting zone lining.

MR. SWAIN described the load test and he said that it has no value in a cupola. It does have in an open hearth furnace, where the steel foundrymen melt a great deal of their steel. It is valuable in checker brick, because those brick are heated all around, just as the specimen is heated in the load test. The load test might be used to predict service of refractories in open hearth checker work.

The test the author mentioned for plastic refractories, where the sand rammer is used, is being widely used in the industry and by a good many users of these refractories. It does show whether a plastic is too soft or whether it is too hard to be rammed in a ladle lining or to be rammed in a furnace, so it does not squeeze out or crumble out. So these ASTM tests which have been described will have to be used with caution in any expanded program of testing that the foundryman might carry on. I do say, however, that it might well be part of a handbook on foundry refractories, which we need.

CO-CHAIRMAN KLOPF: The ASTM has been and is going to continue doing a great deal of work on standardization in regard to testing. The foundry industry in all its phases has always had a decided lack of standardization so when we pursue our problems let us keep that one factor in mind. We have to have standardization and development if we are going to have beneficial results.

C. R. SHOOK:<sup>4</sup> We do not test fire brick, but we do test our crucibles. In asking for bids for crucibles, we regularly ask the crucible makers, who are interested in supplying them, for five crucibles for testing purposes. They are paid for those crucibles. We run all five and specify that the order will be awarded on the performance of the best three of the five. Testing has the advantage of keeping the crucible makers in close touch with us and it keeps us in close touch with them. We frequently change from one manufacturer to another. We like this procedure and recommend it to crucible users.

We have two types of furnaces, a gas furnace with one burner and also an Ajax high-frequency induction furnace. The gas furnaces have been in use almost ever since the mint has been in existence. The Ajax furnaces have been in use since the Fall of 1942. We have had six years' experience, and that includes war time operation of 24 hr a day 7 days per week. They have had some hard service. The crucible for the gas furnace is a regular bilge type. The crucible for the Ajax furnace is a straight-walled-cylinder type. We believe we are getting the best possible crucible as a result of these tests.

<sup>2</sup> Cornell University, Ithaca, N. Y.

<sup>3</sup> Ironton Fire Brick Co., Ironton, Ohio

<sup>4</sup> United States Mint, Philadelphia

## EFFECT OF GATING DESIGN ON METAL FLOW CONDITIONS IN THE CASTING OF MAGNESIUM ALLOYS

By

H. E. Elliott \* and J. G. Mezoff \*

### ABSTRACT

*In this paper the authors relate the design of the gating system to the degree of turbulence which occurs during the pouring of magnesium alloy castings, and to the resulting quality of the castings.*

*It was found that improper sprue design may lead to casting defects by the entrainment of gases in the metal stream. Studies were made of the effect of sprue design upon such entrainment of gases.*

*Further work indicated that three casting defects may be caused or exaggerated by gating turbulence; namely, skins, blows, and microporosity. The authors show how the degree of gating turbulence affects the occurrence of these three defects, using test castings of various designs.*

*The value of skim-gates as a method of controlling gating turbulence, and thereby preventing these defects, is shown.*

### Introduction

THE PRESENT PAPER deals with studies of the effect of gating design on the nature of flow conditions in the pouring of magnesium alloys into molds. Attempts were made to determine from a practical point of view just what factors of gating design were critical in the control of pouring turbulence. Observations were also made as to what foundry defects were likely to occur or to be aggravated if turbulence in pouring were not prevented. Attempts were then made to apply the principles evolved to the gating of a commercial casting.

In beginning this work, it was anticipated that contributions to gating practice of commercial significance could be made by laboratory studies. It was recognized, however, that practical foundrymen had already learned a great deal about gating practice, and that conventional gating systems in use at the time this work started embodied a valuable background of practical information. It was therefore proposed to use a conventional gating system, arrived at by practical foundrymen, as a prototype; and then to examine critically the constituent elements of this conventional system.

A study of gating systems in use at the time revealed that in general they were constructed at follows: First,

in casting magnesium alloys, a pouring basin was almost invariably used. This basin was flooded in the early part of the pour, and kept brim-full throughout the duration of the pour. By this means, the pouring time was standardized from one cast to another; the avoidance of certain casting defects was also attributed to this practice. Beneath this pouring basin, a down-sprue carried the metal (in common practice) to the level of the lowest part of the casting cavity. From the base of the sprue, the metal flowed in a horizontal runner, and from this runner the metal passed into the base of the casting cavity through a number of gates. It was most common practice for the choke in this entire gating system to occur in the sprue; that is, the relative dimensions of sprue, runner, and gates were so adjusted that the cross-section available for the flow of metal was at a minimum in the sprue, which hence controlled the rate of flow. Dimensions of the other elements of the gating system were then so adjusted as to avoid restriction to the flow of the metal after it had passed through the sprue.

### Screening of Metal Stream

At the time this work was begun, it was apparent that the above features of gating design had become fairly standard for magnesium alloys. There was considerable unresolved difference of opinion, however, about a number of points. Chief among these, perhaps, was the question as to whether the metal stream need be screened at any point in the gating system. Some foundries invariably placed a skim-gate with or without a pad of steel wool, at a point intervening between the sprue and the gates, and at a level in the mold lower than the base of the casting cavity. In other foundries, the use of screens was regarded as unnecessary, as long as the design of the gating system was proper.

An examination of this conventional gating practice indicated that the most probable sources of pouring turbulence resided in the design of the sprue part of this system, and in the design of that portion of the runner lying immediately under the sprue. It was reasoned that if the metal could be brought into a state of quiet flow in the horizontal runner, it would

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be reasonably easy to prevent turbulence thereafter. Consequently, it seemed reasonable to single out the sprue and its design for initial study.

### A Study of Sprue Design

In conjecturing as to what design of sprue might give best freedom from turbulence when metal was poured into a mold, the system sketched in Fig. 1 was visualized. In this sketch, the pouring basin is repre-

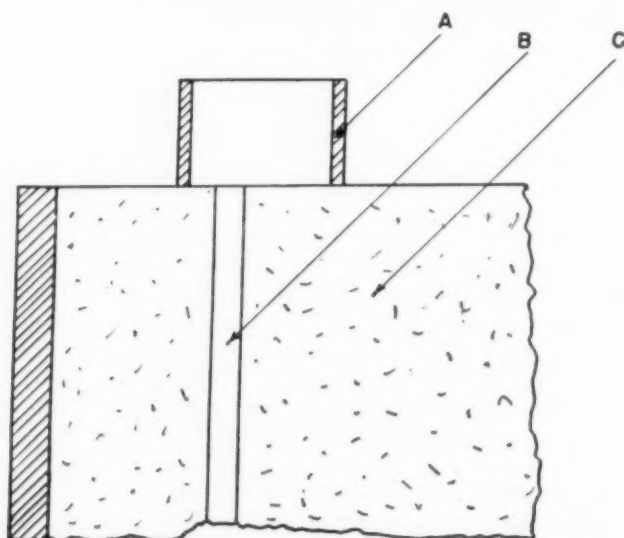


Fig. 1—Conventional sprue system. (A), pouring basin; (B), sprue, and (C), mold material.

sented by "A", while the sprue "B", which is circular in cross-section, is surrounded by the permeable mold material "C". It is assumed that there is no restriction of the rate at which the metal can leave the base of the sprue (this is approximated in conventional gating practice, at least for the early part of the pour). In the pouring of a mold of which the sprue is sketched in Fig. 1, the runner cup is flooded and kept full throughout the pour. When steady state conditions of flow are attained, consider the conditions which must prevail in this system. The following hypothesis seemed a reasonable starting point:

Assuming that the sprue is untapered, the maximum resistance to flow will occur at the mouth of the sprue, due to the turbulent entry conditions at this point. Consequently, the metal will be able to fall through the sprue at a faster rate than the rate at which it can enter the sprue-mouth. The falling metal will tend to flush mold gases out of the sprue ahead of it, creating a partial vacuum just below the mouth of the sprue. Since the mold material is permeable, however, the net result is an aspiration of mold gases into the metal stream, and the admixture of these mold gases with the metal stream.

For a preliminary test of this hypothesis, the apparatus indicated in Fig. 2 was set up. In this sketch, "A" was a  $\frac{3}{4}$ -in. diameter porous alundum tube (representing a sprue through a porous mold material).

This tube was surrounded by a jacket, "B", as indicated. Above the tube was mounted a glass reservoir, corresponding to the pouring basin of Fig. 1. Arrangements were made by which water could be supplied to the reservoir at such a rate as to maintain any desired level in the reservoir.

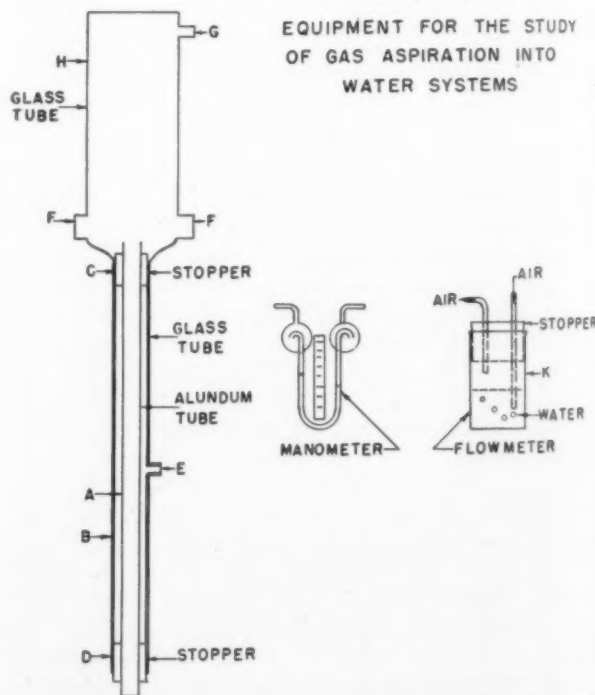
It was found when water was passed through this apparatus, keeping the reservoir at the top full, that a negative pressure of about 2 in. of water formed in the annulus between the tube and the glass jacket. Further evidence of the formation of a negative pressure in the alundum tube was the occurrence of a vortex extending from the top of the reservoir into the top of the alundum tube.

When the bottom of the alundum tube was restricted (corresponding to a restriction of the base of a sprue), water seeped through the walls of the alundum tube, and a positive pressure was developed in the annulus.

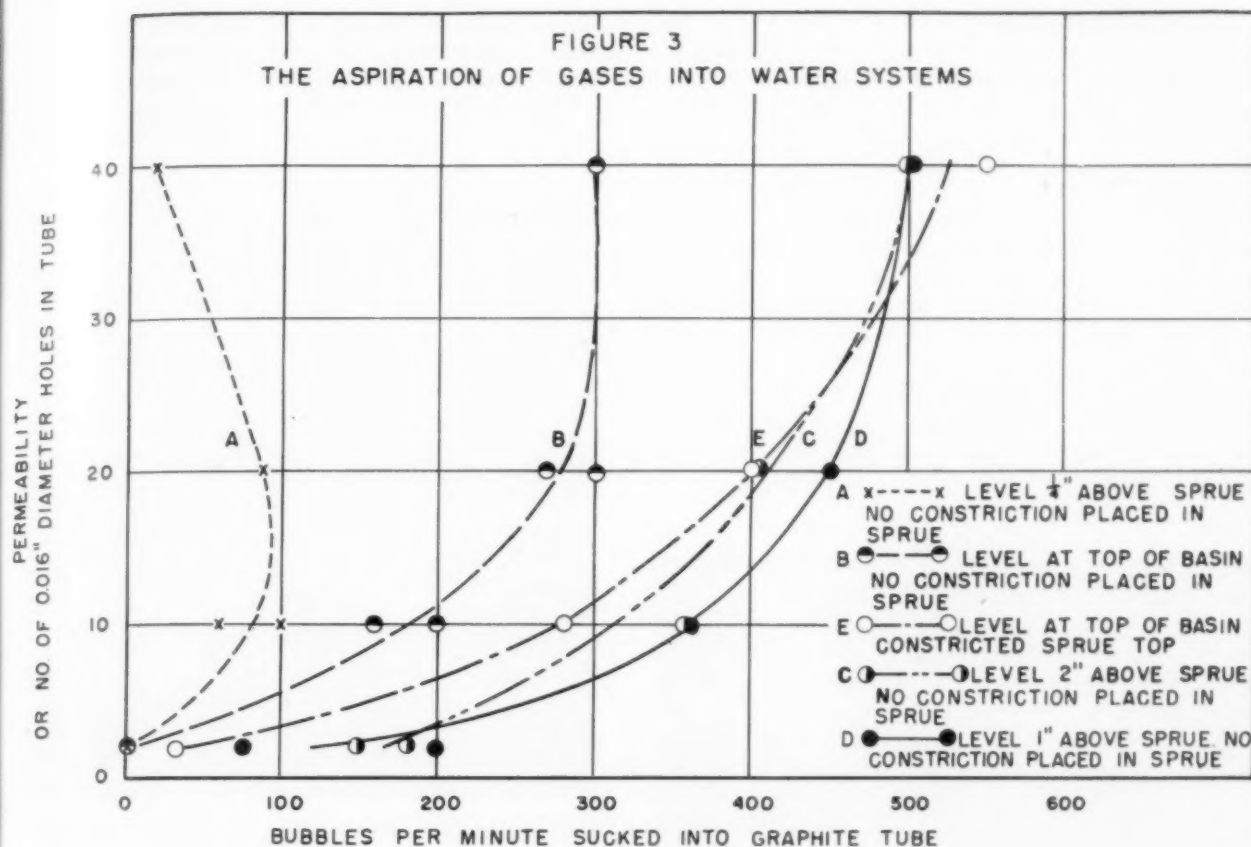
A  $\frac{3}{4}$ -in. O.D. graphite tube was then secured. About twenty 0.013-in. diameter holes were drilled in the side of the tube to give it permeability. The experiment was repeated replacing the alundum tube with the drilled graphite tube. When the water flow was so adjusted as to hold the level in the reservoir at various heights, it was found that no vortex formed with this more permeable tube as long as the depth in the reservoir was greater than one inch over the top

FIGURE 2

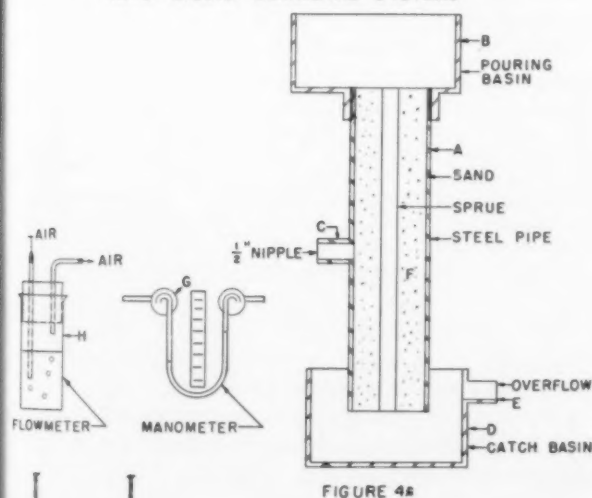
EQUIPMENT FOR THE STUDY OF GAS ASPIRATION INTO WATER SYSTEMS



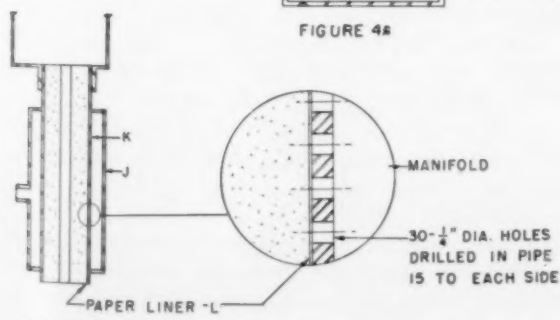
of the tube. When the reservoir was kept full of water, a negative pressure of 5 in. was developed in the annulus. When the bottom of the tube was restricted, a positive pressure of greater than 12 in. of water was developed; when the top was restricted, a negative pressure of more than 12 in. of water was produced.



**FIGURE 4**  
**EQUIPMENT FOR THE STUDY OF GAS ASPIRATION**  
**INTO LIQUID DOWMETAL SYSTEMS**



**FIGURE 4a**



**FIGURE 4b**

Further work with another graphite tube of the same size was conducted over a range of permeabilities. The permeability was varied by starting with two holes each 0.016 in. in diameter, and increasing the number of holes to 40 during the course of the experiments.

For each permeability value, the effect of the static head of water in the reservoir was studied. Figure 2 illustrates the simple flowmeter used in this study to compare the degree of aspiration under various conditions. Figure 3 shows curves indicating the results obtained.

#### Further Studies of Sprue Design

If the phenomena observed in these studies of the flow of water through porous vertical tubes could be assumed to coincide with those occurring when metal is poured down the sprues of sand molds, a number of interesting relations between sprue design and turbulence were indicated. It was necessary to determine whether such phenomena actually occurred when the metal-sand system replaced the water-aluminum or water-graphite system. For this purpose the apparatus sketched in Fig. 4a was prepared.

A standard steel pipe (A), 3 in. in diameter, 20 in. long, and threaded on one end was used as a flask. Fastened to the top of the flask was a removable pouring basin (B) made of boiler plate. A side-arm (C) was welded to the side of the flask, by means of which pressure and gas-flow measurements were to be made. A catch basin (D) with overflow (E) were placed under the flask in such a way as to provide a liquid seal at the bottom.

Core sand was rammed up and baked within the flask to form the sprue core (F), which was 20 in. long. An open-end manometer (G) and a crude flowmeter (H) were available for making rough determinations of pressure variations and gas flow.

It was anticipated that the large volumes of gas produced by the agents and binders normally used in the foundry would interfere with flow and pressure measurements. Therefore, beryllium was added to the commercial AZ63 alloy used in the tests so that it would be unnecessary to use agents in the sand. Blast sand was used as the base sand because of its high permeability, and a commercial grade of water glass was employed as the binder. When such a core sand mixture was baked for one-half hour at 1000 F, little further gas evolution occurred when magnesium alloy at 1350 F was poured through it.

#### Use of Liquid Seal

In the earliest experiments, it was found that if a liquid seal were not provided at the bottom of the mold, no aspiration could be measured. A tapered sprue 1 in. in diameter at the top and 1½ in. in diameter at the bottom was found to aspirate considerable quantities of gas when a liquid seal was provided. It was not possible to obtain a quantitative comparison in degree of aspiration among various aspirating sprues tested, either visually or by means of available motion picture equipment.

This equipment was abandoned after a few tests in favor of the equipment indicated in Fig. 4b. Using this equipment, a variety of sprue designs were tested as to comparative tendencies to aspirate gases.

A 1½-in. diameter sprue (untapered) gave a large amount of aspiration. When the bottom of this sprue was restricted, however, by placing a section of skim-gate over it in such a way as to restrict flow, a positive pressure was generated within the manifold. It is believed that the air in the manifold heated and expanded, that no aspiration occurred, and that gas was therefore forced out of the manifold through the flowmeter. This observation of the effect of the gas expansion during heating shows how significant previously noted cases of aspiration were.

A tapered sprue, 2½ in. in diameter at the top by 1 in. in diameter at the bottom gave a slight positive pressure, again indicating little or no aspiration. A sprue tapered from 2 in. in diameter at the top to ¾ in. diameter at the bottom produced a large positive pressure in the manifold. These experiments showed that if the sprue is tapered sufficiently from top to bottom, aspiration can be markedly reduced.

At the outset of this experimental work, a hypothesis had been formulated which assumed that in a sprue system consisting of a pouring basin mounted over an untapered sprue, the major resistance to flow would be at the upper mouth of the sprue. This assumption led to the prediction of aspiration effects. The experimental demonstration that such aspiration occurs is regarded as confirmation of the assumption that the major resistance to flow in such systems does occur at the mouth of the sprue. In another paper<sup>1</sup>, the authors present other evidence that this is the

case. At any rate, the effect of tapering the sprue from large at top to small at bottom is readily understood. Such alterations in sprue design shift the major restriction to flow to the base of the sprue, and thereby eliminate the formation of a negative pressure beneath the sprue mouth and resulting aspiration.

It was questioned whether other means might be found of shifting the major resistance to flow away from the sprue mouth by changing the sprue design. A possible clue to such designs lay in the knowledge that some magnesium alloy foundries used "slot" sprues with the purpose of reducing pouring turbulence. This suggested the advisability of testing sprues with a large ratio of cross-sectional perimeter area, with respect to whether they aspirated gases.

Accordingly, a battery of four ½-in. diameter sprues was molded up in the equipment sketched in Fig. 4b. When the metal was poured through this sprue system, a slight positive pressure was developed in the manifold. With batteries of two and three ½-in. diameter sprues, a similar positive pressure resulted. A rectangular sprue ½ in. x 1 in. cross section aspirated large quantities of gas. On the other hand, a ¾ in. x 2 in. rectangular sprue, or a battery of three such slots, produced a significant positive pressure in the manifold. Table 1 tabulates the work relating sprue design to aspiration.

TABLE 1—EFFECT OF SPRUE DESIGN ON DEGREE OF ASPIRATION

Type of Sprue	Size at Top, in.	Size at Bottom, in.	Restriction at Bottom	Amount of Aspiration
Round, tapered...	1 Dia.	1½ Dia.	No	Much
Round, tapered...	1½ Dia.	1 Dia.	Yes	Moderate
Round, untapered...	1½ Dia.	1½ Dia.	No	Much
Round, untapered...	1½ Dia.	1½ Dia.	Yes, by skim-gate 28% voids	None
Round, tapered...	2½ Dia.	1 Dia.	Yes	None
Round, tapered...	2 Dia.	¾ Dia.	Yes	None
Four pencil sprues...	½ Dia.	½ Dia.	No	None
Three pencil sprues...	½ Dia.	½ Dia.	No	None
Rectangular.....	2x1	2x1	No	Much
Rectangular.....	2x½	2x½	No	Little
Rectangular.....	2x¾	2x¾	No	None
Three slot sprues...	2x¾	2x¾	No	None

It is assumed that the reason for the failure of slot type sprues (or other sprues of large ratio of perimeter to area) to aspirate gases lies in the fact that with sprues of this type the major resistance to flow is not concentrated at the sprue mouth. Probably the resistance to flow of the sidewalls of the sprue becomes significant with respect to that of the sprue mouth, when the cross section of the sprue has a sufficiently high ratio of perimeter to area.

#### Relation of Gating Design to Casting Quality

##### Description of test casting

It was desired to continue these studies of gating turbulence along the following lines: (a) to determine what other elements of the conventional gating system (other than the sprue) are critical with respect to the effect of design on gating turbulence;



(b) to determine what defects are caused or aggravated by permitting uncontrolled amounts of turbulence to occur in the gating; (c) to determine the relative roles of good sprue design and the use of skim-gates in assuring good casting quality.

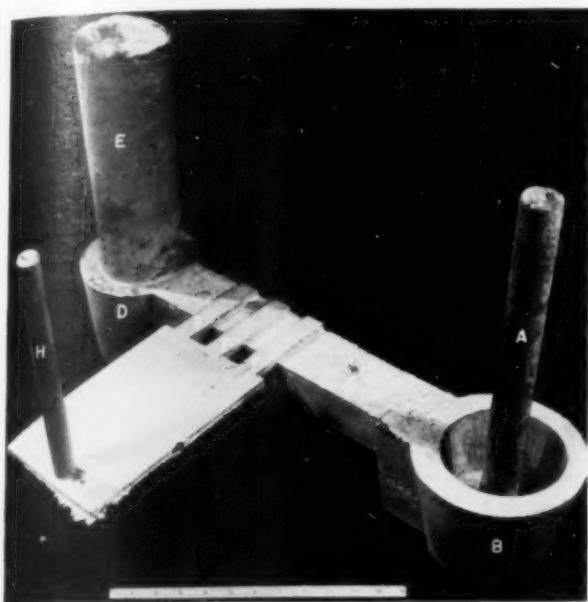


Fig. 5—Experimental casting used in studies of effects of gating turbulence on casting quality.

To accomplish these aims the test casting pictured in Fig. 5 was designed. In this casting, *A* was the pouring sprue. It could be terminated at the parting line of the mold, or it could be extended through the cope and into the drag to within 1 in. of the bottom of the well, *B*, as indicated. Thus, a liquid seal could be provided at the base of the sprue. From the well *B*, the metal passed into the runner *C*, which was designed to have several times the cross section of the largest sprue to be studied. This was done to insure that the choke would occur in the sprue as in conventional gating practice, although in commercial practice the ratio of runner area to sprue area would not be exaggerated to the extent indicated in this test casting.

At the opposite end of the runner *C*, a second well, *D*, was placed and over this well there was a large riser or pop-off, *E*. The well *D*, was amply proportioned to serve as a catch-basin for the first metal poured, for it was desired that steady-state conditions of flow be attained before the actual test panel filled. The riser *E*, served as a pop-off, preventing the metal from entering the test panel at too great a velocity.

The test panel itself is shown as *G*. This was a 6 x 10 x 1/4-in. horizontal panel, gated into one 6-in. end and vented, *H*, at the opposite end. Its purpose was simply to "tap-off" a sample of the metal flowing in the runner *C* after steady-state flow conditions were attained. It was expected that this sample of metal would freeze rather rapidly, entrapping any included gases. It was hoped that by examining the quality of this test panel, inferences could be drawn as to how

much turbulence was occurring in this gating system under steady-state conditions of flow.

This test casting was sufficiently flexible to permit investigation of the independent effects of many gating variables. For example, it was easy to substitute sprues of almost any desired design for the untapered round sprue shown (Fig. 5). The liquid seal at the sprue base could be used or omitted at will. A screen area, as indicated in Fig. 6, could be incorporated into the system readily. The gates into the panel could be varied as in Fig. 7, as could the venting of the test panel.

#### Description of defects encountered

Preliminary studies indicated that two different defects in the test casting were sensitive to variations in gating design. One of these defects consisted of



Fig. 6—Experimental casting showing screen areas used in some tests.

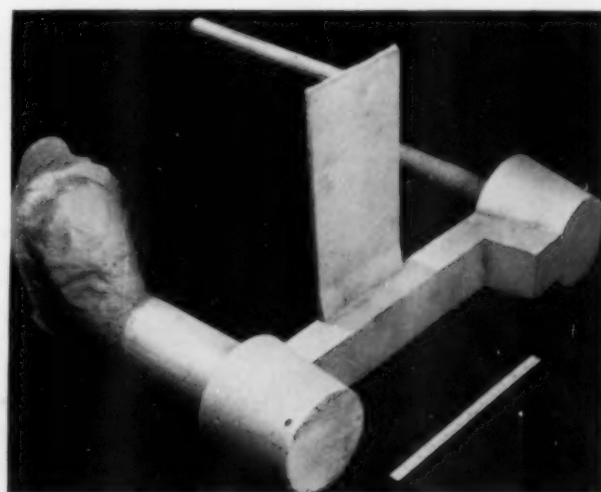


Fig. 7—Experimental casting showing web-gating of the test panel.

irregularly shaped holes varying in size from a tiny hole the size of a pin head up to as much as an inch in diameter. This defect was called blows.\*

These blows were most readily detected by radiographing the panel. The radiograph also provided a convenient method of comparing the extent to which a number of panels contained this defect. Figures 8 to 12 indicate the rating system which was arrived at in comparing these test castings as to the severity of the blow condition. If there were no blows in the castings, its rating was "0". Panels containing various numbers of blows were assigned ratings from "1" to "5" as indicated.

The other defect which appeared when gating turbulence was permitted was called "skins". This consisted of a reaction product between the metal and the mold gases which became entrapped in the metal. The aggregate of reaction-products which were called skins were not in general very dense, but instead consisted of films of oxidized metal mixed with some unreacted gases. They had the appearance of a "foam" of metal films mixed with gases, rather than

\* As will be developed in this paper, such blows were clearly shown to be induced by gating turbulence. Other studies at Dow not covered in this paper indicate that there are other sources of green sand blows, aside from gating turbulence. The presence of such impurities in the molding sand as bituminous coal will result in a form of green sand blow which is characterized in appearance by its almost perfectly spherical shape, and hence by the extreme sharpness and roundness of its appearance on a radiograph. The occurrence of this type of green sand blow also seems to be affected by the moisture content of the sand. The blows described here, on the contrary, were not related either to sand impurities or to sand moisture content, but appeared to be the result of entrapment of mold gases in turbulently flowing metal. It is probable that all of these sources of blows are of commercial importance.

the appearance of a pocket of metal oxide in a powdered form.

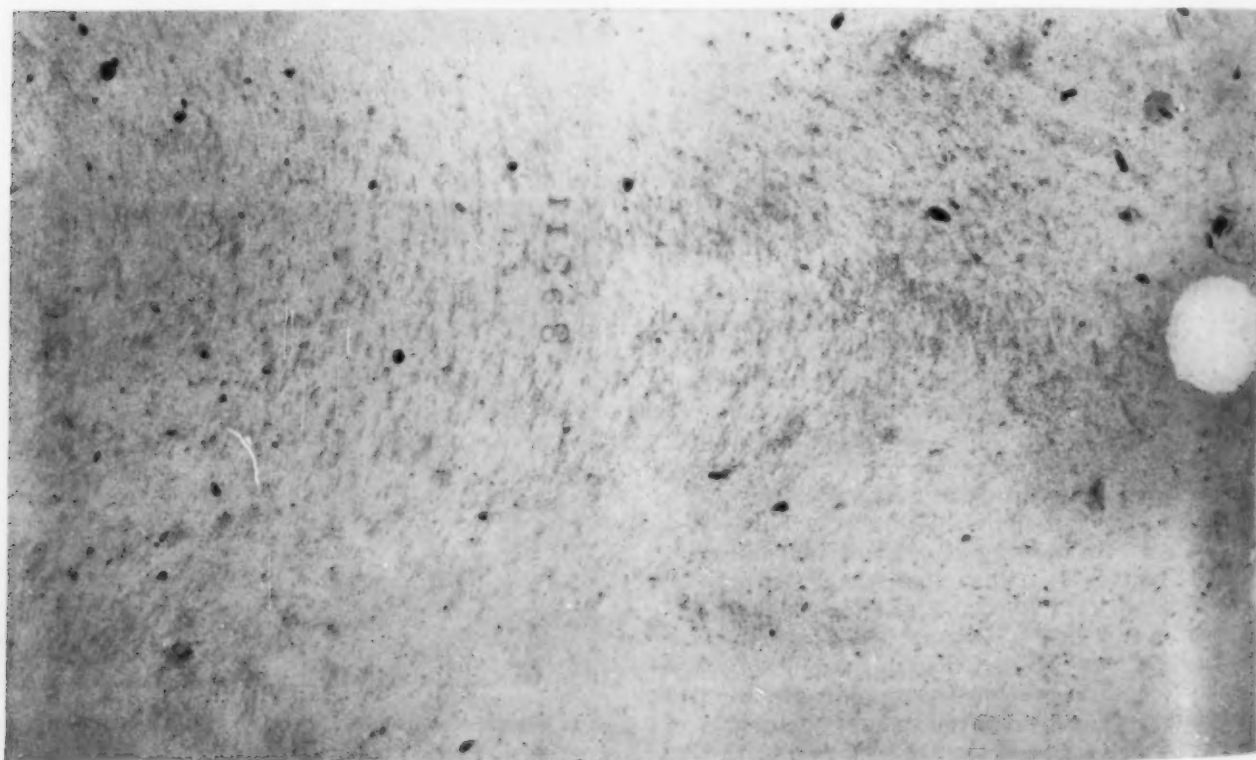
Skins were best detected by fracturing the test panel and examining the fresh fractures. A rating system was set up to compare the relative amounts of skin in panels cast using various gating systems as shown in Fig. 13. If the fracture was free of skins, a "0" rating was assigned. Ratings of "1" through "5" were assigned to indicate increasing amounts of entrapped skin.

It was not uncommon for both blows and skins to be evident in the fractures of a single panel. For this reason, the fractures of test panels were classified in the following types, as indicated in Fig. 14: Type I fractures exhibited skins, but no blows; Type II fractures had both blows and skins; Type III fractures had blows, but no skins, and Type IV fractures were clean, showing neither blows nor skins.

#### *Effect of gating design on occurrence of blows*

Preliminary tests made with the experimental casting shown in Fig. 5 showed that by use of a 1½-in. diameter untapered sprue and a low pouring temperature, panels containing large numbers of blows were obtained. Since this sprue was an aspirating type of sprue, previous experience had led the authors to expect that the use of a sprue of this type would lead to intimate admixture of the metal with mold gases. Hence, the occurrence of the blows was attributed to the entrapment of mold gases in the flowing metal, to the failure of these gases to react at this low pouring

Fig. 8—Rating system for describing severity of blows in test panels; upper limit of a "1" rating. Reduced approximately ¼ in reproduction.

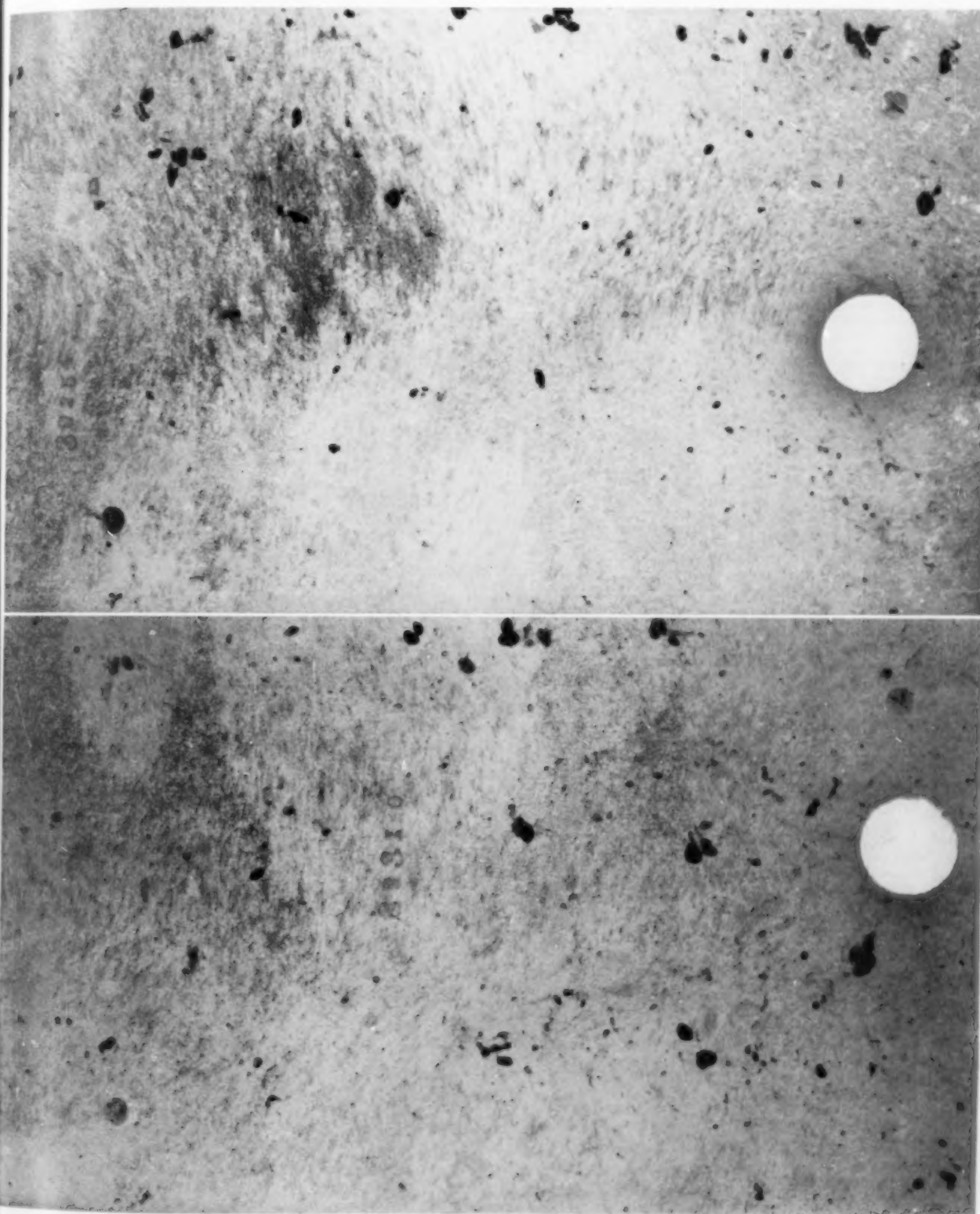


temperature, and to the failure of the gas bubbles to escape from the rapidly freezing panel.

It was believed, then, that a successful method of detecting gating turbulence had been found. It seemed

Fig. 9 (above)—Rating system for describing severity of blows in test panels; upper limit of a "2" rating. Fig. 10 (below)—Upper limit of a "3" rating. Reduced  $\frac{1}{4}$  in reproduction.

logical that the occurrence of blows in the test panel could be used as an indication of the presence of turbulence in the gating; and that if variations in the gating design led to variations in the severity of blows in the test panel; such variations could be attributed to variations in gating turbulence. A rather comprehensive program was therefore conducted to relate various features of gating design to the resulting degree of gating turbulence, using the occurrence of





blows in the test panel as a criterion of turbulence. The results of this program are tabulated in Table 2.

*Fig. 11 (above)—Rating system for describing severity of blows in test panel; upper limit of a "4" rating.*

*Fig. 12 (below)—Upper limit of a "5" rating. Reduced  $\frac{1}{4}$  in reproduction.*

In all of the tests listed in Table 2, the melt preparation was the same. Virgin ingot AZ63 alloy was melted in steel crucibles under crucible type flux. No grain-refining or degassing treatment was applied, but inasmuch as virgin ingot metal was used, and precautions were taken to avoid gas pick-up in melting, there was no danger that the poured metal contained suf-

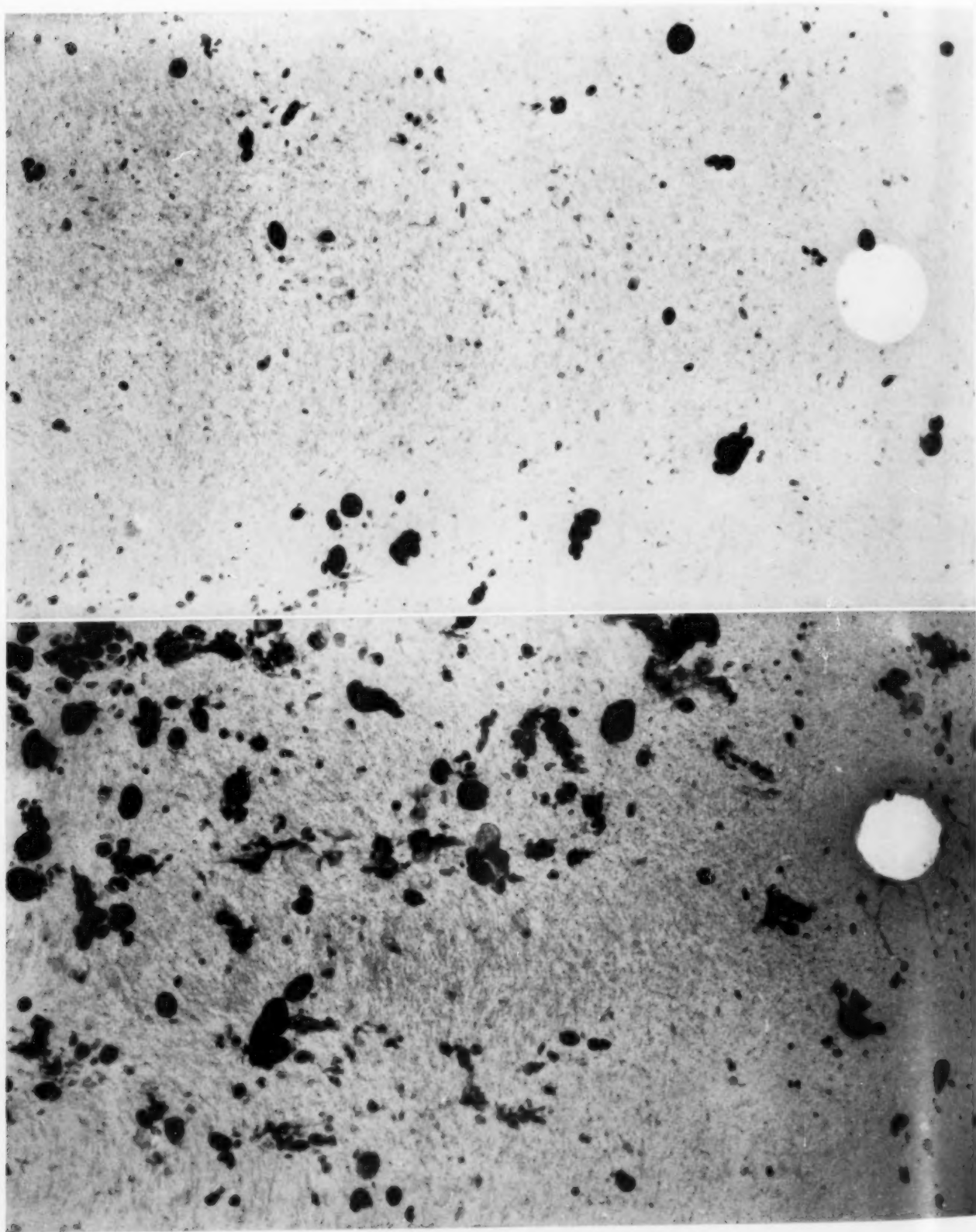




Fig. 13—Rating system for Type I fracture.

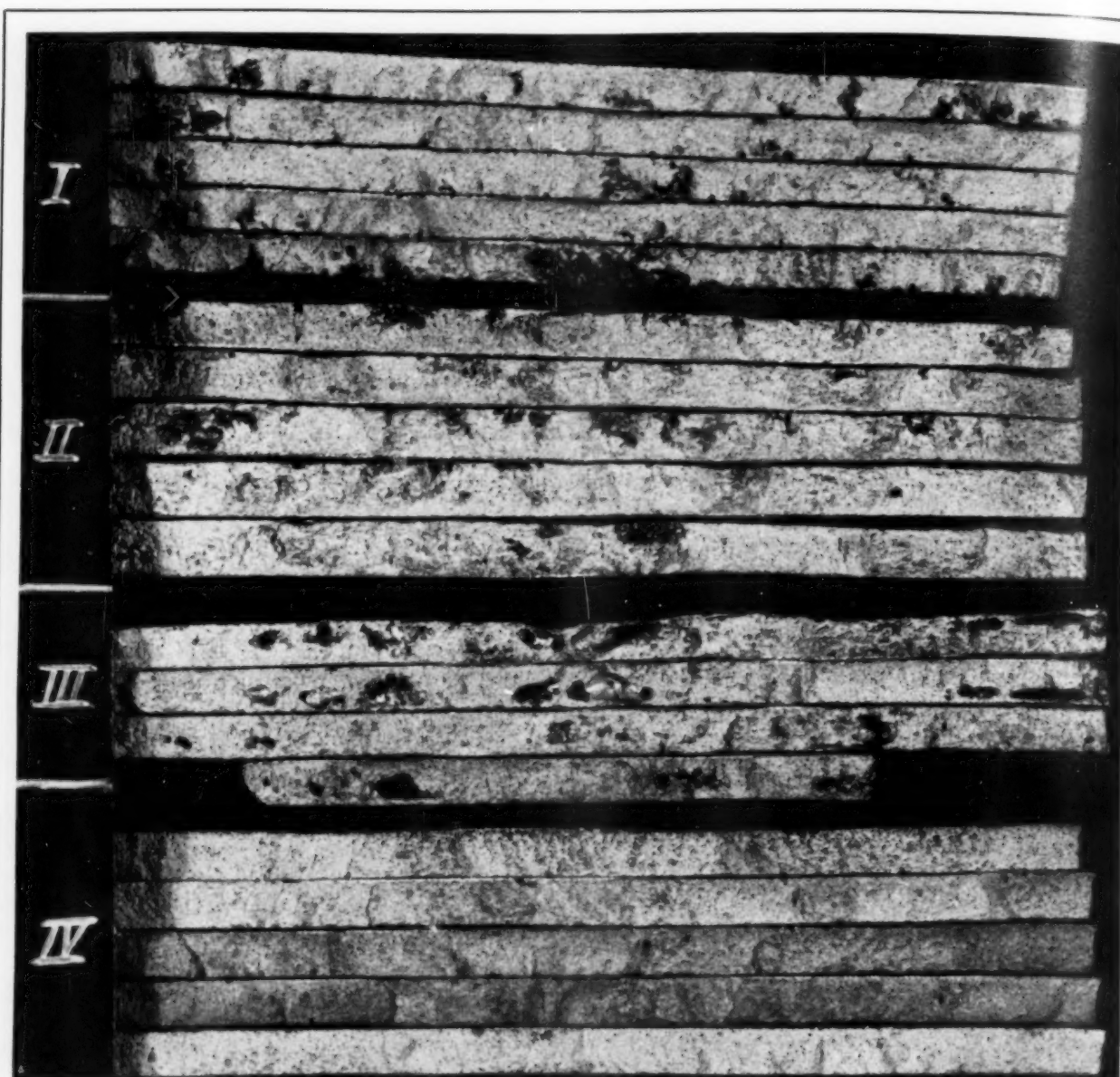


Fig. 14—Fracture types observed in study of experimental castings.

ficient hydrogen gas to cause gas holes from this source. In all tests, the mold material was the same—a fluoride-inhibited sand of the composition given in Appendix A. The same pouring basin was used in each test—a bottomless steel cup 10 in. long by 5 in. wide by 5 in. deep. Pouring technique was to flood this cup as early as possible, and to keep it full throughout the pour.

The significance of the various headings in Table 2 is described:

**Column 1: "Test No."** In this column a serial number is assigned to each kind of test that was made.

**Column 2: "Sprue Description."** The design of the sprue used in each test is described in this column. All round sprues were molded in green sand. Batteries of slots were made in a plaskon-bonded core sand, and these cores were thoroughly baked before use.

**Column 3: "Liquid Seal."** In this column it is indicated whether the sprue terminated at the parting line, or extended into the drag to within 1 in. of the bottom of the well as in Fig. 5.

**Column 4: "Screen area."** Here it is indicated whether a screen area with skim-gate as in Fig. 6 was used. In those cases in which a screen area was used, the skim-gate material was tinned perforated steel sheet with the No. 4 perforation.

**Column 5: "Steel wool."** In most cases in which a screen area was used, a loose pad of steel wool was placed on the sprue side of the skim-gate. In this column it is indicated whether this practice was followed.

**Column 6: "Pouring temperature."** This column is self-explanatory.

**Column 7: "Vents."** In this column it is indicated



TABLE 2—EFFECT OF GATING DESIGN ON OCCURRENCE OF BLOWS IN THE TEST CASTING

Test No.	Sprue Description	Liquid Seal	Screen Area	Steel Wool	Pouring Temp., °F.	Vents	Panel Gating	A.P.T. sec.	No. of Tests	A.B.R.	Remarks
1	1½ in. Dia.	No	No	No	1200	One	Finger	6.0	4	5.0	
2	1½ in. Dia.	Yes	No	No	1200	One	Finger	5.5	6	4.5	
3	2¼ in. Dia. at top x 1 in. Dia. at Bottom	Yes	No	No	1200	One	Finger	5.0	3	4.5	
4	1½ in. Dia. at top x 1 in. Dia. at Bottom	Yes	No	No	1200	One	Finger	5.5	3	3.0	
5	5 slots, each ⅜ in. x 2 in.	Yes	No	No	1200	One	Finger	3.0	2	5.0	
6	3 slots, each ⅜ in. x 2 in.	Yes	No	No	1200	One	Finger	4.0	3	3.0	
7	2 slots, each ⅜ in. x 2 in.	Yes	No	No	1200	One	Finger	5.5	5	3.0	
8	1½ in. Dia. at top, ¾ in. Dia. at bottom	Yes	No	No	1200	One	Finger	7.0	1	1.0	
9	1½ in. Dia.	Yes	Yes	Yes	1200	One	Finger	6.5	5	0.5	
10	2 slots, each ⅜ in. x 2 in.	Yes	No	No	1200	One	Finger	5.5	5	3.0	Special Crucible Lip
11	5 slots, each ⅜ in. x 2 in.	Yes	Yes	Yes	1200	One	Finger	4.0	3	4.0	
12	3 slots, each ⅜ in. x 2 in.	Yes	Yes	Yes	1200	One	Finger	4.5	2	1.0	
13	5 slots, each ⅜ in. x 2 in.	Yes	Yes	No	1200	One	Finger	4.0	3	3.0	
14	2 slots, each ⅜ in. x 2 in.	Yes	Yes	No	1200	One	Finger	6.0	3	1.0	
15	1½ in. Dia.	Yes	No	No	1200	One	Finger	5.0	1	5.0	Runner restricted to 2 in. Dia. half round opening.
16	1½ in. Dia.	Yes	No	No	1200	One	Finger	5.5	2	4.5	Runner restricted to 1½ in. Dia. half round opening.
17	1½ in. Dia.	Yes	No	No	1200	One	Finger	5.5	5	3.5	Runner restricted to 1½ in. Dia. half round opening.
18	1½ in. dia.	Yes	No	No	1200	One	Finger	6.5	5	...	1 in. Dia. half round opening — all panels misran.
19	5 slots, each ⅜ in. x 2 in.	Yes	Yes	Yes	1200	One	Web	4.0	5	2.0	
20	3 slots, each ⅜ in. x 2 in.	Yes	Yes	Yes	1200	One	Web	7.0	3	2.0	
21	2 slots, each ⅜ in. x 2 in.	Yes	Yes	No	1200	One	Web	8.0	1	2.0	
22	2 slots, each ⅜ in. x 2 in.	Yes	Yes	Yes	1200	One	Web	7.0	3	1.0	
23	5 slots, each ⅜ in. x 2 in.	Yes	Yes	Yes	1200	Three	Web	...	5	1.0	
24	5 slots, each ⅜ in. x 2 in.	Yes	Yes	Yes	1200	One	Web	4.0	3	2.0	
25	3 slots, each ⅜ in. x 2 in.	Yes	No	No	1200	Three	Web	...	3	3.0	
26	1½ in. Dia.	Yes	Yes	Yes	1200	Three	Web	...	5	1.5	
27	2¼ in. Dia. Top, 1 in. Bottom	Yes	Yes	Yes	1200	Three	Web	6.0	5	0.5	
28	5 slots, each ⅜ in. x 2 in.	Yes	No	No	1200	Three	Web	4.5	5	2.0	
29	3 slots, each ⅜ in. x 2 in.	Yes	No	No	1200	Three	Web	5.5	5	1.5	
30	1 slot ⅜ in. x 2 in.	Yes	No	No	1300	Three	Web	19.0	1	0.0	⅜ in. thick panel
31	2 slots ⅜ in. x 2 in.	Yes	No	No	1300	Three	Web	9.0	1	0.0	⅜ in. thick panel
32	3 slots, ⅜ in. x 2 in.	Yes	No	No	1300	Three	Web	6.0	1	0.0	⅜ in. thick panel
33	4 slots ⅜ in. x 2 in.	Yes	No	No	1300	Three	Web	5.0	2	1.0	⅜ in. thick panel
34	5 slots ⅜ in. x 2 in.	Yes	No	No	1300	Three	Web	4.0	1	0.0	⅜ in. thick panel
35	1½ in. Dia.	Yes	No	No	1300	Three	Web	6.0	1	1.0	⅜ in. thick panel

whether a single 1-in. diameter vent was used at the panel end as in Fig. 5, or whether this venting was increased by the use of three such vents.

**Column 8: "Panel Gating."** This column indicates whether the metal entered the test panel through three finger-gates (as in Fig. 5), or through a continuous web-gate (as in Fig. 7).

**Column 9: "A.P.T."** This column gives an average of the pouring times for the castings poured in each type of test. The pouring times were measured by an electric time-recorder, which measured the lapse of time between the moment the metal reached a contact placed in the pouring cup, and the moment the metal reached a second contact in the well *D* (Fig. 5) at the level of the parting line.

**Column 10: "No. of Tests."** Here is indicated the number of times each test was repeated.

**Column 11: "A. B. R."** Here is recorded the average blow rating (Fig. 8 to 12) of the panels resulting from each test.

A detailed study of the tests listed in Table 2 and

their results yields many significant conclusions as to the effect of gating design on the occurrence of turbulence. A brief survey of these tests follows.

In Tests 1 and 2, an aspirating sprue is used (1½ in. diameter, untapered). In neither test was a screen area employed, but in both tests the panel was gated through three finger gates and vented by means of a single 1-in. diameter vent. In Test 1, no liquid seal was provided, while in Test 2, a liquid seal was used. Test 1 was repeated four times; Test 2, six. Both tests yielded panels containing a very large number of blows, as indicated by the high A.B.R. (average blow rating). It was inferred from the appearance of these panels that at some point in the gating system, an intimate admixture of metal with mold gases occurred, resulting in the entrainment of bubbles of gases in the metal stream, and that these bubbles were carried into the test panel and entrapped there by rapid freezing. It was considered probable that the major source of this turbulence was the design of the sprue, inasmuch as an aspirating type of sprue was used. Test

2 indicated that the sprue base was probably not the major source, since provision of a liquid seal did not significantly reduce blows.

In an attempt to determine whether the sprue in Tests 1 and 2 was the major source of turbulence, it was decided to duplicate the procedure of these tests with the exception that a non-aspirating type sprue would be used. In Tests 3, 4, and 8 this was done by using sprues which were tapered with the large diameter up. In each case, a liquid seal was provided at the sprue base. In Test 3, a sprue  $2\frac{1}{4}$  in. in diameter at the top by 1 in. in diameter at the bottom was used. It was not effective in reducing the A.B.R. of the test panel. In Test 4, a sprue  $1\frac{1}{2}$  in. in diameter at the top and one inch in diameter at the bottom was used. This sprue was successful in reducing the A.B.R. to 3.0. This was regarded as a very significant test. The test appeared to be comparable in every way to Test 2. All factors of gating design except the taper of the sprue were held constant, and the pouring time was the same in both tests. A significant reduction in blow rating attended the change to a non-aspirating type of sprue. The failure of the non-aspirating type sprue of Test 3 was believed to be due to excessive velocity at the sprue base; the excessively fast rate of flow through the 1-in. diameter opening was believed to lead to turbulence in the well at the base of the sprue. In Test 8, a sprue  $1\frac{3}{4}$  in. diameter at the top by  $\frac{3}{4}$  in. in diameter at the bottom gave the very low A.B.R. of 1.0. However, this test could not be compared directly with Test 2, because of the much slower pouring rate in Test 8.

#### Effect of Sprue Design

While the panels from Test 4 contained significantly fewer blows than those from Test 2, clearly demonstrating an effect of sprue design on turbulence, they still contained so many blows as to indicate the occurrence of considerable turbulence in the gating system. Attempts continued, therefore, to find means of further reducing gating turbulence. In Tests 5, 6, and 7, slot sprues were tested. In Test 7, conditions compared with those of Test 2, except that a sprue consisting of two 2-in. x  $\frac{3}{8}$ -in. slots replaced the  $1\frac{1}{2}$ -in. diameter sprue. Again a significant effect of sprue design on panel quality was seen. At similar pouring rates, the non-aspirating type sprue gave better quality than did the aspirating type. In Test 5, a non-aspirating type sprue consisting of five slots, each 2 in x  $\frac{3}{8}$  in., failed to give a low A.B.R., apparently because of excessive pouring rate.

The best quality panel thus far attained still contained so many blows as to indicate considerable turbulence somewhere in the system. The crucible in use in these tests had a pouring lip of such design as to give a wide flat stream in pouring. It was suspected that this type of pouring stream might be resulting in entrapment of gases. Therefore, in Test 10, the methods of Test 7 were repeated with the exception that the crucible used was fitted with a pouring lip which gave a tubular stream. The results indicated no advantage of this practice, and the A.B.R. of 3.0 which was obtained still left a question as to where the

responsible turbulence was occurring. The quest for this source was abandoned for a time, and attention was turned to the effect of screening practice on panel quality.

To determine the effect of filtering the metal stream, Test 9 was first made. This test duplicated the procedure of Test 2, except that a screen area with skim-gate and steel wool as illustrated in Fig. 6 was added to the runner. Although Test 9 gave longer pouring times than Test 2, this is accounted for by the larger volume of metal required to fill the screen area, and it does not appear that the screen area reduced the rate of metal flow, under steady-state flow conditions, appreciably. Test 9 was repeated 5 times, yielding the remarkably low A.B.R. of 0.5. It was evident that the use of the screen area was remarkably effective in improving test panel quality. Whether the screen area acted purely as a filter, or whether it acted in some way to reduce turbulence was not determined.

Further studies of the effectiveness of screen areas were made in Tests 11 to 14. It is seen from the data on these tests in Table 2, that a screen area with skim-gate, with or without steel wool was very effective in giving good panel quality as long as moderate pouring rates prevailed. Excessive pouring rates, however, caused poor panel quality in spite of the use of a screen area in the runner.

Later, in Tests 19 to 26, a clue was found as to the cause of poor panel quality at excessive pouring rate when a screen area was used in the runner. In this series of tests, the effect of web gating, in place of finger gating, and the effect of using three panel vents instead of one, were investigated. It was found that with three panel vents and a web-gate into the panel, good panel quality resulted from the use of a screen area even when a very fast pouring rate, obtained by five  $\frac{3}{8}$ -in. x 2-in. slots, were used (Test 23). The use of this extra panel venting and improved gate was of particular advantage, however, when a moderate pouring rate, comparable to that obtained with a  $1\frac{1}{2}$ -in. diameter sprue, was used. This indicated that when the effectiveness of the screen area broke down with excessive pouring rates, the cause of the poor quality obtained was turbulence beyond the screen area, probably within the panel itself; and that such turbulence beyond the screen area could be corrected by improving the gating and venting of the test panel.

#### Use of Tapered Sprue

In Test 27, the combined effects of reducing aspiration by use of a tapered sprue, and the use of a screen area, with good gating and venting of the panel, was tested. Five repeats of this test gave a very good A.B.R. of 0.5.

Attention was then returned to the problem of securing good panel quality without screening by so designing the gating system as to minimize turbulence. In Tests 15 to 18, an attempt was made to eliminate turbulent flow in the sprue and in the sprue-base by choking the horizontal runner. In each test a  $1\frac{1}{2}$ -in. round sprue with liquid seal was used. It was found that the runner could be choked down at a position close to the well *B* of Fig. 5 to a semi-circular opening

1 1/4 in. in diameter, without reducing the pouring rate. When this amount of choke was used, an A.B.R. of 3.5 resulted, representing some improvement over the quality obtained with no choke. It was suspected that this choke was effective in preventing turbulence in previous parts of the gating system, but that "spurtling" of the metal through the choke in the runner generated turbulence at this point, nullifying any gains made in early parts of the system.

In Tests 28 and 29, attempts were continued to find a gating method which would give high freedom from blows in the absence of screening. In these tests, batteries of slots 3/16 x 2 in. in cross section, using a liquid seal were tested. Test 29 used three such slots, and gave a pouring rate comparable to that attained with a 1 1/2-in. diameter sprue. A quite low A.B.R. of 1.5 resulted. Test 29, when compared with Test 2, is a very striking demonstration of the effectiveness of using a non-aspirating type sprue to reduce pouring turbulence. Comparing Test 29 with Test 7 or Test 10, it is seen that the 3/16-in. thick slots were more effective in reducing turbulence than were the 3/8-in. slots first tested. Comparing Test 29 with Test 9 and Test 26, it is indicated that the use of a screen area was somewhat more effective in giving good panel quality than was the use of the best sprue system found, in the absence of a screen area.

#### Further Studies of Effect of Gating Design on Casting Quality

The studies thus far had established a relationship between gating turbulence and the occurrence of blows in the test panel at low pouring temperatures. The occurrence of this defect had then been used as a measure of the degree of turbulence resulting from various gating systems which were compared. It was desired now to study the relationship between gating design and the occurrence of skins in the panel fracture, and to determine under what conditions skins result from turbulence and under what conditions blows occur. It was also desired to determine what effect other variables, such as molding material, mold atmosphere, and composition of the melt, might have on the occurrence of these defects.

Accordingly, the tests outlined in Table 3 were conducted. The same basic test casting was used as in previous tests. The significance of the various headings in Table 3 is described as follows:

**Column 1: "Test No."** In this column a serial number is assigned to each test that was made. A test, as the term is used here, consisted of the casting of a number of test castings under identical conditions.

**Column 2: "Sprue Description."** In this column, the design of the sprue is indicated, as in Table 2.

**Column 3: "Screen area."** In this column it is indicated whether screen area, with skim-gate and steel wool, was used as previously described.

**Column 4: "Liquid seal."** Here it is indicated whether a liquid seal was provided at the sprue base, as previously described.

**Column 5: "Metal."** In this column, the composition of the melt is indicated. Variations in melt composition are described in more detail in the following discussion of the results of these tests.

**Column 6: "Pouring Temperatures."** In this column it is indicated at what temperature the metal was poured.

**Column 7: "Sand."** In this column, it is indicated whether the fluoride or the non-fluoride type sand was used. These two types of sand are described in detail in Appendix A.

**Column 8: "Panel Venting."** In this column it is indicated whether three 1-in. diameter vents were used on the panel end, or whether these were omitted.

**Column 9: "Fracture Type."** Panels were fractured to evaluate quality as discussed above and as shown in Fig. 13. Fractures were classified in four types as shown in Fig. 14.

**Column 10: "Fracture Rating."** Type I and IV fractures were rated as discussed above and pictured in Fig. 13.

**Column 11: "No. of Castings."** In this column it is indicated how many castings were made with an identical set of conditions, constituting a test.

The results of the tests described in Table 3 are briefly discussed in the following paragraphs.

#### Fracture Quality Standards

**Test 1.** It was first necessary to establish a standard fracture quality with which the quality obtained by a variation in gating could be compared. The standard chosen was the quality obtained by gating the panel in such a way as to provide numerous opportunities for turbulent flow to occur, and taking no steps to strain out resulting bubbles or skins. The procedure was to use a sprue of moderately fast pouring rate (1 1/2-in. diameter) which previous studies showed to aspirate gases. No liquid seal was provided at the base of the sprue, which set up another condition leading to turbulence. There was also opportunity for turbulent motion in the pouring cup. No effort was made to strain out defects originating from these three sources. An amply large runner and panel gate, combined with adequate panel venting, insured that mixture of mold gases with the metal would not occur in these parts of the system. The metal used was commercial AZ63 melted under crucible type flux, and poured at 1250 F. The sand was the fluoride type, of which the composition is given in Appendix A. Frequent reference will be made in the following discussion to this standard test. It will be referred to as the standard turbulent gating system. Note (Table 3) that this system produced Type I fractures, and the skins were present in quantities which gave an average rating of 4.4.

#### Effect of pouring temperature

**Tests 2 and 22.** To check the effect of pouring temperature on panel quality using the standard turbulent gating system, metal was poured at 1500 F and at 1200 F into molds similar to those used in Test 1. As at 1250 F, the panels poured at 1500 F gave Type I fractures, with an average rating of 4.7, not significantly different from those of Test 1. Those panels poured at 1200 F, however, gave distinctly different fracture appearance. Open blows containing little skin characterized these fractures. This indicated a lower reaction rate between the mold gases and the



TABLE 3—EFFECT OF GATING DESIGN ON FRACTURE QUALITY OF THE TEST CASTING

Test No.	Sprue Description	Screen Area	Liquid Seal	Metal	Pouring Temp., °F.	Sand	Panel Venting	Frac- ture* Type	Aver. Fracture Rating**	No. of Cast- ings	Remarks
1	1½ in. Dia.....	None	None	AZ63 Alloy	1250	F.	3-1 in. D.	I	4.4	5	(Standard turbulent gating system)
2	1½ in. Dia.....	None	None	AZ63 Alloy	1500	F.	3-1 in. D.	I	4.7	3	
3	1½ in. Dia.....	Yes	None	AZ63 Alloy	1250	F.	3-1 in. D.	I	0.7	3	
4	1½ in. Dia.....	Yes	None	AZ63 Alloy	1500	F.	3-1 in. D.	I	0.7	4	
5	5 slots, 2 in. x ½ in....	None	Yes	AZ63 Alloy	1350	F.	3-1 in. D.	I	1.5	2	
6	5 slots, 2 in. x ½ in....	None	Yes	AZ63 Alloy	1500	F.	3-1 in. D.	I	0.7	3	
7	1½ in. Dia.....	None	None	AZ63 Alloy	1250	F.	3-1 in. D.	I	2.0	2	Screen in pouring cup
8	1½ in. Dia.....	None	None	AZ63 Alloy	1500	F.	3-1 in. D.	I	3.7	3	Screen in pouring cup
9	2 in. Dia.....	Yes	Yes	AZ63 Alloy	1250	F.	None	II	...	3	Very severe blows on one panel
10	2 in. Dia.....	Yes	Yes	AZ63 Alloy	1500	F.	None	II	...	2	Very severe blows on one panel
11	½ in. Dia.....	None	None	AZ63 Alloy	1500	F.	3-1 in. D.	I	1.0	3	
12	½ in. Dia.....	None	Yes	AZ63 Alloy	1500	F.	3-1 in. D.	I	1.7	3	
13	1½ in. Dia.....	None	None	AZ63 Alloy	1250	F.	3-1 in. D.	I	3.0	2	
14	1½ in. Dia.....	None	None	AZ63 and Be and Be	1500	F.	3-1 in. D.	I	3.0	1	
15	1½ in. Dia.....	None	None	M1 Alloy	1350	F.	3-1 in. D.	I	4.5	2	
16	1½ in. Dia.....	None	None	M1 Alloy	1500	F.	3-1 in. D.	I	5.0	2	
17	1½ in. Dia.....	None	None	AZ63 Alloy	1250	N.F.	3-1 in. D.	II	...	3	
18	1½ in. Dia.....	None	None	AZ63 Alloy	1500	N.F.	3-1 in. D.	I	4.0	1	
19	1½ in. Dia.....	None	None	AZ63 Alloy	1250	F.	3-1 in. D.	I	4.0	3	Mold skin-dried by torching
20	1½ in. Dia.....	None	None	AZ63 Alloy	1500	F.	3-1 in. D.	I	4.7	3	Mold skin-dried by torching
21	6 slots, 2 in. x ½ in....	None	Yes	AZ63 Alloy	1500	N.F.	3-1 in. D.	I	1.0	2	
22	1½ in. Dia.....	None	None	AZ63 Alloy	1200	F.	3-1 in. D.	III	...	5	
23	1½ in. Dia.....	None	None	AZ63 Alloy	1200	N.F.	3-1 in. D.	III	...	3	
24	1½ in. Dia.....	None	None	AZ63 Alloy	1500	N.F.	3-1 in. D.	I	4.5	2	Mold baked 12 hours at 300° F.
25	6 slots, ½ in. ....	Yes	Yes	AZ63 Alloy	1500	F.	3-1 in. D.	IV	0.0	4	
26	1½ in. Dia.....	None	None	M1 Alloy	1250	F.	3-1 in. D.	II	...	3	
27	6 slots, ½ in. ....	None	Yes	AZ63 Alloy	1500	F.	3-1 in. D.	I	1.0	2	Mold gassed with SO
28	1½ in. Dia.....	None	None	AZ63 Alloy	1500	F.	3-1 in. D.	I	4.0	2	⅜ in. thick panel used
29	1½ in. Dia.....	None	None	AZ63 Alloy	1200	F.	3-1 in. D.	I	4.0	3	⅜ in. thick panel used
30	1 in. Dia.....	None	None	AZ63 Alloy	1500	F.	3-1 in. D.	I	2.0	3	
31	1½ in. Dia.....	None	None	AZ63 Alloy	1500	F.	3-1 in. D.	I	3.0	3	Cope — 6 in. deep

\*Average Fracture Rating:

"0" rating = freedom from skins.

"1"-"5" rating = with increasing amounts of skin.

\*\*Fracture Type:

Type I fracture = Skins, but no blows.

Type II fracture = Both blows and skins.

Type III fracture = Blows, but no skins.

Type IV fracture = Clean, neither blows nor skins.

metal at the lower metal temperature. Gases entrained in the metal appeared not to react, but rather to remain as bubbles in the metal, appearing in the fracture as blows.

#### *Effect of strainer practice*

*Tests 3 and 4.* These tests showed the effectiveness of a screen area, with skim-gate and steel wool, in the runner in preventing the defects which occur when the standard turbulent gating system was used. Presence of the screen and steel wool in the runner was the only variation in these tests from the procedures of Tests 1 and 2. Skin ratings were reduced from greater than 4.0 to less than 1.0.

*Tests 7 and 8.* In these tests, the standard turbulent gating system, as well as the metal and mold treatment, of Tests 1 and 2 was reproduced, with the exception that a screen "basket" containing steel wool (as shown in Fig. 15) was placed in the pouring cup.

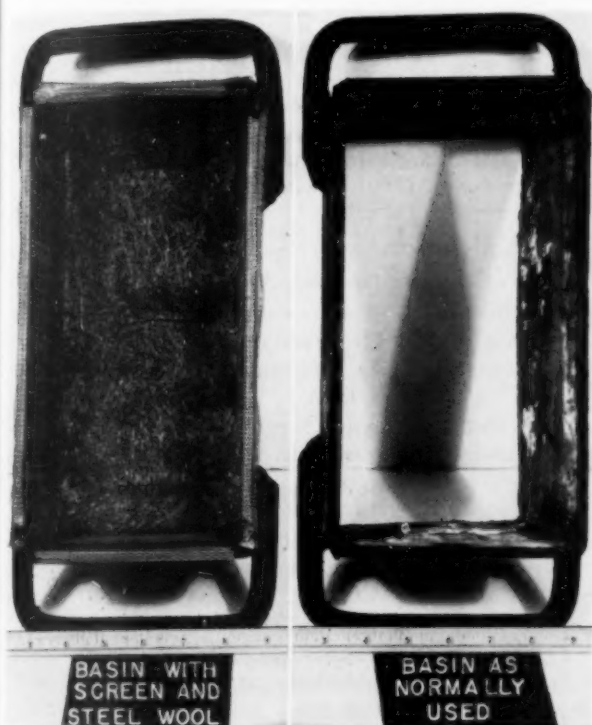


Fig. 15—Photographs showing pouring basin use, with and without a screen basket.

Poor fractures of Type 1 resulted. The ineffectiveness of a strainer at this point showed that many reaction-product skins form in the gating, and are not merely a carry-over from the melt. Strainers are effective only in the absence of further sources of gas-metal mixture between the strainer and the casting. This point is further demonstrated in the following tests.

*Tests 9 and 10.* In these tests a very fast pouring rate (with a 2-in. diameter sprue) was used in conjunction with a screen and steel wool in the runner and inadequate panel venting. The same temperatures and molding sand were used as in Tests 1, 2, 3, and 4. It is seen that with these high pouring rates,

highly defective panels are possible even with good strainer practice in the runner. It is interesting that Type II fractures were obtained. Blows were prevalent, although at similar pouring temperatures, only skins had been previously seen. It is concluded that these blows must have had a source which occurred after the metal passed through the strainer. Mold gases were unable to leave the panel cavity fast enough for the metal to enter without admixture. Due to rapid freezing in the panel and the low metal temperature at the time this mixture occurred, the mold gases did not react with the metal to be converted completely to skins.

It is readily seen that a set of conditions such as were set up in these tests could result in blows in a commercial casting despite careful screening practice and/or sprue design.

#### *Effect of Reducing Metal Turbulence*

*Tests 5 and 6.* While the above tests showed that the effect of turbulence in causing skins can be largely corrected by careful screening practice, it was not known whether the screens exerted their beneficial effect by a filtering action or by preventing the responsible turbulence from occurring. In Tests 5 and 6, a non-aspirating sprue was employed, with a liquid seal provided at its base. Thus two important sources of gas-metal mixture were minimized. In addition, the sprue used has a pouring rate of 40 per cent lower than that of the standard 1½-in. diameter sprue, which further insured reduced turbulence. No strainer was used in the runner or pouring cup. Fluoride type sand was used as in previous tests. Good fracture quality resulted, about the same as, or slightly poorer than, that attained with the use of the standard turbulent gating system used in conjunction with screen area and steel wool. This improvement over standard turbulent gating system practice is attributed to reduced turbulence in the sprue and sprue-base.

#### *Effect of Pouring Rate*

*Test 30.* In this test, the standard turbulent gating system was used, having a pouring rate approximately equal to that used in Tests 5 and 6. Quality was much better than that obtained with a 1½-in. diameter sprue, but not as good as that obtained using the slot sprue of like pouring rate with a liquid seal.

*Tests 11 and 12.* In these tests, a very low pouring rate was obtained by using a small (½-in. diameter) sprue. Although in Test 11, no liquid seal was provided at the sprue base, allowing opportunity for turbulence, fracture quality was good. This again exemplified the often observed fact that low pouring rates are conducive to good freedom from blows and skins. It is not clear whether this can be attributed wholly to reduced turbulence at lower pouring rates, or whether the effect may be partly explained by the fact that more time is provided for skins and bubbles to settle out of the metal stream before the metal enters the casting cavity.

#### *Effect of Mold Materials and Mold Atmospheres*

*Tests 17, 18, and 23.* In Tests 1, 2, and 22 above described, the fracture quality of panels poured at

1200 F, 1250 F, and 1500 F was compared using the standard turbulent gating system and the fluoride type sand and described in Appendix A. In Tests 17, 18, and 23, a similar comparison was made using the non-fluoride type sand, also described in Appendix A. The following tabulation summarizes those comparisons:

Type of Sand	Pouring Temperatures		
	1200 F	1250 F	1500 F
Non-Fluoride Sand	Blows	Blows and Skins	Skins
Fluoride Sand	Blows	Skins	Skins

Thus at a metal temperature of 1200 F, the mold gases generated in either sand are unreactive or only slowly reactive with the metal; the gases therefore appear in the fracture as empty cavities. At a metal temperature of 1500 F, the mold gases react rapidly with the metal to form skins in both sands. Only at intermediate temperature (in the case of this particular casting, at 1250 F) is a difference in reaction rate of the metal with the mold gases of the two sands seen. In the fluoride sand, the mold gases react rapidly with the metal to form skins, giving a Type I fracture. In the non-fluoride sand, the metal gases react only partially with the metal to give a mixture of skins and blows in the panel—a Type II fracture. This slight difference in behavior of the metal in contact with the mold atmospheres developed by the two sands does not suggest any particular advantage of one type over the other.

*Tests 19 and 20.* In these tests, the standard turbulent gating system of Tests 1 and 2 was used. Fluoride sand was used, but the molds were thoroughly skin-dried by torching before closing. If the reaction by which skins are formed is associated with moisture content of the mold atmosphere, one might expect this relation to be reflected in a lower skin rating for the panels poured into skin-dried molds. No reduction in skin rating nor change in fracture type was associated with this change in mold preparation, however.

*Test 24.* The standard turbulent gating system, non-fluoride sand, and a 1500 F pouring temperature were used. It was attempted to dry these molds effectively by baking 12 hr at 300 F, to reduce moisture content of the mold atmosphere. The molds were still steaming when removed from the oven, and it is not clear how effectively mold-atmosphere water-content was reduced. The fracture quality ratings indicate no improvement associated with this mold treatment.

*Test 27.* In this test, a fluoride sand mold was flushed with  $\text{SO}_2$  before pour-off, as a variation of mold-atmosphere. The purpose was to determine, if possible, whether high  $\text{SO}_2$  content affects the tendency toward the formation of skins in the casting. Turbulence was kept to a minimum by using a slot sprue and a liquid seal, but no screen area was used in the runner. A pouring temperature of 1500 F was chosen. Comparing results with those of Tests 5 and 6, no appreciable effect of the  $\text{SO}_2$  on fracture quality is seen.

#### Comparison of various gating techniques

An attempt was made to evaluate the relative effectiveness of attempts to control skin formation by (a) reducing turbulence, (b) screening, and (c) reducing turbulence plus screening.

*Tests 3 and 4.* These tests were described above. The methods of Tests 3 and 4 are considered to simulate those of foundries which rely on skim-gates to secure freedom from skins and blows.

*Test 21.* In this test, a non-aspirating type sprue was used in conjunction with good sprue-base design (liquid seal) to minimize turbulence. The non-fluoride sand was used, without screen or steel wool in the runner. This test was an attempt to simulate the practice of foundries said to rely solely on careful gating design to control defects due to turbulence. Comparison of the fracture results from Tests 3 and 4 and Test 21 shows that the two systems gave approximately equivalent quality.

*Test 25.* In this test, a non-aspirating type sprue was used in conjunction with good sprue base design and with a skim-gate in the runner. Of four panels cast by this method, all were visually perfect fracture quality. This indicated a slight advantage of the non-aspirating type sprue even when a skim-gate was used. Previous tests had indicated that the sprue design had no significant effect on blows, as long as a skim-gate was used.

#### Effect of Melt Composition

*Tests 13 and 14.* In these tests, the value of addition of beryllium to AZ63 alloy in reducing skin formation was investigated. The standard turbulent gating system was used with fluoride sand and pouring temperatures of 1250 F and 1500 F. Comparing results with Tests 1 and 2, one sees a significant but not outstanding reduction in average skin rating. Fractures were of Type I at both pouring temperatures.

*Tests 15 and 16.* In these tests, the tendency toward skin formation of M1 alloy was tested. The standard turbulent gating system was used. Results were comparable to those obtained when AZ63 alloy was poured into similar molds.

*Test 26.* In this test M1 alloy was poured at 1250 F into the standard turbulent gating system in fluoride sand. Whereas AZ63 alloy under these conditions gives a Type I fracture containing only skins, the M1 alloy showed very numerous blows and a little reaction product skin, Type II fractures. The comparison of the behavior of M1 and AZ63 alloys cast at various pouring temperatures with the standard turbulent gating system in fluoride sand is summarized as follows:

Alloy	Pouring Temperature			
	1200 F	1250 F	1350 F	1500 F
AZ63	Blows	Skins	.....	Skins
M1	.....	Blows and Skins	Skins	Skins

#### Effect of casting wall thickness

*Tests 28 and 29.* All previous panels had been of  $\frac{1}{4}$ -in. thickness; in these tests, this thickness was increased to  $\frac{3}{8}$  in. The standard turbulent gating sys-



tem was used, with pouring temperatures of 1200 F and 1500 F. At both pouring temperatures, Type I fractures were obtained with skin ratings of "3" and "5." In a 1/4-in. panel, blows would have occurred at the 1200 F pouring temperature, but only skins were observed in this thicker panel. This is probably accounted for by the longer time of solidification of the thick panel, allowing time for the gas to react with the metal with the formation of skins.

#### Effect of Cope Height

Test 51. In this test, Test 2 was duplicated except that the cope depth was reduced from 12 in. to 6 in. Otherwise the standard turbulent gating system was used. Type I fractures were obtained; skin ratings were "3" in all panels, representing a moderate reduction in skin formation.

These studies thus provided a reasonably clear understanding of the effects of turbulence in producing skins and blows, and the relationship between various elements of gating design and the degree of turbulence that results. The question was asked as to whether any other foundry defects, such as microporosity tendency, were aggravated by turbulence in the gating. The work described in the following section

of this paper was directed toward answering this question.

#### Effect of Gating Turbulence on Tendency Toward Micro-Porosity

It is evident that when liquid magnesium alloys are allowed to come into admixture through turbulence with mold gases, reactions occur which lead to the formation of non-metallic skins and films. Such non-metallics may form through various reactions—through the reaction of magnesium with free oxygen or with free nitrogen, for example. When metal is poured into a green sand mold, much water vapor is formed through the vaporization of the tempering moisture. Thus it would seem reasonable that considerable reaction between magnesium and water vapor might occur, if the metal is mixed intimately with the mold gases. This reaction might be expressed:



Thus such a reaction would release nascent hydrogen in intimate contact with the liquid alloy. Another effect of such a reaction might be to form quantities of MgO in a fine state of distribution throughout the metal.

TABLE 4—EFFECT OF GATING TURBULENCE ON POROSITY TENDENCY OF THE WEDGE PANEL

Description of Gating				Inches of Porosity		
Sprue	Screen	Liquid Seal	Chlorination	Test Wedge	Control Wedge	Control Wedge
A. Low-turbulence gating systems						
5 slots (2 in. x 1/8 in.)	Yes	Yes	15 min.*	0	...	...
5 slots (2 in. x 1/8 in.)	Yes	Yes	15 min.	8	...	...
5 slots (2 in. x 1/8 in.)	Yes	Yes	15 min.	0	...	...
5 slots (2 in. x 1/8 in.)	Yes	Yes	15 min.	0	...	...
5 slots (2 in. x 1/8 in.)	Yes	No	15 min.	0	...	...
5 slots (2 in. x 1/8 in.)	Yes	No	15 min.	0	...	...
5 slots (2 in. x 1/8 in.)	Yes	Yes	15 min.	3	...	...
5 slots (2 in. x 1/8 in.)	Yes	Yes	15 min.	7 1/2	...	...
6 slots (2 in. x 1/8 in.)	Yes	Yes	15 min.	6 1/2	0	0
6 slots (2 in. x 1/8 in.)	Yes	Yes	15 min.	0	0	0
4 slots (2 in. x 3/16 in.)	Yes	Yes	15 min.	0	0	0
3 slots (2 in. x 3/16 in.)	Yes	Yes	15 min.	8 1/2	0	0
4 slots (2 in. x 3/16 in.)	Yes	Yes	15 min.	0	0	0
6 slots (2 in. x 3/16 in.)	Yes	Yes	15 min.	0	0	0
1 slot (2 in. x 3/16 in.)	Yes	Yes	15 min.	0	0	0
4 slots (2 in. x 3/16 in.)	Yes	Yes	15 min.	0	0	0
6 slots (2 in. x 3/16 in.)	Yes	Yes	15 min.	0	0	0
B. High-turbulence gating systems						
1 1/2 in. Dia.	None	None	15 min.	4 1/2	...	...
1 1/2 in. Dia.	None	None	15 min.	0	...	...
1 1/2 in. Dia.	Yes	None	15 min.	0	0	0
1 1/2 in. Dia.	Yes	None	15 min.	Some Porosity	0	0
1 1/2 in. Dia.	Yes	None	15 min.	7	0	0
2 in. Dia.	Yes	None	15 min.	4	0	0
1 1/2 in. Dia.	Yes	None	5%	8	0	0
1 1/2 in. Dia.	Yes	None	5%**	7 1/2	0	0
1 1/2 in. Dia.	Yes	None	5%	6 1/2	0	0
1 1/2 in. Dia.	Yes	None	5%	0	0	0
1 1/2 in. Dia.	Yes	None	5%	Some Porosity	0	0
1 1/2 in. Dia.	Yes	None	5%	6	0	0
1 1/2 in. Dia.	None	Yes	5%	2	0	0
1 1/2 in. Dia.	None	Yes	5%	3 1/2	0	0
1 1/2 in. Dia.	None	Yes	5%	0	0	0

\*Chlorine gas was bubbled rapidly through the melt for 15 min. at 1400°F. This was equivalent to about 3% Cl<sub>2</sub> by weight.

\*\*Chlorine was bubbled through the melt at 1400°F. from a weighed cylinder, until 5% Cl<sub>2</sub> by weight had been used.

It was our purpose to determine whether turbulence such as might occur in the pouring of a poorly gated mold might result in the formation of sufficient nascent hydrogen to gas the melt by its solution, and consequently to increase the porosity tendency of the metal.

Previous experience (as previously described) had established known means of varying the degree of turbulence in the gating of the test casting pictured in Fig. 5. In earlier tests a 6-in. x 10-in x 1/4-in. panel had been attached to the runner of this casting to sample the metal stream in such a way as to detect blows and skins. It was now desired to sample the metal stream in such a way as to evaluate its porosity tendency.

Busk, Marande, and Newhams<sup>2</sup> have described a porosity test casting by means of which they evaluated the relative porosity tendencies associated with various melt preparatory techniques. It was decided to attach this wedge panel, with its riser, to the runner of our test casting as a sample of the metal stream. The gate into the riser of the wedge panel was made to have the same dimensions as in the wedge panel casting described by the authors referred to. It was proposed to use the radiographic quality of this panel as a measure of the porosity tendency of the metal stream.

Into molds of this type was poured metal which had been degassed by thorough chlorination, at a pouring temperature of 1500 F. Gating variations were made as indicated in Table 4 to vary the degree of turbulence. The resulting radiographic quality of the attached wedge panel was measured and recorded. With each casting of this kind, two wedge panel molds (as described in the paper<sup>1</sup> above referred to) were poured, as a check on the initial quality of the metal poured. Table 4 summarizes the results of these tests. It is seen that the "low-turbulence" gating systems gave a much higher proportion of sound panels than did the high turbulence systems, although unaccountable amounts of porosity are found in panels from both groups. It is noted that the control panels always gave a radiographically sound wedge, indicating that variations in the quality of the test wedges were not due to variations in gas content of the melts.

These preliminary indications that turbulence may affect the porosity tendency of the metal were confirmed in tests to be described in the following section of this paper.

#### Application of Various Gating Systems To a Commercial Casting

It was desired to determine what commercial significance the principles of gating derived from our studies with a test casting might have. To evaluate these principles, it was decided to pour a commercial casting, varying the degree of turbulence in the gating in accordance with the principles evolved in earlier work. It is probable that thin-walled magnesium castings are more sensitive to defects caused by gating turbulence than are heavy-walled castings. Accordingly, an oil pan (Fig. 16) was chosen for study. This casting is typical of a wide range of thin-walled castings, for

which the properties of magnesium alloys make them particularly well suited.

The experimental procedure used was to apply various gating systems to this pan casting, to radiograph the resulting castings, and then to fracture them and examine the fractures for skins. It was first essayed to relate the occurrence of blows and of skins to the degree of turbulence permitted in the gating, just as had been done before with the test casting.

Attempts to relate the occurrence of blows to the gating method used were unsuccessful. The reasons for this lack of success were several: It was necessary to pour at relatively high temperatures to avoid misruns. This favored the complete reaction of entrapped mold gases to form skins, which were radiographically visible or invisible depending on their density. There were occasional radiographically visible blows which had the sharp, circular appearance of blows due to other causes than gating turbulence (see footnote, page 6). Also, occasional surface pits on the castings produced irregularly shaped dense spots on the radiographs, indistinguishable from those caused by skins or blows. Under these circumstances, it was not possible to assign a meaningful "blow-rating" to the castings, and hence blows off the green sand could not be related to gating practice for this casting. This was not interpreted as evidence that gating turbulence is unlikely to cause blows in commercial castings. A less well-vented casting of other design might be very sensitive to gating variations.

#### Skin Ratings by Fracturing

Attempts to relate skins in the casting fractures to gating turbulence were more successful. Skin ratings were obtained by fracturing the pan into approximately 20 pieces of equal size, and examining all fractures for skins. The total number of agglomerates of skin found in the fractures of a given pan was divided by the number of pieces examined, and this quotient was multiplied by 100 to give the skin rating for the pan. A rating of about "5" thus represented a pan of very clean fracture quality, since such a rating indicated that only one agglomerate of skin could be found in the fractures of the approximately 20 pieces examined.

In Table 5, it is seen that the relationship between skins and gating design discovered in earlier work

TABLE 5—EFFECT OF GATING DESIGN ON FRACTURE QUALITY OF THE OIL PAN CASTING

Test No.	Sprue Description	Screen	Liquid Seal	Chlorination	Average Skin Rating	No. of Castings
1	2 in. Dia.....	Yes	No	No	6	5
2	2 in. Dia.....	No	No	No	85.0	2
3	6 slots (2 in. x 3/16 in.)...	No	No	No	40	2
4	6 slots (2 in. x 3/16 in.)...	No	Yes	No	12	5
5	6 slots (2 in. x 3/16 in.)...	Yes	Yes	No	13	3
6	1 3/8 in. Dia.....	No	No	No	90	5
7	1 3/8 in. Dia.....	Yes	No	No	2	2
8	1 3/8 in. Dia.....	No	No	5%	36	5
9	1 3/8 in. Dia.....	Yes	No	5%	2	4
10	6 slots (2 in. x 3/16 in.)...	No	Yes	5%	7	5
11	6 slots (2 in. x 3/16 in.)...	Yes	Yes	5%	2	4

with the test castings as well borne out in the tests with the oil pan. Non-aspirating sprue systems, used in conjunction with a liquid seal at the sprue base, gave better quality than was attained with sprues of round cross-section. A more effective way of controlling the amount of skin in the fractures was by use of a screen area with steel wool in the runner.

It is to be noted that with comparable sprue systems, the fractures of pan castings showed much less skin than did the fractures of the 6-in. x 10-in. x 1/4-in. panel of the test casting. It is believed that this was

TABLE 6—EFFECT OF GATING TURBULENCE ON POROSITY IN THE OIL PAN

System No.	Sprue	Liquid Seal	Screen Area	Chlorination	Porosity Rating*
I	1 3/8 in. Dia.....	None	None	None	60
II	1 3/8 in. Dia.....	None	None	5%	67
III	1 3/8 in. Dia.....	None	Yes	None	52
IV	1 3/8 in. Dia.....	None	Yes	5%	34
V	6 slots (2 in. x 3/16 in.)...	None	None	5%	55
VI	6 slots (2 in. x 3/16 in.)...	None	None	None	95
VII	6 slots (2 in. x 3/16 in.)...	Yes	None	5%	47
VIII	6 slots (2 in. x 3/16 in.)...	Yes	Yes	None	63
IX	6 slots (2 in. x 3/16 in.)...	Yes	Yes	5%	31

\*For description of rating method, see Appendix B.

due to the much shorter runner distance between the sprue base and the casting in the case of the test casting (Fig. 5) as compared to the pan casting (Fig. 16). The longer runner distance in the pan casting allowed time for skins in the metal stream to be settled out and entrapped by the sand walls surrounding the runner.

The tests tabulated in Tables 6 and 7 are relevant to the relationship between gating turbulence and porosity in the pan casting. The porosity rating systems used in these tables are described in Appendix B. In Table 6, sprue systems were compared which gave approximately equal over-all filling rates of the casting. The tests were so chosen as to determine the effect on porosity of three variables: (a) degassing of the melt; (b) use of a non-aspirating sprue; and (c) use of a screen area in the runner. Three significant trends are noted. (a) Chlorination of the melt was required to get a low porosity rating; (b) in the absence of screen areas, the non-aspirating sprue system gave a lower porosity rating than the aspirating system; (c) the lowest porosity rating resulted when a

Fig. 16—Photographs showing various views of the pan casting.

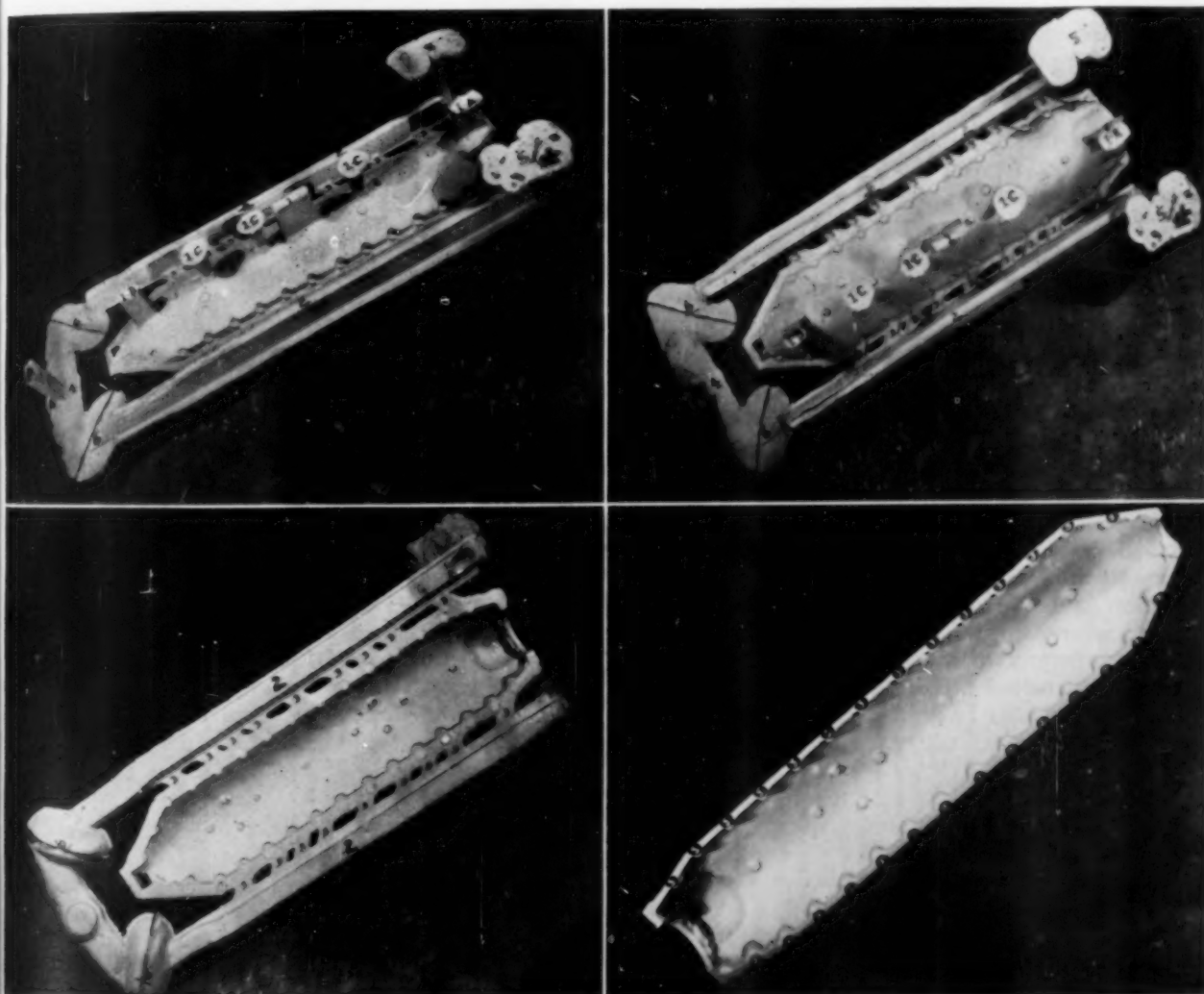




TABLE 7—EFFECT OF GATING DESIGN ON POROSITY IN THE OIL PAN

System No.	Sprue	Liquid Seal	Screen Area	Chlorination	Porosity Rating*	System Average
II 1 $\frac{3}{8}$ in. Dia.....		None	None	5%	80	113
					149	
					84	
					118	
IV 1 $\frac{3}{8}$ in. Dia.....		None	Yes	5%	133	70
					79	
					79	
					69	
VII 6 slots (2 in. x $\frac{3}{16}$ in.).....		Yes	None	5%	54	89
					132	
					84	
					68	
IX 6 slots (2 in. x $\frac{3}{16}$ in.).....		Yes	Yes	5%	66	48
					97	
					57	
					63	
					38	
					35	

\* Rating system similar to that used in Table 6, but derived from a different group of castings and hence leads to different absolute values.

screen area was used, regardless of the sprue type, when the metal was chlorinated. In Table 7, the results of more intensive studies of various sprue systems confirm the trends established in Table 6.

#### Discussion

Three common foundry defects have been seen to be affected by the degree of turbulence permitted in the gating of magnesium-alloy castings, namely; blows, skins, and microporosity. While it is known that factors other than gating turbulence may also effect the occurrence of green sand blows, it is clear that such blows may result from the admixture of the metal with mold gases as the casting is poured—either due to turbulence in the gating, or to turbulence in the casting cavity itself. Skins in the casting result from similar sources. Whether blows or skins will result from a given admixture of metal with mold gases seems to depend on the extent to which the entrapped mold gases react with the metal. The degree of reaction will in turn be dependent upon the temperature of the metal-gas mixture, and upon the length of time during which the metal and mold gases are in intimate admixture at high temperatures.

In the above paragraphs, the relationship between gating design and gating turbulence has been traced in considerable detail. It was shown that significant reduction in the occurrence of blows and skins in test castings accompanied a reduction in the degree of turbulence permitted in the gating of the casting. It was seen that turbulence could be reduced in a number of ways, such as by reducing the pouring rate or by using "non-aspirating" sprues. It was shown that an even more effective way of reducing the occurrence of skins and blows in the casting was the use of a skim-gate, with steel wool, in the runner. As long as our observations were limited to the study of blows and skins, it was indeterminate whether the effect of such screen areas was merely one of filtering out agglomerates of gas and oxide from the metal stream, or whether the function of the screen area was one of preventing the formation of such inclusions. Not until

attention was given to the third defect—microporosity—was any light thrown on this subject.

In studying the effect of gating design on microporosity, it was again established that the occurrence of the defect could be reduced by reducing gating turbulence. And again it was found that an even more potent way of controlling the defect was the use of a skim-gate in the runner.

It is easy to visualize a skim-gate with steel wool acting effectively as a filter for the removal from the metal stream of gas bubbles and the large agglomerates of reaction-products which appear in castings as skins. It is less easy to see how such skim-gates could act to filter out any material, such as a dissolved gas, which would increase the tendency of the metal toward microporosity. It is the belief of the authors that the function of a screen area in the runner of a magnesium alloy casting is not merely one of filtering oxides and gas bubbles out of the metal stream. It is believed rather that a properly designed screen area serves actually to reduce the formation of such inclusions in earlier parts of the gating system. It is believed that the mechanism of this effect is as follows: The screen area, even when amply proportioned so as not to retard the pouring rate of the casting, offers just sufficient resistance to metal flow to keep all elements of the gating system prior to the screen area full of metal. The metal is caused to exert a slight positive pressure on the mold walls surrounding the sprue and runner in all parts of the gating up to the screen area. This prevents aspiration of gases and other sources of metal-gas admixture. By this means, the reaction of the metal with mold gases, with resulting liberation and solution of hydrogen gas, is limited, and by this means does the use of a screen area reduce the tendency of the metal toward microporosity.

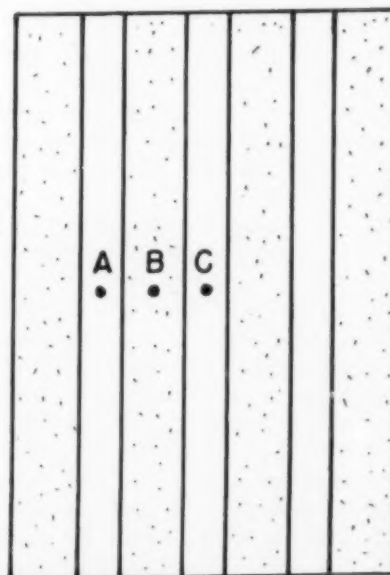
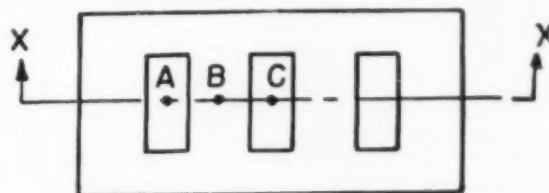
The authors have seen other evidence that a screen area acts in this way, in the study of metal flow in partial molds. For the purposes of flow study, arrangements were made to pour metal into a mold consisting only of a pouring basin, sprue, runner, screen area, and gates, the upper portions of the mold being

mitted. In the pouring of this mold, it was observed that the metal flowed along the runner until it came to the vertical screen area; then, instead of flowing through the screen at once, the metal built up behind the screen until it eventually flowed over it. Had there been a mold section covering the runner, the metal would eventually have been forced to flow through the screen, but not until sufficient hydrostatic head had been built up behind it. Only on this basis has the effect of a screen area in reducing the porosity tendency of the metal been successfully explained.

At the outset of this work, the value of the use of a screen area in the runner in making magnesium alloy castings was questioned. It was a purpose of this work to evaluate the relative merits of such screen areas as compared with gating systems designed to prevent turbulence. The results here presented would appear to be a complete confirmation of the value of such screening systems. For repeatedly, screening has been seen to be the most effective means of eliminating skins and blows, and it is indicated strongly that screening is the most effective way of reducing gating turbulence. On the basis of the present work, it would seem that screen areas—with a skim gate and steel wool—should be used whenever maximum freedom from skins and blows is desired, and that when such screens are used, little further advantage is to be gained by refinements in the design of the sprue. Care, however, must be taken to avoid intermixture of the metal with mold gases after it has passed through the screen. Hence the screen area should if possible be at the level of the lowest part of the casting cavity. Care must also be exerted that the filling rate of the casting cavity is not too fast in relation to its venting capacity.

It is conceivable that cases might arise in which the use of a screen area would be disadvantageous. For example, if extremely close control of the iron content of the metal were desired, the presence of steel in the remelt material would be objectionable. In such cases, the gating principles described above might serve as a guide whereby adequate freedom from turbulence in the gating could be attained without the use of screen areas. These principles, too, would be applicable in general to alloys other than magnesium-base alloys, wherein the use of skim-gates might be out-ruled.

In this regard, a possible pit-fall in the use of the slot sprue is to be pointed out. Referring to Fig. 17, it is easy to visualize the venting problem in the core material at point B, when metal is flowing through channels A and C. At high metal temperatures, the authors have observed evidence of the passage of evolved gases into the metal stream in using such sprues. Figure 18 shows a section of a slot sprue, as



#### SECTION X-X

Fig. 17—Sketch of a slot sprue composed of core sand.

Fig. 18—Section of a slot sprue from a casting poured at 1600 F. Note indications of inter-mixture of metal with gases.



it appeared after solidification, from a casting poured at 1600 F. Another effect of such gas evolution was a retardation of flow rate at high metal temperature. It is to be expected that gas evolution of this sort would lead to the formation of skins in the casting.

#### References

1. J. G. Mezoff and H. E. Elliott, "A Study of Factors Affecting The Pouring Rates of Casting," A.F.A. PREPRINT No. 48-33.
2. R. S. Busk, R. F. Marande and W. C. Newhams, "Effect of Gas on The Properties of Magnesium Sand Casting Alloys," TRANSACTIONS, American Foundrymen's Association, vol. 53, pp. 72-282 (1945).

### APPENDIX A

#### DESCRIPTION OF THE FLUORIDE AND NON-FLUORIDE SANDS USED

Material	Fluoride Sand	Non-fluoride Sand
Base Sand	Bank Sand, 60 to 65 FN	Same
Inhibitor	4% Dow #190 Agent	2.0% $H_3BO_3$ , 1.5% S
Binder	3.5 to 4.0% W. Bentonite	3.0% W. Bentonite
Glycol	1.0% to 1.25%	1.5%
Moisture	2.0 to 2.5% $H_2O$	2.0 to 2.5% $H_2O$

### APPENDIX B

#### METHOD OF ASSIGNING POROSITY RATINGS TO PAN CASTINGS OF TABLE 6

All pan castings were radiographed. Three 15-in. x 17-in. films were required for each pan casting. These three films were termed the "sprue-end", "far-end" or "middle" radiographs, depending on the portion of the casting radiographed in relation to the sprue position.

All films for the group of nine pans listed in Table 6 were arranged serially in order of increasing porosity as follows:

- |                   |                    |
|-------------------|--------------------|
| 1. IX Middle      | 14. II Sprue-end   |
| 2. III Middle     | 15. VI Middle      |
| 3. IX Sprue-end   | 16. VII Far-end    |
| 4. VIII Middle    | 17. II Far-end     |
| 5. IV Middle      | 18. V Sprue-end    |
| 6. VII Middle     | 19. I Sprue-end    |
| 7. I Middle       | 20. I Far-end      |
| 8. IV Sprue-end   | 21. IX Far-end     |
| 9. V Middle       | 22. III Far-end    |
| 10. V Far-end     | 23. VI Far-end     |
| 11. IV Far-end    | 24. III Sprue-end  |
| 12. II Middle     | 25. VIII Far-end   |
| 13. VII Sprue-end | 26. VIII Sprue-end |
| 27. VI Sprue-end  |                    |

Each film was then assigned its serial number as a rating; these "film ratings" were treated as follows:

Order of "Middles"	Order of "Sprue-Ends"	Order of "Far-Ends"
1. IX	3. IX	10. V
2. III	8. IV	11. IV
4. VIII	13. VII	16. VII
5. IV	14. II	17. II
6. VII	18. V	20. I
7. I	19. I	21. IX
9. V	24. III	22. III
12. II	26. VIII	23. VI
15. VI	27. VII	25. VIII
61	152	177

The porosity rating of a pan was then the weighted sum of the ratings of its middle, far-end, and sprue-end films. The "middle" film ratings were weighted three times as heavily as the "sprue-end" or "far-end" ratings, since the sum of all "middle" ratings was only about one-third the sum of the "far-end" or "sprue-end" ratings. Pan ratings for the various systems were, therefore, as follows:

- I. Rating =  $3(7) + 19 + 20 = 60$
- II. Rating =  $3(12) + 14 + 17 = 67$
- III. Rating =  $3(2) + 24 + 22 = 52$ , etc.

### DISCUSSION

Chairman: L. BROWN, Magnesium Fabricators Div., Bohn Aluminum and Brass Corp., Adrian, Mich.

Co-Chairman: A. T. RUPPE, Bendix Products Div., Bendix Aviation Corp., South Bend, Ind.

RICHARD ALLCHIN:<sup>1</sup> In this investigation did the authors assume they had very little back pressure on the down sprue?

MR. ELLIOTT: We did make the assumption that if the runner cross-section is large enough, the flow rate is controlled by the sprue. Under these conditions, it was further assumed that the greatest problem in getting away from turbulence would be in carrying the metal from the top of the mold down to the horizontal runner via the down sprue, and that it would be a much less difficult problem to carry the metal in the horizontal runner around to the various gates into the casting without turbulence. Since the horizontal runner was located at the level of the lowest part of the casting cavity there would be no further drop of metal after it left the sprue. Gravity would help quiet the flow in its upward flow into the casting cavity.

RAYMOND BALL:<sup>2</sup> If the authors used a rectangular-shaped instead of a round sprue would they eliminate some of the metal turbulence?

MR. ELLIOTT: Only if the rectangular-shaped sprue selected were quite thin, say one-half inch or less in thickness. If a 1 5/8-in. diameter sprue were replaced by a 2-in. x 1-in. sprue giving the same cross-sectional area, similar aspiration effects would result in both cases. However, if a battery of four sprues were used, each about 2 x 1/4 in. (an equivalent cross-sectional area) then you would have a system which would permit a greatly reduced amount of aspiration.

DR. BLAKE LORING:<sup>3</sup> I have an observation, not especially with magnesium but with some of the copper-base alloys, which would lend support to the authors' conclusions. A few years ago R. G. Hardy did some experimental work at the Naval Research Laboratory [See "Conditions of Flow in Bronze Castings" by J. T. Robertson and R. G. Hardy, TRANSACTIONS, A.F.A., vol. 54, p. 732 (1946)], in which he was trying to find a suitable means for choking off the bottom of the sprue, to choke the metal off and to suitably lead the metal into the casting. At that time he found there was a very distinct aspirating action, so much so that letting the metal flow out from the sprue with no casting present, the metal, which was a bronze, came out in bursts. It worried us for a long time as to the exact magnitude of the aspirating action. We are glad that this has been quantitatively measured.

CO-CHAIRMAN RUPPE: Why, contrary to general commercial practice, was the test casting gated off the top of the runner?

MR. ELLIOTT: In answer to that question, I think that it is not uncommon to gate magnesium alloys in that way. We could have avoided doing so had we chosen, but our purpose in doing it was so that the first metal that flowed into the casting would not flow into the test panel but instead would flow into the well at the far end of the horizontal runner, and in that way, not until we got steady conditions of flow, set up in the system, would we actually take our sample, that is, would the test casting actually fill.

<sup>1</sup> Thompson Products, Inc., Cleveland

<sup>2</sup> Ball Brass Co., Tacoma, Wash.

<sup>3</sup> U. S. Naval Research Laboratory, Washington, D. C.



CO-CHAIRMAN RUPPE: We assume there is and take steps to adjust our gating systems to minimize the channeling of the molten metal up into the mold cavity. I wonder if there is a channeling effect through the screen block? In other words, I wonder how effective the enlarged screen block area really is when passing metal to the mold cavity. I believe the first metal to reach the screen spreads out to form a block and actually only the center portion of the screen passes metal. If you have ever noticed metal rising up through a screen you will note that it does channel. The metal will come through at a point and not uniformly through the whole screen.

MR. ELLIOTT: I think it is very unlikely that all this screen area is effective. In other words, metal does not flow through all the holes in the screen throughout the pour. In fact, I think that is one of the features that contributes to the functioning of a vertical screen in the horizontal runner. A screen not only filters out skins and bubbles that have already formed due to turbulence, but it tends to hold the metal back in the entire system up to that point. It tends, in other words, to put a restriction in the running system; even though the screen is amply proportioned, it will offer just enough restriction of the flow to support the metal in the sprue, and thus helps to avoid aspiration effects. I think that the screen area not only acts as a filter to strain out oxide and bubbles, but it helps to prevent the formation of those defects in prior parts

of the system.

MR. MEZOFF: We have opened molds and watched the metal as it came to a vertical screen area. The very first metal instead of going immediately through the screen would stop at the screen, that is in a horizontal runner, and the level of the metal in the channel would rise until the runner was full and then flow over the screen. When we had a cope on, the metal would come up and hit the cope surface and then pass through the screen. With the cope on we could not see whether it went through in the whole area, but I would assume that the metal passed through the screen uniformly all over, at least at the start of passage through the screen. I have seen this type of flow at screens in permanent molds also.

CHAIRMAN BROWN: Did the authors find in their investigation that a certain proportion between the sprues, runner gates and end gates had to be maintained?

MR. ELLIOTT: We did not actually make a study of that.

CHAIRMAN BROWN: Were you concerned only with the aspirating gases?

MR. ELLIOTT: Yes. In the system that we used we did not worry about the yield of the casting. We made the runner so large in relation to any sprue that we were going to test that we felt that turbulence effects in the horizontal runner would be of little significance in relation to those occurring in the down sprue.

# OXYGEN-ENRICHED CUPOLA BLASTS

By

W. C. Wick \*

## ABSTRACT

Use of oxygen-enriched cupola blasts has proven successful for increasing melting rates and tapping temperatures with a given coke ratio and blast volume. Melting rates have been increased as much as 40 per cent by adding approximately 14 per cent oxygen to the blast. Melting and tapping temperature has been raised from 25 to 150 F by manipulating the oxygen addition and total blast volume. Savings in coke with no sacrifice in operating conditions are available by the use of oxygen enrichment. Most important of all, oxygen is useful to overcome bridging and cold metal conditions and often can be used to save a heat from freezing in the furnace rather than resorting to dropping the bottom. Oxygen enrichment can be used as a control tool to regulate melting rate and tapping temperature.

FOR THE PAST FEW YEARS the steel industry in cooperation with several oxygen producing companies has been investigating the possibility of using commercial oxygen to facilitate melt down and quality control in steel making processes. These experiments apparently are showing encouraging results in the form of increased production capacity by faster melting time and more rapid chemical refining.

New and diversified applications for commercial oxygen in large volume are rapidly coming into existence. One of these new applications is in conjunction with cupola operation in the foundry industry. A number of scattered verbal reports on the use of high purity commercial oxygen as an addition to the cupola blast air have been heard, and brief mention of it has been made in trade publications.

Realizing the importance of the subject, a research project was initiated to determine the advantages or disadvantages that would be obtained from using oxygen-enriched air blasts in the cupola, and to obtain experimental data which would be presented to the foundry industry as an indication of the results that could be expected in commercial operations. In order to obtain such data, a number of heats had to be made with as many constant conditions as possible, using the oxygen content of the air blast and the coke ratio as the controllable variables.

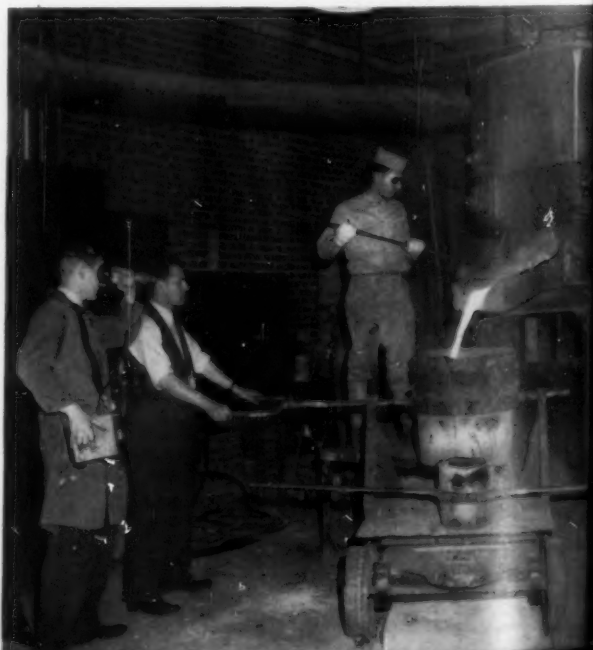
The following presentation is offered as a progress

report of the work completed on the use of oxygen-enriched cupola blasts. A number of results are indicated, all of which undoubtedly will be subject to some degree of change when applied to larger scale operation.

Equipment used in the experiments consisted of commercially available items of standard types and design. Figure 1 shows the cupola in operation, tapping an oxygen-enriched blast heat. Because of the nature of the work, a large volume of oxygen (for the proposed program) was required to be available for instant use. Figure 2 shows a small oxygen generating plant which was made available for this research project. This oxygen generator has a rated capacity of 1000 free cu ft per hr of high purity oxygen. A bank of eight hydrotubes with a total capacity equivalent to 32,000 free cu ft was used as a source of oxygen supply for the experimental cupola heats.

The flow of oxygen from the hydrotubes to the cupola air blast line was controlled through a pressure reducing station. The volume of oxygen added to the blast was measured by an orifice and manometer system. The blast and oxygen lines and control equipment used in the project are shown in Fig. 3.

Fig. 1—Tapping cupola. Oxygen-enriched blast used.



\* Associate Metallurgist, Armour Research Foundation, Chicago.



Fig. 2—Oxygen generating plant used in this project.

A standard type cupola, lined to 19 in. inside diameter, with a skip hoist charging unit was used in the experiments. The air blast entering the cupola was controlled by an air weight control assembly which operated independently from the oxygen pressure reducing station. In order to obtain a thorough mixture of the high purity oxygen addition with the air blast, a 3-in. diameter oxygen line was connected to the 10-in. diameter blast line downstream from the air control assembly and at right angles to the air blast line. So far as could be determined by gas analysis made of samples taken at the wind box, the blast air and oxygen were thoroughly mixed before entering the tuyeres.

Consistency and repetition of a number of procedures from one heat to the other was highly desirable to eliminate as many variable conditions as possible from the tests.

#### Cupola Operation

Charging practices were standardized for all heats. Two hundred-pound metal charges consisting of 50 per cent pig iron, 40 per cent iron foundry returns and 10 per cent steel flashings were used. When a change in iron to coke ratio was made, the quantity of coke was varied. This represented the smallest change in stock conditions and metal weight adjustments. Raw materials for the experiments were obtained from the same source for all the experiments; however, slight variations in coke and metal quality could not be avoided because of present day material shortages.

Coke for these heats was sized to range from 2 to 4-in. diameter pieces. Limestone was used as the fluxing material in all but two heats and constituted 32 per cent by weight of the coke charge. The air blast

was drawn from inside the foundry which maintained a substantially constant temperature range. The oxygen used was of 99.5 to 99.6 per cent purity and free of moisture. Temperature measurements were made inside the cupola with a radiamatic pyrometer sighted into a closed-end ceramic tube immersed in the cupola well. Optical pyrometer temperature readings were taken at the cupola spout throughout each heat. These temperatures form the basis for the data and the results presented.

#### Research Program and Results Obtained

In order to have a basis for comparison, the initial heats Nos. 1 and 2 were made with a normal oxygen content air blast. Heat No. 2 (Table 1) is considered to be an average operation for this cupola with an 8:1 coke ratio and normal volume air blast of 850 cfm. The melting rate, melting loss and level off temperature were used for comparison with those of the heats made later.

Table 1 is a tabulation of data obtained from individual heats made with varying oxygen additions to the air blast. A series of 8:1 coke ratio heats with oxygen additions to the blast ranging from 0 to 13.8 per cent of the total blast volume show almost a straight line increase in the melting rate. Figure 4 illustrates this increase in melting rate graphically and serves to demonstrate one of the effects of oxygen-enriched cupola blasts. The melting rate was increased as much as 40 per cent in one case. The total blast volume was higher in this heat, which probably had some effect on the melting rate. One of the most satisfactory oxygen-enriched blast heats was No. 4, Table 1, in which the melting rate was increased 19 per cent above the blank heat (No. 2). If the "level off"\* tem-

\* Level off tapping temperature is the average temperature obtained after the coke bed is burned down to the operating level and the temperature remains nearly uniform.



perature is noted for the various 8:1 coke ratio heats tabulated, a temperature rise will be indicated corresponding to the increase in oxygen content of the blast. Still higher tapping temperatures can be obtained with an 8:1 coke ratio if the total blast volume is reduced 15 to 20 per cent and the oxygen content maintained at approximately a 10 per cent addition. Under these conditions the melting rate will not be increased as much as with the normal blast volume heats with equivalent oxygen additions. Table 1, heat No. 13, shows the figures derived from a reduced volume, oxygen-enriched blast heat. The melting rate, however, is still 5.9 per cent above the blank heat. This combination of conditions should be appealing to a shop interested in higher tapping temperature rather than to one desiring greater production.

#### Oxidation Losses

Contrary to expectations, oxidation losses diminished with increased oxygen content blasts. If the melting loss and oxygen content of the blast are plotted, the relationship is essentially a straight line function, the melting loss being reduced as the oxygen content of the blast increases. In an effort to determine why the melting loss diminished, metal and slag samples were analyzed. The metal analyses generally indicated silicon content increases rather than the normal 10 per cent loss. This is brought about by increased quantities of CO in the combustion reactions and the formation of slightly reducing conditions shown by flue gas analyses. The results of these

conditions are to eliminate the oxidation loss of silicon and also to reduce some of the silica lining to silicon.

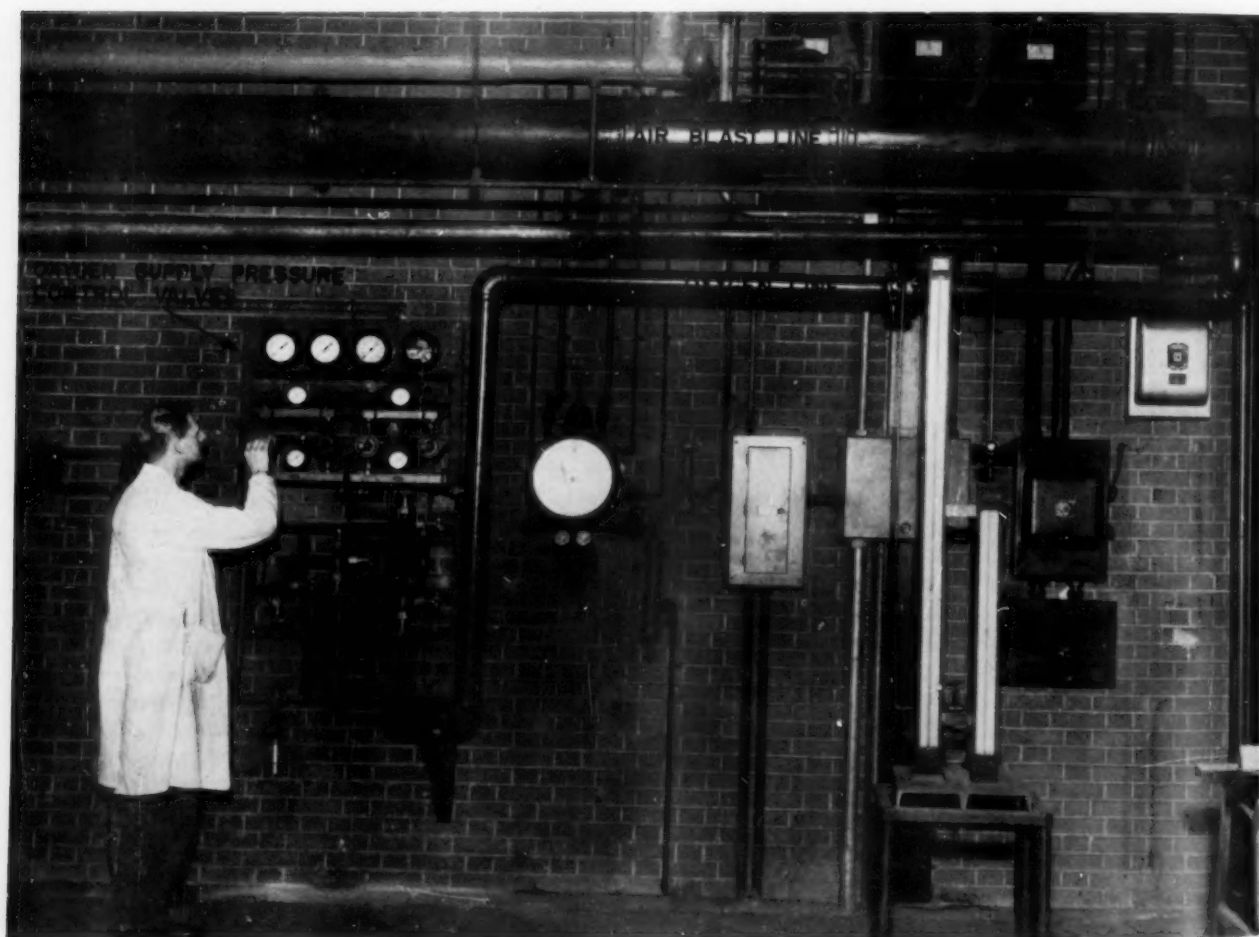
Further support of the reduction in melting loss with increased oxygen blast heats is shown in Table 2. Slag analysis of samples taken from the heats Nos. 2, 3 and 4 indicated reductions in manganese, manganese dioxide, and iron oxide contents as the oxygen of the blast is increased. A reduction of these elements in the slag reflects the increased yield of metal charged. Heats Nos. 6 and 13 show the same trend for the iron oxide but because of other changes in the heat the other elements are slightly out of line with the indicated trend.

Referring again to Table 1, the effect of oxygen enrichment and reduced volume blasts with a 12:1 coke ratio is shown. By a simple change in the total blast volume, maintaining the same percentage of oxygen addition, the melting rate can be reduced and the tapping temperature increased. This method of control offers a greater flexibility in cupola operation.

#### Effect of Oxygen Content and Coke Ratio

Several heats using a 5:1 coke ratio and normal oxygen content air blast were made for comparative purposes with the 7:1 and 8:1 coke ratio enriched blast heats. Operations with the 5:1 ratio heat (No. 11) were very unsatisfactory due to the high ash con-

Fig. 3—Blast and oxygen lines, and control equipment.



Heat  
No.

2  
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4

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12

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15

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TABLE 1—COMPARATIVE DATA, NORMAL AND OXYGEN ENRICHED BLAST HEATS

Heat No.	Melting Time (min)	Melting Rate (lb/hr)	Coke Ratio	Total Blast Volume (cfm)	Oxygen Addition (% of Blast)	Level Off Average Temp. (°F)	Metal Charged (lb)	Metal Melted (lb)	Metal Re-Covered (lb)	Melting Loss (%)	Increased Melting Rate (%)	Remarks
2	115	2922	8-1	850	0	2675	5800	5600	5250	6.3	0	
3	113	3186	8-1	860	4.5	2700	6000	6000	5713	4.8	9.0	
4	100	3600	8-1	850	8.6	2775	6000	6000	5775	3.8	19.4	
6	82	4098	8-1	892	13.8		6000	5600	5596	0.07	40.2	Booster Charge of Coke Added During Heat—No Level Off
13	103	3096	8-1	700	9.9	2860	6000	5317	5191	2.3	5.9	
7	95	3789	12-1	817	8.2	2650	6000	6000	5825	2.9		
14	120	3200	12-1	550	8.1	2800	6400	6400	6362	0.6		
8	119	3000	7-1	850	8.7	2810	6000	5921	5835	1.5		
12	179	1836	5-1	850	0	2600	6000	5486	5389	1.6		
11	147	1956	5-1	850	0		4800	4800				Oxygen Used to Save Heat
				750	8.0							
				700	9.4							
15	104	3228	8-1	850	0	2790	5600	5600	5473	2.2	10.7	
				670	7.1							

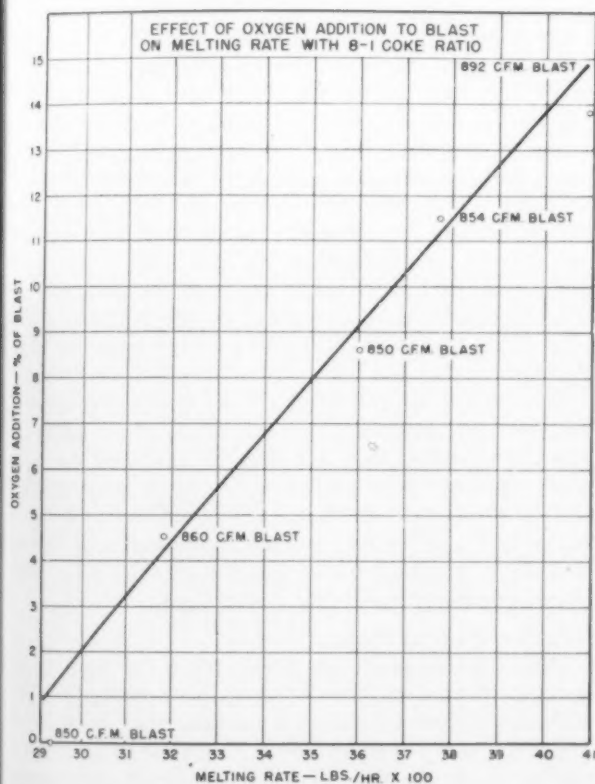


Fig. 4—Effect of oxygen addition to blast on melting rate with 8:1 coke ratio.

tent of the coke. A study of Table 1 will reveal that higher temperatures and faster melting rates were obtained with a 12:1 coke ratio oxygen-enriched blast than was possible with a 5:1 ratio normal volume air blast. Savings of fuel (and even salvage of a complete heat in some instances) can be effected without a sacrifice in vital temperature requirements. The amount of fuel saved would be the difference between the 5:1 and 12:1 coke ratios or approximately 233 lb of coke per ton of metal melted. The increased temperature and melting capacity of these two heats must likewise be considered when weighing the economics of each operation.

Comparison of heats Nos. 7 and 14, Table 1, indicates similar temperature and melting rate results. Economically, however, heat No. 14 would be more desirable than No. 7 because of the smaller quantity of coke and oxygen used and decreased melting loss for a comparable tapping temperature and melting rate.

Another economic aspect to be considered when thinking of the oxygen-enriched blast is the effect it might have on the cupola refractory. A quantitative study of lining loss during each heat was made and the results are shown in Table 3. The patch material used in all heats was as follows: 30 per cent— $\frac{1}{8}$ -in. silica quartz ganister; 50 per cent— $\frac{3}{8}$ -in. silica quartz ganister; 14 per cent—No. 60 silica sand; 6 per cent—western bentonite; 5 per cent water.

The data given represent the quantity of patch

TABLE 2—SLAG ANALYSIS FROM 8:1 COKE RATIO HEATS

Heat No.	Oxygen Addition (% of Blast)	Si	SiO <sub>2</sub>	Mn	MnO <sub>2</sub>	FeO	Remarks
2	0	22.0	47.1	1.53	2.41	6.3	
3	4.5		47.4	1.43	2.26	4.4	
4	8.6		53.8	1.32	2.08	3.0	
		23.24	49.7	0.86	1.36	3.87	
6	13.8	23.00	49.2	1.45	2.29	6.10	Excessive O <sub>2</sub> for Coke Used
		20.92	44.78	1.90	3.00	1.47	Reduced
13	9.9	21.46	45.93	2.11	3.34	1.74	Volume Blast

material of a constant quality and mixture required to restore the inside diameter of the cupola to 19 in. Dimensionally speaking, the oxygen-enriched blast can be considered less destructive to the lining than the normal blast. However, the zone of maximum temperature is extended or elongated, which also means the "scar" or burned out area is increased accordingly. This condition accounts for the apparent added lining loss in the oxygen-enriched heats. It is possible that, due to the slight decrease in dimensional loss of the lining, longer heats could be run if desirable. Close observation of the quantities of patch material listed in Table 3 will reveal that the reduced blast volume heat No. 13, (oxygen enriched) caused less lining damage dimensionally and volumetrically than the normal oxygen blast heats. Heat No. 14, also of reduced volume enriched blast, was slightly more destructive to the lining than the blank heat. Heat No. 15 was a mixed blast heat which was changed from normal to oxygen-enriched blasts intermittently during the heat. This procedure was not particularly harmful to the lining and did show interesting effects of periodic usage of oxygen-enriched blasts for controlling temperature.

Mere tabulation of the data from these heats is insufficient to emphasize the importance of the developments which are possible by the use of oxygen-enriched blasts. The more interesting and valuable aspects of this research program are brought out by a brief discussion of the various figures which are shown herewith.

#### Temperature Control With Oxygen and Coke Ratios Required

Figure 5 represents tapping temperature curves from Heats Nos. 2, 3, and 4. The value of oxygen enrichment in the raising of tapping temperatures is illustrated. Using normal blast volumes, an oxygen addition of 8.6 per cent was sufficient to increase the metal temperature at the spout 100 F. This amount of added temperature would be helpful in eliminating misrun defects if such conditions prevailed.

If little or no extra temperature is required, oxygen enrichment will be effective in reducing the quantity of coke required per ton of metal melted. Figure 6 shows the similarity of results that can be obtained by the use of a 12:1 coke ratio and an 8.2 per cent oxygen addition to the blast as compared with a heat of 8:1 coke ratio and normal oxygen content air blast.

Another method of temperature control is demonstrated in Fig. 7. Manipulation of the coke ratio with approximately the same total blast volume and oxygen addition will result in raised or lowered tapping temperatures. This condition, if known to the cupola operator, could be used to great advantage in producing iron at the most suitable temperature for the type of work to be poured. Greater economy of coke consumption can be obtained if oxygen is used to control the tapping temperature.

#### "Oxygen as Cupola Medicine"

Figure 8 emphasizes the importance and value of oxygen-enriched blasts when a heat gets off to a poor start or when below average fuel is encountered unexpectedly. Heat No. 11 illustrates the action of a 5:1 coke ratio normal oxygen content blast heat which

TABLE 3—LINING MATERIAL REQUIRED AFTER VARIOUS HEATS

Heat No.	Lining Diameter After Heat (Inches)	Patch Material Required (Pounds)	Coke Ratio	Oxygen Addition (% of Blast)
2	25	210	8-1	0
3	25	245	8-1	4.5
4	25½	275	8-1	8.6
6		290	8-1	13.8
7	24	275	12-1	8.2
8	22	225	7-1	8.7
*11		350	5-1	0
*12	27	490	5-1	0
13	22	90	8-1	9.9
14	22	225	12-1	8.1
15	23	215	8-1	Variable

\* Soda Ash + Limestone Flux.

started out under average conditions. About half way through the heat, the temperature started to drop and severe bridging occurred. The heat was losing temperature steadily and under ordinary circumstances the bottom would have been dropped, thereby losing the balance of the day's run. Knowing the potentialities of oxygen, the blast was enriched and almost immediately the temperature increased. The balance of the heat was completed without additional difficulty. The coke analysis indicated a high ash content which was blamed for most of the trouble encountered. Another brand of coke with a lower ash and higher carbon content was used for heat No. 12 and the results were somewhat better with the same coke ratio and normal blast as that used in heat No. 11.

The importance of these results cannot be stressed too strongly, especially to the small shop operator who occasionally may get into trouble and have to "drop bottom." For such cases as these, the economy and advantages of having a few cylinders of oxygen available for emergency use is immeasurable. The loss of a day's production can mean many times the cost of ten or twenty small tanks of oxygen.

#### Reduced Volume Blasts

Figure 9 compares the results of a 7:1 coke ratio, normal volume oxygen-enriched blast heat (No. 8) with two reduced volume oxygen-enriched blast heats (Nos. 13 and 14) of 8:1 and 12:1 coke ratio, respectively. Aside from a slight bridging of one tuyere during heat No. 14, which caused a temperature variation, approximately the same accomplishment was possible with a 12:1 coke ratio reduced volume enriched blast heat as with a 7:1 coke ratio normal volume enriched blast heat. The 8:1 coke ratio heat was slightly hotter and more uniform in tapping temperature than either of the other two heats. The significant feature of this particular illustration is the demonstration of the ability to produce hotter iron with less coke using reduced volume oxygen-enriched blasts than is possible with normal volume blasts even with added oxygen. The only important variable that could cause such a variation in operating conditions is the quantity of inert nitrogen entering the cupola. Nitrogen adds nothing to the combustion of



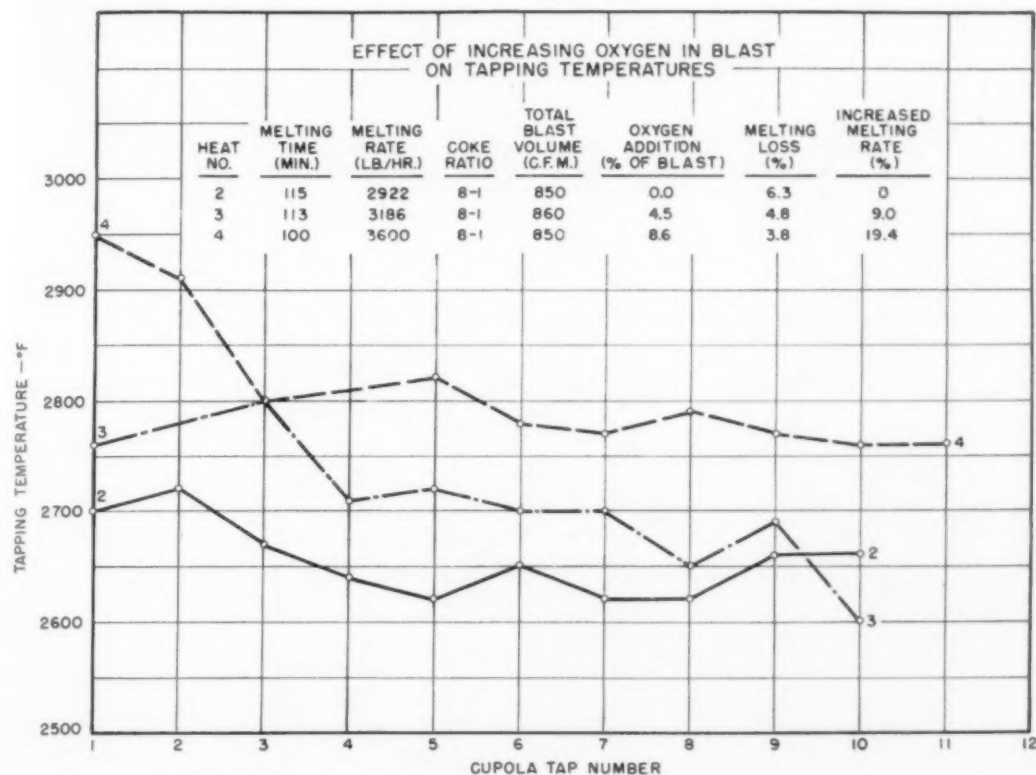


Fig. 5—Effect of increasing oxygen in blast on tapping temperatures.

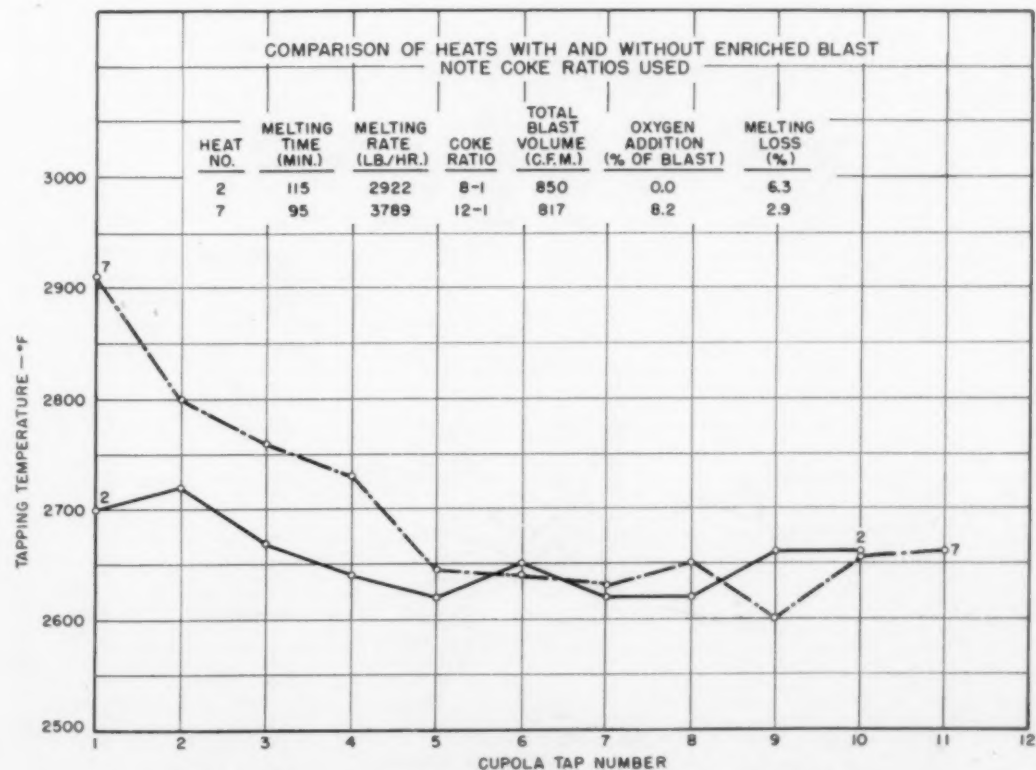


Fig. 6—Comparison of heats with and without enriched blast.

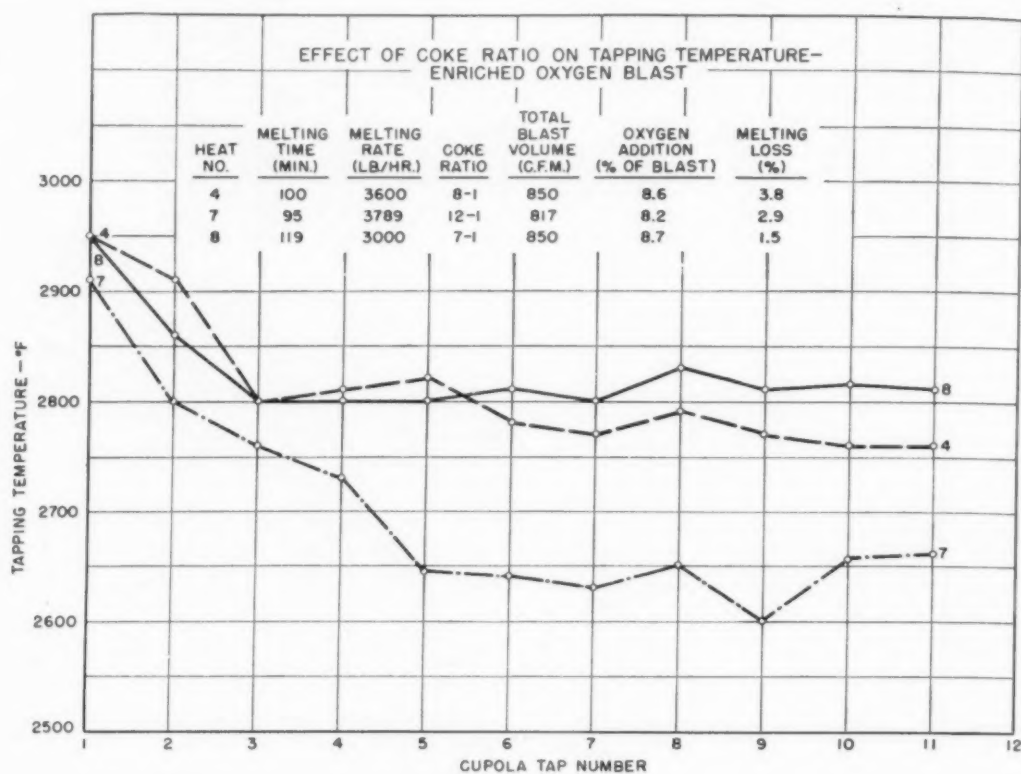


Fig. 7—Effect of coke ratio on tapping temperature—enriched oxygen blast.

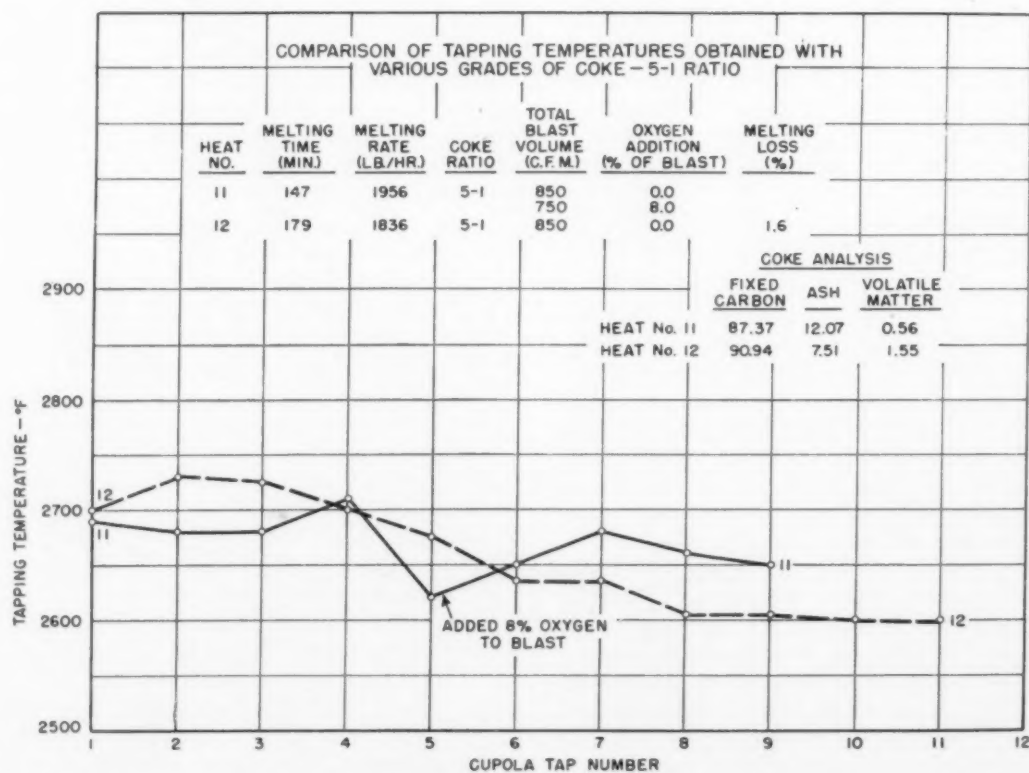


Fig. 8—Comparison of tapping temperatures obtained with various grades of coke—5:1 ratio.

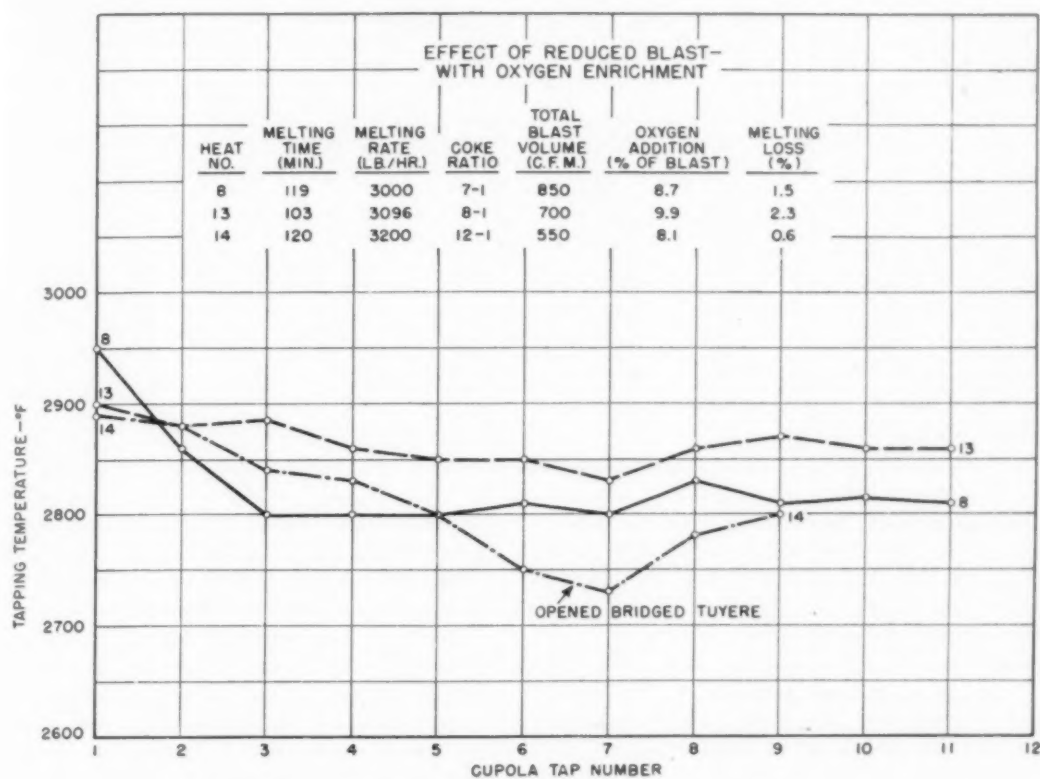


Fig. 9—Effect of reduced blast with oxygen enrichment.

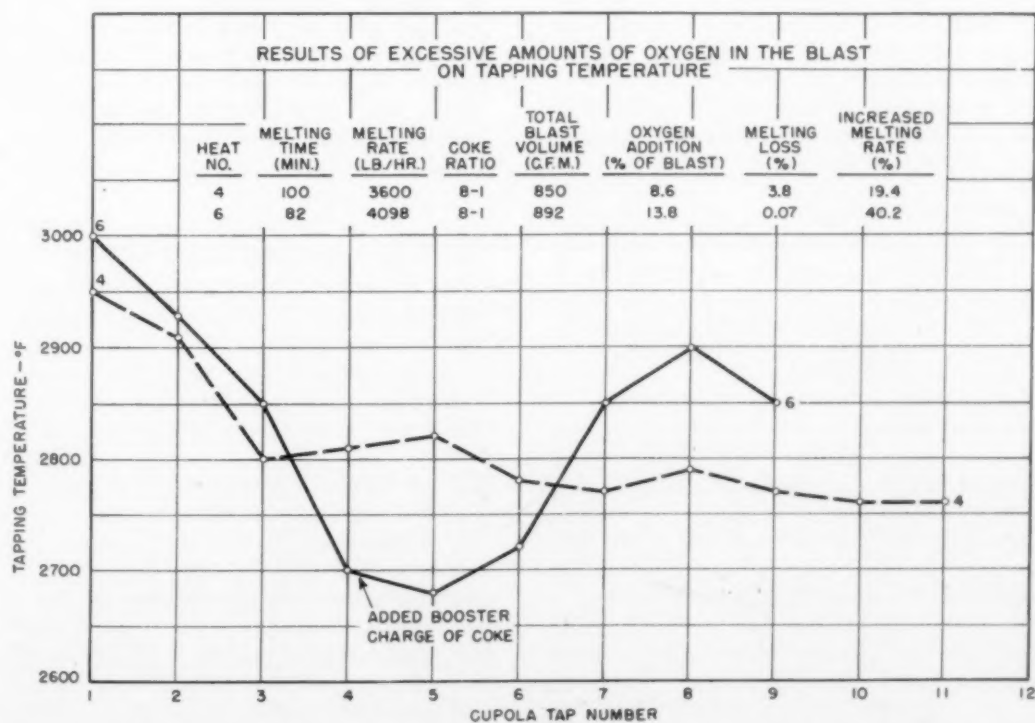


Fig. 10—Results of excessive amounts of oxygen in the blast on tapping temperature.



TABLE 4—OXYGEN ADDITIONS VS. INCREASED MELTING RATES

Cupola Size	Total Blast C.F.M.	Oxygen Addition Cu Ft/Hr Percent of Blast				Normal Melting Rate (*) Tons/Hr	Melting Rate — Tons/Hr With Oxygen Addition Percent of Blast				(*)
		4	6	8	10		4	6	8	10	
1	950	2,280	3,420	4,560	5,700	1.5	1.63 <sup>x</sup>	1.71 <sup>x</sup>	1.78 <sup>x</sup>	1.86 <sup>x</sup>	
2	1,300	3,120	4,380	6,240	7,800	2.25	2.44	2.56	2.66	2.78	
3	1,800	4,320	6,480	8,650	10,800	3.25	3.54	3.70	3.86	4.03	
3½	2,425	5,810	8,760	11,600	14,500	4.25	4.63	4.83	5.05	5.27	
4	3,100	7,450	11,100	14,850	18,550	5.50	5.99	6.28	6.5	6.80	
5	3,600	8,650	12,920	17,500	21,600	6.25	6.8	7.12	7.45	7.75	
6	4,100	9,840	14,720	19,600	24,600	7.25	7.9	8.26	8.6	9.0	
7	5,200	12,450	18,650	24,900	31,200	9.25	10.0	10.5	11.0	11.5	
8	6,400	15,350	23,000	30,700	38,400	11.25	12.2	12.8	13.3	13.9	
9	7,700	18,450	27,600	36,850	46,200	13.75	15.3	15.6	16.3	17.0	
9½	9,200	22,050	33,100	44,100	55,100	16.25	17.7	18.5	19.3	20.0	
10	10,700	25,600	38,400	51,400	64,250	19.0	21.1	21.6	22.6	23.5	
12	12,500	30,000	45,000	60,000	75,000	22.25	24.2	25.3	26.4	27.1	

\* Based on 8-1 coke ratio.

x Estimated from experimental heats.

the coke to produce heat, but does add an extra volume of gas which must necessarily be heated as it passes through the cupola. The Btu value developed by combustion is partially used to heat this nitrogen, thereby reducing the amount of heat available to melt and superheat the iron.

#### Excessive Oxygen Addition

The amount of oxygen added to the blast may be in excess of that which can be used satisfactorily for the quantity of coke present in the bed and subsequent splits. A typical example of such a condition was produced in heat No. 6 shown graphically (Fig. 10) in comparison with a satisfactory heat of 8.6 per cent oxygen addition. Both heats represented were made with the same coke ratio (8:1), and approximately the same total blast volume.

An oxygen addition of 13.8 per cent was sufficient to burn the coke faster than melting could proceed; therefore, the coke bed was burned down with an

accompanying loss of temperature. Later in the heat, a booster charge of coke was added and the temperature increased until this coke also was burned away. A very high temperature would have resulted if sufficient coke was supplied for the quantity of oxygen present in the blast.

To use oxygen in the blast throughout the entire heat might be considered unnecessary or too costly by some operators. An alternate method of using oxygen that is worthy of consideration is the intermittent "shot in the arm" procedure. Figure 11 shows the effect of periodic additions of oxygen to the blast on tapping temperature. The first part of the heat was run with the "shot in the arm" method and the last part was run using oxygen primarily as a control tool to obtain and maintain a given temperature. The results are self evident of the value of oxygen as a control tool and for quick temperature pick-up to avoid or eliminate bridging or cold iron difficulties.

TABLE 5—COST OF OXYGEN ADDITIONS/TON OF MELT

Cu Ft of Oxygen Per Mo. (20 Days)	Tons of Oxygen Per Mo. (20 Days)	Unit Cost of Oxygen Dollars/Ton	Unit Cost of Oxygen Cents/100 Cu Ft	Dollars Cost/Ton of Melt Per Cent Addition			
				4	6	8	10
Small Quantity Purchase		180.00	75.0	9.40	13.30	17.20	20.60
Small Quantity Purchase		120.00	50.0	6.25	8.90	11.50	13.70
600,000	25.0	72.00	30.0	3.75	5.35	6.90	8.25
1,200,000	50.0	62.40	26.0	3.25	4.63	5.97	7.16
1,800,000	75.0	55.20	23.0	2.88	4.10	5.30	6.35
3,840,000	160	45.60	19.5	2.44	3.47	4.49	5.35
4,800,000	200	44.40	18.5	2.32	3.29	4.26	5.10
5,760,000	240	43.20	18.0	2.25	3.20	4.14	4.96
6,720,000	280	42.00	17.5	2.19	3.12	4.02	4.82
7,680,000	320	40.80	17.0	2.12	3.02	3.92	4.69
8,640,000	360	39.60	16.5	2.06	2.94	3.80	4.54
9,600,000	400	38.40	16.0	2.00	2.84	3.68	4.41
10,560,000	440	37.20	15.5	1.94	2.76	3.56	4.27
11,520,000	480	37.20	15.5	1.94	2.76	3.56	4.27
15,360,000	640	36.00	15.0	1.87	2.66	3.45	4.14
19,200,000	800	34.80	14.5	1.81	2.58	3.34	4.00
23,040,000	960	33.60	14.0	1.75	2.50	3.22	3.86
Very Large Quantity Purchase		15.00	06.2	0.78	1.11	1.44	1.72
Huge Quantity Purchase		5.00	02.1	0.25	0.36	0.46	0.55

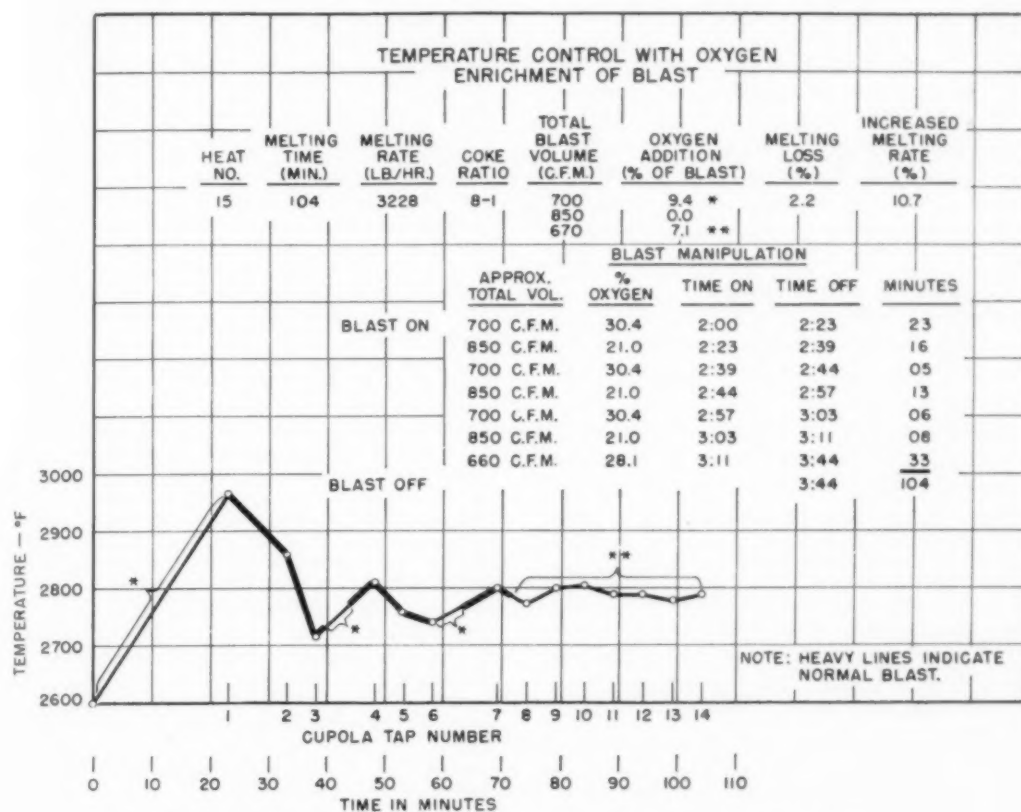


Fig. 11—Temperature control with oxygen enrichment of blast.

Probably the most vital consideration, and naturally the first thought, concerns the economics of the use of oxygen-enriched blasts. The question of the cost of large volume purchases of oxygen has many answers, depending on how much is required, the purity and the method of obtaining it.

#### Cost of Oxygen Addition to Blast

Since the advantages gained from the use of oxygen in the cupola blast are somewhat intangible economically (fewer misrun castings, improved furnace operation, higher melting rate, less oxidation loss, higher pouring temperatures, etc.), the real value in dollars and cents would be difficult, if not impossible, to evaluate unless savings were calculated over a prolonged period of time.

Table 4 consists of a number of assumed melting rates for various size cupolas using data from the experimental heats as a basis for the calculations.

The melting rate per unit of time will influence the cost per ton of melt when using oxygen in the blast. In normal blast heats with approximately 4, 6, 8, and 10 per cent oxygen additions to the blast, the melting rates were increased by roughly 9, 14, 19, and 24 per cent, respectively. Using these approximations derived from the experimental heats, a table (No. 4) of oxygen consumption per hour and melting rate per hour was prepared for various sized cupolas.

Table 5 shows approximate costs of quantity oxygen at various levels of consumption per month as

supplied by one of the leading makers of oxygen production equipment. The cost of using oxygen per ton of melt at four levels of addition are indicated for the unit cost represented. These costs are only approximations and will vary according to the source of oxygen supply.

The real cost of oxygen-enriched cupola blasts as stated previously will depend on the amount of help it will offer to the particular foundry, and it is believed may be more than balanced by economies in operation, if effectively used.

#### Conclusions

The following conclusions have been drawn from the experimental data obtained from the use of oxygen additions to the cupola blast.

1. With approximately a 10 per cent oxygen addition to the normal volume cupola blast and an 8:1 coke ratio, the melting rate was increased roughly 20 per cent.

2. Under the above conditions the tapping temperature of the metal was approximately 100 F higher than with a normal oxygen content blast. Higher operating temperatures also provided less bridging and more fluid slag.

3. By control of the oxygen content and total volume of the blast, tapping temperature and melting rate for a given coke ratio can be controlled to suit the prevailing conditions in cupola operation.

4. Oxygen additions of more than 10 per cent of

the total blast require a low coke ratio and would result in high tapping temperatures.

5. Temperature control by adjustment of the coke ratio is possible with oxygen-enriched blasts.

6. Operating conditions and tapping temperatures similar to a 5:1 coke ratio normal blast heat can be obtained by increasing the iron to coke ratio to 12:1 and adding from 8 to 9 per cent of oxygen to the blast.

7. Reduced volume oxygen-enriched cupola blasts caused a smaller increase in melting rate and higher tapping temperature than the normal volume oxygen-enriched blast heats. This offers an additional tool for temperature and production control.

8. Periodic "shot in the arm" methods of using oxygen additions to the blast can be beneficial in obtaining hot iron on short notice without serious interruption of the charging routine.

9. Cupola heats that start to freeze up due to some operating difficulty, (poor coke, bridging, etc.), can be saved from "dropping bottom" by the emergency use of bottled oxygen added into the blast line. The temperature can be increased almost immediately and the heat saved, not to mention the day's production.

10. By proper manipulation and use of oxygen-enriched blasts, little additional refractory loss occurred and improved operating conditions resulted.

11. Less oxidation loss of the silicon occurred with the use of oxygen-enriched blasts.

12. Dollars and cents cost of oxygen additions will range from \$0.25 to \$20.00 per ton of melt depending upon quantity added and price paid per 100 cu ft. The true cost or economy of oxygen-enriched cupola blasts will have to be determined by each shop after all conditions of operation are considered.

### Acknowledgments

The author wishes to express his appreciation to the Whiting Corporation, Harvey, Illinois, for sponsoring the research work, and to Air Products Incorporated, Emmaus, Pennsylvania, for the use of the oxygen generating unit. Special acknowledgment and appreciation goes to A. W. Gregg, P. Poelstra, W. Jaeschke, and A. J. Grindle of the Whiting Corporation, Harvey, Illinois; and to W. E. Mahin, R. D. Haworth, C. Locke, M. Baeyertz, and other staff members of Armour Research Foundation for their interest and assistance in the completion of the reported work.

### DISCUSSION

*Chairman:* R. G. McELWEE, Vanadium Corp. of America, Detroit

*Co-Chairman:* E. C. JETER, Ford Motor Co., Dearborn, Mich.

A. K. HIGGINS (*Written Discussion*):<sup>1</sup> The first paper on a new subject of any complexity is always a difficult one. This is particularly true of a subject as involved as the one presented by Mr. Wick. He is to be congratulated on his carefully conducted experiments and his able presentation of the results of his work. When an industrial process has been stabilized for as long a period as has the operation of melting iron in the cupola, we tend to feel that major changes in methods are of doubtful value, and are reluctant to accept modifications in practice that are beyond our normal experience.

Modifications in cupola practice have been confined (for many years) to improvements in control, and the introduction of labor saving devices.

The introduction of added oxygen to the cupola blast may

be said to go beyond this type of change, involving as it does, variations in the chemical behavior of the cupola, as well as in the physical and metallurgical aspects of its use.

We have, in recent months, been conducting limited explorations in the use of oxygen on commercial melts, and our results differ in degree but not in trend, with those presented by Mr. Wick. The single exception to this statement, being the effect of the cupola refractories.

It must be understood that our work was done far more crudely, as is usually the case when development work is done on shop equipment. We did not meter our oxygen as precisely, nor did we have the control on blast volume that was available to Mr. Wick. We were somewhat in the position of making a run under approximate conditions and finding out what we had done after we were through.

Under these conditions, we feel that the measure of agreement that resulted between Mr. Wick's work and our own is generally good.

Lack of agreement in magnitude, is probably due to the differences in dimension of the equipment used.

Some of you may have seen a description of our work, published in *Iron Age* (April 22, 1948) and if so you will note that we have reported approximate doubling of the rate of melting when using 5 per cent oxygen on a run with a 3.75 to 1 coke ratio. Another run, at a 6 to 1 coke ratio with approximately 5 per cent added oxygen resulted in a melting rate slightly more than double the estimated rate for a cupola 37 in. ID as reported in Mr. Wick's Table 4 (for an 8 to 1 coke ratio). We do not believe that this second run melted at anything like its best capacity because we were not prepared for the extremely high rate attained and could not keep the cupola charged.

This difference in results is presumably due to the difference in proportion (between the relationship of cross section area to wall area) of the two cupolas used, and the consequent lower efficiency of the smaller cupola due to heat losses into and through the wall. As a consequence, our outlook on increased melting rate is considerably more optimistic than the results reported in this paper.

Mr. Wick has touched briefly on the lower loss of silicon in his melt analysis. In the case of the extremely low silicon melts we were using, this became a positive gain (in one case of 0.60 per cent). Again we would ascribe this to the higher temperatures reached in our larger operation. The unfortunate thing is that the silicon content seems to be largely a function of coke temperature and as a result, uniformity of operating conditions would have to be maintained very carefully if we wished to keep our metal analysis at all constant. This would make periodic additions of oxygen difficult and would indicate the continual additions of smaller amounts of oxygen as more desirable.

Melting temperatures reported by Mr. Wick are in substantial agreement with our results for the initial tap, but with normal charging, we found general maintenance of temperatures in excess of 2800 F. This would presume a reduction in bed height in Mr. Wick's operations and might mean that it would be as well to start with a somewhat lower bed.

The main difference in results is of course in the amount of refractory lost during the operation and its resultant effect on the slag produced.

We found that the cupola lining (made of first grade cupola block) was almost melted through in 2 hr although we have no difficulty in running all day, without added oxygen. This factor alone would prevent the continued use of oxygen in the amounts noted.

Because of the rate of refractory failure, the slags produced were high in alumina, and extremely viscous. Our second run, with its highly oxidizing conditions due to low bed and low charge, produced a slag apparently high in iron that eroded the slag hole block badly although the well temperatures were not high.

The prices quoted for high rates of oxygen consumption (Table 5) are most attractive, but the volumes required to attain these prices exceed any predictable use in existing foundries and would also require operation "around the clock" since the operation of the oxygen generator must be continuous.

We might combine some of the figures in Table 4 and Table 5 of the paper to find the size of operation that would be required to use oxygen in some of the cheaper classes.

If we consider only the size 12 cupola as being suited to large

<sup>1</sup> Allis-Chalmers Mfg. Co., Milwaukee



operations and further, assign probable values of 180,000,000 cu ft per month for the "Very Large Quantity Purchase," and 360,000,000 cu ft per month for the "Huge Quantity Purchase," we may in turn compute the number of cupolas that must be in operation at all times during the 24 hr to use the amounts noted. On smaller operations we can assume liquid storage of the oxygen, and operation of cupolas on an 8-hr basis.

We find for 25 cents per ton of iron oxygen (at 4 per cent addition) we would require 16 cupolas in operation 24 hr a day 30 days a month. With 10 per cent added oxygen (at \$.55 per ton) this drops to 6.5 cupolas in constant operation.

When we consider smaller operations we find that with 4 per cent additions we can use \$1.75 per ton of iron oxygen cost with 770 cupola-hours a month or an average of slightly over one cupola in operation continuously. Thus we can assume that our costs for oxygen alone will exceed \$1.75 per ton of iron and will more likely be around \$2.50 for a normal size foundry.

It may be gathered from the above that we do not wholeheartedly subscribe to the general use of oxygen. We do, however, agree with Mr. Wick that a bank of manifolded cylinders, attached to the blast line would be desirable since they would serve to bolster falling temperature and in addition would be used to raise the temperature of the first tap and to make iron available more rapidly at the start of the day.

#### MR. CROCKETT'S DISCUSSION

J. M. CROCKETT (*Written Discussion*):<sup>2</sup> This paper presents data that would be difficult to obtain from production tests, and the author is to be congratulated for a very interesting summary of the experimental work.

About two years ago, we assisted in some oxygen-enriched cupola blast tests, made under operating conditions in a plant manufacturing wrought iron. At that time, it was obvious that there were some advantages in the use of oxygen, but it was also apparent that oxygen-enriched blast introduced new problems. Subsequent work involving production cupolas confirm the conclusion of this paper that oxygen will increase the metal temperature and production rate, but differ from the conclusions regarding refractory consumption and oxidation losses.

It should be emphasized that oxygen enriched blast accomplishes two purposes in a cupola. It causes a higher flame temperature, or higher reaction zone temperature, because of the reduced nitrogen content of the blast. In addition, more oxygen per unit time can be readily supplied to the cupola, and consequently the rate of fuel consumption, and production, can be increased. Both of these objectives may be obtained by other means the first by drying or heating the blast and the second by increasing the blast rate. Therefore, the application of oxygen to the cupola should be considered as merely another one of several means of improving cupola operation, and not as a cure all for present operating difficulties.

The question of oxidation loss summarizes the major problems that are encountered in applying oxygen-enriched cupola blast. This paper definitely shows that metal loss is less when using oxygen, than in normal operation. This would be expected if the coke ratio remained constant. It has been found in open hearth operation that the loss due to oxidation is a function of time as well as atmosphere during melting, and if the time of melting is shortened, and the atmosphere remains essentially of the same oxidizing characteristics, then the loss of carbon and silicon will be less with oxygen. The combustion problem in the cupola is more complicated than in the open hearth. Coke is burned to provide heat to melt iron, but it must be burned in such a manner that the chemical changes of the charge can be controlled within definite operating limits. Consequently, in order to prevent oxidation loss, there will be a definite minimum coke consumption at various oxygen enrichments. This is very important, since it has been found that coke savings of 10 to 15 per cent are the maximum attainable without oxidation loss. With such a limitation on coke savings, it is apparent that the use of oxygen is not economical under present conditions.

There is still concern that the refractory problem in using oxygen will be difficult to overcome, particularly at high enrichments. It has been our experience, that, with enrichments above 3 per cent added oxygen, refractory consumption is definitely greater.

The use of oxygen during an extended period of cupola operation is not economical at the present time. Cupola melting is an efficient means of utilizing fuel, as compared with other melting units. Because of its efficiency, the magnitude of the inherent improvement from the use of oxygen is limited. We have found in using oxygen for open hearth combustion, that the greatest benefits could be obtained in shops having old furnaces operating over their rated checker capacity, while new, efficient furnaces showed less benefit in using oxygen. It seems evident from this paper and other work that the saving in coke using oxygen will balance the cost of oxygen only in extremely unusual cases, if at all.

The intermittent use of oxygen, as recommended, offers the best prospect. However, it should be used carefully and under competent supervision, otherwise the cost of oxygen will quickly overshadow its benefit. The supply of oxygen for this purpose should be planned so that the desired flow rate can be maintained during the time the oxygen is needed. This means the use of adequate-sized pipe lines from the oxygen supply and proper regulating and control equipment.

Oxygen has been established as an important—and in some metallurgical operations an essential—reagent, but it is not a substitute for more economical means of obtaining satisfactory operations and fuel consumption.

#### MR. CLARK'S DISCUSSION

R. A. CLARK (*Written Discussion*):<sup>2</sup> The writer has reviewed the paper on "Oxygen-Enriched Cupola Blasts," by W. C. Wick, with great deal of interest. As the author has indicated, reports from the steel industry concerning the application of oxygen to open hearth melting has aroused the curiosity of many foundrymen concerning the possibility of its use to enrich the cupola blast for the purpose of obtaining more efficient combustion with the view of increasing metal temperature and reducing the amount of coke necessary to melt their iron.

Mr. Wick and the sponsors of this program have published the most complete report on this subject which has come to our attention to date. Their data can be used as the basis of further investigations which will show the possible value of enriched blast in larger commercial units.

We believe it should be stressed that the results obtained with this small experimental cupola of 19 in. ID, may not be checked exactly using the larger furnaces which are usually found in commercial practice. The high proportion of wall area to the volume of the combustion zone has a tendency to hold the thermal efficiency of very small cupolas to a somewhat lower level than that of the larger furnaces and it is very likely that the optimum oxygen added to the blast might be found to be considerably lower in a larger cupola.

We also believe that we should not overlook the fact that the cupola used in these experiments was of a standard design which has been developed over a period of generations as suitable for operation on a normal cold air blast. Presumably the use of oxygen has the effect of increasing the rate of combustion at the tuyere level. This condition would be expected to give an unusually high temperature in the lower part of the coke bed and a correspondingly lower temperature in the upper portion of the bed. We would expect such a condition to cause burning away of the cupola bed to a level somewhat lower than when using a straight air blast. There are definite indications that such a condition existed in some of Mr. Wick's heats, particularly when the heaviest oxygen additions were used. However, to melt iron at the maximum temperature for a given coke ratio, we must not only maintain an extremely hot and active bed, but must maintain it at sufficient depth so the metal dripping through it will be exposed to the high temperature a sufficiently long time to accumulate the desired superheat.

The use of substantial amounts of oxygen would be expected to upset the normal combustion equilibrium to the point where major changes in the design of the cupola furnace might be indicated to obtain the best results. As an example, it is possible that introduction of the blast at several levels rather than through one row of tuyeres might help maintain a sufficiently deep bed to considerably improve the absorption of available heat by the melting metal.

Mr. Wick has shown price figures indicating that the cost of an oxygen addition of 4 per cent to the cupola blast might vary

<sup>2</sup> Air Reduction Sales Co., New York

<sup>2</sup> Electro Metallurgical Co., Chicago

from \$9.40 per ton of melt for a very small foundry, to as little as \$0.25 for an extremely large purchaser of oxygen. We believe it may be of interest to place several foundries of different capacity in this table. For this purpose let us consider three plants melting 20, 100 and 1000 tons of metal per day. Cost of a 4 per cent oxygen addition to the cupola blast at each of these foundries would then run in the range of:

Daily Melt, tons	Monthly Usage of Oxygen, in cu ft	Price of Oxygen per cu ft	Cost per ton of Metal
20	403,200	\$0.0050	\$5.04
100	2,016,000	\$0.0023	\$2.32
1000	20,160,000	\$0.00145	\$1.46

With the coke for an overall metal to coke ratio of 5 to 1 casting some \$4 to \$6 per ton of melt, depending on locality, it is obvious that the average small foundry is going to have some difficulty justifying the use of oxygen. However, it is possible that the very large foundries might be able to justify a cost in the neighborhood of \$1.50 per ton of melt for oxygen if they could make substantial savings in coke or attain higher and more consistent temperature.

In papers of this kind which present a large mass of experimental data, most authors feel a hesitancy about reporting more of the fine details than are absolutely necessary. In all probability, Mr. Wick has much pertinent information on details of the experiments such as initial coke bed heights on each heat, chemical analysis of the metal produced at each tap, the weight of metal tapped and the time necessary to accumulate it for each tap, etc. It is hoped that the author may see fit to publish this information at an early date, as such data would be of definite value to others who may wish to carry on further experimental work along this line.

#### Cost Calculations on Use of Oxygen Based on Wick's Prices.

##### 4% Oxygen Addition to Blast

30,000 cu ft air to melt 1 ton iron

$30,000 \times .21 = 6,300$  cu ft  $O_2$  to melt 1 ton iron.

If we enrich the blast to 25%  $O_2$ , we will require

$6300 \times \frac{25}{100} = 1,575$  cu ft total blast volume.

$25,200 \times 0.04 = 1,008$  cu ft of  $O_2$  would be added per ton of iron melted.

Melting 20 tons/day or 400 tons/month:  $400 \times 1008 = 403,200$  cu ft  $O_2$  consumed at a price of \$0.0050 per cu ft

$1008 \times 0.0050 = \$5.04$  per ton of melt.

Melting 100 tons/day or 2,000 tons/month:  $2000 \times 1008 = 2,016,000$  cu ft  $O_2$  consumed at a price of 0.0023 per cu ft

$1008 \times 0.0023 = \$2.32$  per ton of melt.

Melting 1,000 tons/day or 20,000 tons/month:

$20,000 \times 1008 = 20,160,000$  cu ft  $O_2$  at a price of 0.00145.

$1008 \times 0.00145 = \$1.46$  per ton of melt.

#### MR. BOWERS' DISCUSSION

J. A. BOWERS (Written Discussion):<sup>4</sup> When we were asked to discuss this paper we were a little hesitant to do so because of the meager amount of data we had available. However, after checking our data we found that it checked in many instances with Mr. Wick's work and decided to offer it for what it is worth. The data was collected in somewhat the same manner as Mr. Wick's. The first test being on a pair of experimental heats and the others from a production cupola with the blast enriched with oxygen for a short period of time.

Two experimental heats were run on a cupola lined to 36-in. diameter. The first heat was conducted with normal blast while the second had approximately 4.5 per cent oxygen added. The cupola blast was supplied from a positive pressure blower and the rpm of the blower was the same for both heats with all bleeder valves closed tightly. The iron charges weighed 1000 lb and contained 20 per cent clean steel and 80 per cent pig iron from the same cast. The iron to coke ratio was 8 to 1, and the flux used was dolomite, the amount being 3 per cent of the weight of the iron charge. The cupola held six charges level with the charging door sill and on top of the sixth charge of both heats approximately 12 in. of coke was placed to tend to make the stack resistance even throughout the blow. The metal

from both heats was collected in 3000-lb ladles and inoculated with 3 lb of 75 per cent FeSi per 1000 lb of iron. Test bars, chill specimens and samples for chemical analysis were poured from each ladle. The oxygen was not introduced into the blast for the first 10 min on the oxygen-enriched blast heat as it was felt that most of this time is spent in heating the stack and therefore it would be a waste of oxygen. A comparison of the data is given in Tables 6 and 7.

TABLE 6—SUMMARY OF DATA FROM PLAIN VS OXYGEN-ENRICHED BLAST IN THE CUPOLA

	Plain Blast	Oxygen Enriched Blast
Time from wind on to first tap, min	35	27
Duration of heat (6000 lb iron), min	57	47
Time oxygen on, min	0	37
Per cent oxygen in blast	21.0 (atmos)	25.5
Average metal temperature, °F	2665	2740
Refractory consumption, cu in.	1758	3215
Average silicon loss, per cent	12.9	9.9
Points carbon pick-up	0.24	0.29
Points sulfur pick-up	0.059	0.045

TABLE 7—CHEMICAL ANALYSIS

	Plain Blast	Oxygen Enriched Blast
Average silicon content, per cent	2.02	2.09
Average sulfur content, per cent	0.091	0.080
Average manganese content, per cent	0.80	0.79
Average phosphorus content, per cent	0.35	0.36
Average total carbon content, per cent	3.45	3.50
Total iron oxide in the slag, per cent	4.46	2.64

The next tests consisted of adding approximately 3.5 per cent oxygen to the blast in a production cupola lined to 48-in. diameter. The iron charge in this operation consisted of 20 per cent steep scrap, 10 per cent return scrap and 70 per cent pig iron, the total weight being 2000 lb. The iron to coke ratio was approximately 7 to 1 with the normal amount of flux. All the iron in this foundry is inoculated with 75 per cent FeSi and is poured mostly into cast iron water and gas pressure fittings. No changes were made to normal procedure during the tests except for the addition of oxygen to the blast.

The tests consisted of operating the cupola in the normal way for approximately 5 hr. Frequent temperature readings and chemical samples were taken during a period of approximately 30 min before adding oxygen to the blast. After which 3.5 per cent oxygen was added for 30 min during which time temperature readings and chemical samples taken as stated above. The oxygen was then cut off and the same observations made for 30 more min. The results are shown in Table 8.

TABLE 8—AVERAGE RESULTS OBTAINED DURING APPROXIMATELY 30-MIN INTERVALS BEFORE, DURING AND AFTER OXYGEN WAS ADDED TO THE BLAST OF A PRODUCTION CUPOLA

Time	Aver. Stream Temp., °F	Si	S	Mn	P	T.C.	% $O_2$ Added to Blast
FIRST TEST							
Before enrichment	2650	1.91	.092	.74	.34	3.46	none
During enrichment	2640	2.10	.096	.77	.35	3.46	3.5
After enrichment	2650	2.02	.084	.67	.37	3.40	none
SECOND TEST							
Before enrichment	2620	1.95	.092	.94	.43	3.34	none
During enrichment	2670	2.07	.096	.52	.49	3.39	3.5
After enrichment	2660	2.00	.091	.54	.43	3.36	none

The following conclusions are listed from the short experimental heats. All of them to some degree in agreement with Mr. Wick's conclusions:

1. The speed of melting was considerably increased being approximately 18 per cent with 4.5 per cent oxygen added.

<sup>4</sup> American Cast Iron Pipe Co., Birmingham

2. The temperature of the iron was higher with oxygen-enriched blast by 75 F.

3. There was 3 per cent less silicon loss with oxygen-enriched blast.

4. If the total iron oxides in the slag are indicative of total conversion loss, then there was less conversion loss with oxygen-enriched blast.

5. The one conclusion in Mr. Wick's paper that our data failed to check was with reference to refractory consumption. We consumed nearly twice as much refractory when using oxygen-enriched blast.

The results from the production cupola tests seemed to indicate, as pointed out by Mr. Wick, that additional oxygen in the blast might be used as a "medicine" to increase melting speed or overcome bridging, cold iron, etc. Although our tests in this instance were not too conclusive we might not have waited long enough after "oxygen on" or "oxygen off" to get the full effect, in our recorded data. It might be well to point out that after the oxygen had been on a very short time the tuyeres became so bright it was difficult to look at them with the naked eye.

We would like to point out that the above conclusions should be considered as indications because of the meager data as pointed out in the beginning. Further experimentation and considerable more study is necessary for positive conclusions on the effects of oxygen-enriched blast on cupola operation.

W. R. JAESCHKE:<sup>5</sup> If oxygen should prove to be a very advantageous or desirable medicine to use in the cupola, and the refractory were the limiting condition, there is the possibility that we might go back about 50 years in cupola design and introduce the oxygen into the center of the cupola through the old type of center tuyere. I thought I might offer that as a possible suggestion for future work.

J. T. MacKENZIE:<sup>6</sup> The two points that I should like to mention are that in our results, for instance, the increased consumption of refractories made more slag which might account for the lower iron content of the slag. Another comment I should like to make on the Armour test is on the very poor results they got with the 5 to 1 coke ratio. They used the wrong procedure by varying the coke. As I figured it out, roughly each charge must have been almost 10 in. of coke when they got up to 5 to 1 ratio. That is much too thick a charge of coke to use in such a small cupola. I am sure if Mr. Wick had varied his iron instead of varying his coke he would have gotten better results with that 5 to 1 coke ratio.

D. E. KRAUSE:<sup>7</sup> I should like to ask Mr. Bowers as to the nature of the patching material used in his cupola.

MR. BOWERS: We used the conventional manufactured 5-in. straights and splits for patching and the daubing that we use is what nearly everybody uses, a mixture of fireclay with ground firebrick as a grog.

MR. KRAUSE: I believe that some of the decrease in refractory burnout reported by Mr. Wick is due to the superior quality of the refractory mix shown on page 249. We have experimented with the same type of refractory mix and obtained unusually good results. I would suspect that much of the difference in refractory burnout between Mr. Bowers' operation and Mr. Wick's upon the introduction of oxygen to the blast might be due to the difference in the refractory qualities of the lining material.

H. P. KREULEN:<sup>8</sup> Was there any special reason why in the experiment the metal temperature at the cupola spout was 2700 F and lower? Would it not have been better to melt the iron at the usual temperature of 2760 to 2790 F, then add the oxygen starting from that point?

CHAIRMAN McELWEE: This was an experiment with a small cupola to determine the effect of oxygen. It was not a production shop run. I think they were justified on that basis.

MR. KREULEN: That is true, but there are foundrymen who operate 36-in. lined cupolas who melt the first taps at a temperature of 2760 to 2790 F.

H. F. POPER:<sup>9</sup> I should like to ask Mr. Bowers where they admitted the oxygen in their production cupola?

MR. BOWERS: We put it in the blast line about 10 ft from the blower, between the blower and the cupola.

MR. POPER: Before it comes into the Y?

MR. BOWERS: Outs enters at one point.

A. E. SCHUH:<sup>10</sup> The paper given by Mr. Wick and the remarks by Mr. Bowers indicate that there is need for considerably more work along this line. As to the economic possibilities of the use of an oxygen-enriched blast, even if everything works out all right we are in a rather strange position. It can be seen from the chart that small amounts of oxygen are going to be rather expensive as an expendable operating material. It would be helpful to know what the capital investment would be for a user of a large quantity of oxygen. If this work is going to be pursued further, and it certainly should be, much more information will have to be acquired before any large foundry can afford to take the risk of a large capital investment to take advantage of the possibilities that are indicated.

<sup>5</sup>Whiting Corp., Harvey, Ill.

<sup>6</sup>American Cast Iron Pipe Co., Birmingham

<sup>7</sup>Gray Iron Research Institute, Columbus, Ohio

<sup>8</sup>Grede Foundries, Inc., Milwaukee

<sup>9</sup>Frick Co., Inc., Waynesboro, Pa.

<sup>10</sup>U. S. Pipe & Foundry Co., Burlington, N. J.



# CHANGES IN CHEMISTRY OF LIQUID STEEL IN CONTACT WITH SAND

Mold Surface Committee of A.F.A.

J. B. Caine,\* Chairman

## ABSTRACT

*This newly formed committee presents a study and report on the happenings at the sand-metal interface when a mold surface is subjected to liquid metal. One of the many possible happenings is a change in chemical composition of the sand and metal. The report this year concerns itself with changes in chemical composition of the metal—the instances reported are for steel—when it contacts the sand to form the casting. A corollary of this work has to do with any possible oxidation or volatilization of steel as an explanation of penetration and adhering sand.*

CHANGES OCCURRING in the chemistry of the surface of metal at the sand-metal interface are not generally appreciated. Although the examples given in this report are for steel, changes of a similar nature, except for differences in chemistry must occur in any metal casting. Figure 1 illustrates a typical steel section in contact with a typical production sand. There was no metal penetration into the sand in this instance, the sand peeled perfectly from the casting. Note the decarburized surface and the presence of large inclusions at the sand-metal interface. The decarburized area in this instance is 0.012 in. deep, the oxidized area showing many large inclusions extends about half this distance from the surface.

On higher magnification it will be found that the inclusions are manganese silicate and iron sulphide even though the steel be aluminum killed, showing only alumina and manganese sulphide inclusions a few thousandths of an inch removed from this oxidized surface. Normal inclusions are found even in the area showing appreciable decarburization.

Fortunately, from the standpoint of physical properties of the casting, this oxidized, decarburized area usually scales off during heat treatment and the problem of poor physical properties, especially in fatigue is not too serious. However, it is something to be kept in mind if any thought be given to scale-free, controlled atmosphere heat treatment.

If, for any reason, the void size of the sand as rammed is too large, resulting in mechanical penetration, the penetrating metal will be found in most

instances to be completely decarburized and showing many large silicate inclusions, as shown in Fig. 2. This change in chemistry at the surface of cast steel, regardless of whether the steel penetrates the sand or not, does not seem to be due to any reaction with the sand itself, but due to a reaction between the steel and oxygen and hydrogen of the air occupying the voids of the sand and present in the gases generated by the sand as it is heated. Water vapor, present either as temper water, combined water in the clay substances, or formed when the cereal and organic materials in the sand decompose, is appreciably dissociated into hydrogen and oxygen at the temperatures of liquid steel.

Hydrogen, in the presence of water vapor is about the most efficient decarburizing agent known. Oxygen will combine with silicon, manganese and aluminum forming inclusions. In fact, enough oxygen must be available to react with all the oxidizable elements present at the surface for iron sulphide inclusions are present in the oxidized surface areas. This is an inclusion type that does not form in steel until the manganese content of the steel is almost exhausted.

## Temper Water Content

There is some evidence that the amount of oxidation and especially decarburization, as measured by the depth of metal showing these changes in oxygen and carbon, is dependent primarily on the temper water content of the sand. The amount of combined water and organic content of the sand seem to be secondary variables. Steel cast against dried sand usually shows less oxidation and decarburization than does steel cast against sand containing water, and even dry sand castings show appreciable oxidation and decarburization of their surfaces in contact with the mold surface.

One thought may be timely. The steel foundryman has always recoiled instinctively from the thought of carbon in any form in his sand, due to the possibility of carbon pick up by the steel. Figures 1 and 2 are representative of hundreds of samples. If carbon in any form is ever found to have possibilities in improv-

\* Metallurgist, Sawbrook Steel Castings Co., Lockland, Ohio.

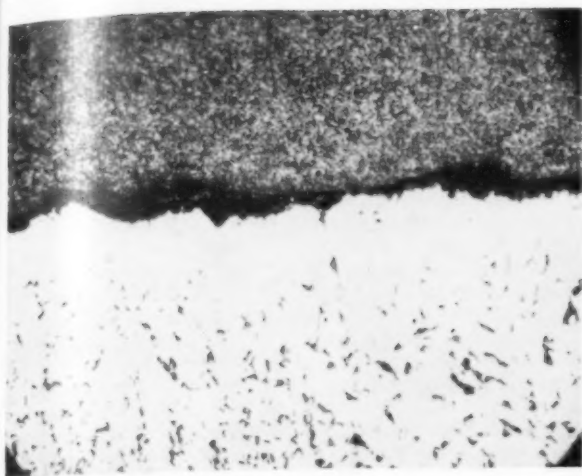


Fig. 1—Steel surface in contact with sand. No penetration. 25X, vertical illumination.

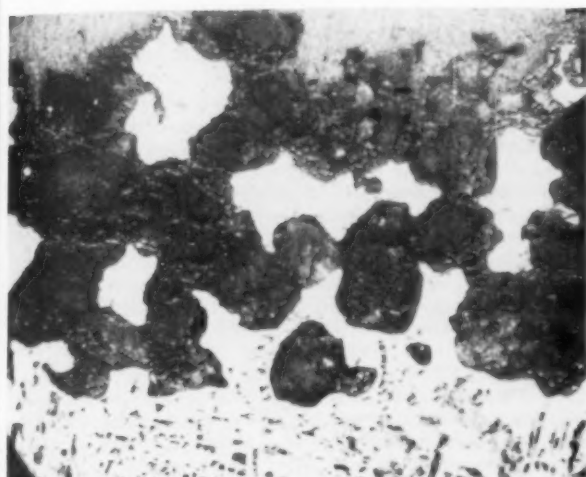


Fig. 2—Steel surface in contact with sand. Bad penetration. 25X, vertical illumination.

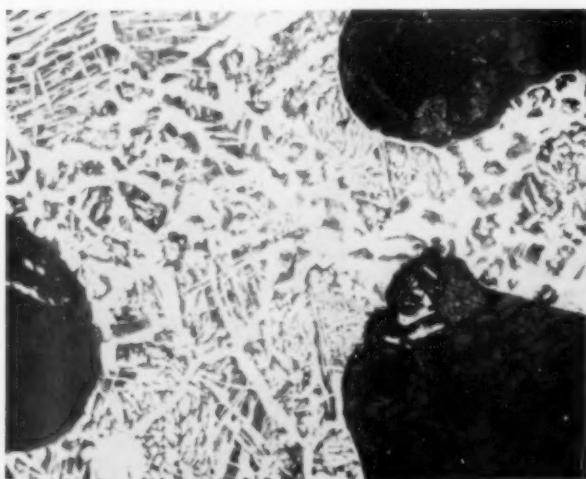


Fig. 3—Example of extreme fusion of sand and penetration. 25X, vertical illumination.

ing the behavior of steel foundry sands when in contact with liquid steel, the mythical danger of carbon pick up should not discourage the thorough investigation of this type of addition. Some carbon pick up from the sand will be advantageous.

That changes in carbon and oxygen of the surface of steel in contact with sand are not due to fusion or sintering of the sand is illustrated by Fig. 3. This particular sand melted completely when in contact with liquid steel, the fusion opening large voids and causing extreme penetration. Note that in this case there is no oxidation or decarburization of the penetrated metal, even though this is an example of the extreme. This behavior has been checked a number of times with a number of types of sands. It would seem that fusion or sintering seals off the metal from any decarburizing or oxidizing gases, effectively preventing any reaction and change in chemistry at the sand-metal interface.

So much for carbon, hydrogen and oxygen. Chemical analysis of penetrated metal should prove interesting if for no other reason than to settle once and for all if there is any possibility that oxidation or volatilization of the steel plays a part in penetration and adhering sand. As a chemical solution must be made as the first step in such studies, carbon cannot be determined, but as has been shown previously in this report, carbon is influenced primarily by the gases of the mold atmosphere. However, an investigation of the relative contents of manganese, nickel, chromium and molybdenum in the base and penetrated metal should throw light on this question.

If metal penetration is due only in part to oxidation and subsequent reduction, oxidizable elements such as manganese and chromium should be oxidized preferentially and be present in the penetrated metal in substantially higher concentrations than in the parent metal. Conversely, stable alloys such as nickel and molybdenum would not be oxidized and would not be present in the penetrated metal, although appreciable amounts be present in the parent metal. If volatilization plays a part in the mechanism of penetration of steel, alloys with a high vapor pressure such as molybdenum will be present in the penetrated metal in larger amounts than in the parent metal.

TABLE I—ANALYSIS OF PENETRATED METAL

Type of Steel	Type of Sand	% Sand in Penetrated Mass	% Steel in Penetrated Mass	Analysis of Penetrated Metal			
				Mn	Ni	Cr	Mo
2315	13% Cement	32	68	..	3.94	..	..
2315	"	37	63	.22	3.42	..	..
2315	"	32	69	..	3.92	..	..
1015	Green	32	68	.32	..	..	..
1030	"	33	67	.84	..	..	..
1030	Core	33	67	.65	.28	..	..
4140	Green	35	65	.50	..	.75	.20
4140	"	38	62	.38	..	.73	.17
0.40-0.60 Mo.	Dry	35	65	.41	..	..	.45

The analyses of nine samples of penetrated metal are shown in Table I. Although the samples are only from two steel foundries, there is no reason to believe that they are not representative. As can be seen from Table I there is no appreciable difference in the alloy contents of the penetrated metal in comparison with that of the parent metal, with the exception of a lower manganese content of the penetrated metal in some cases. This decrease can be explained by oxidation at the sand-metal interface by the mold gases as detailed in the first part of this report.

No silicon contents have been reported due to analytical difficulties. As considerable time is required to dissolve the steel from the steel-sand mixture the silicic acid has an opportunity to oxidize and is filtered off as silica with the sand in the separation of the sand from the steel.

### Summary

From the evidence presented in this report, the following conclusions can be made.

1. Oxidation and decarburization of the surface of steel in contact with sand is due to a reaction with the mold gases. Dissociated water vapor is the most active gas.
2. Scaling during heat treatment will automatically remove the decarburized surface, unless excessive.
3. Evidence from chemical analyses of penetrated metal discounts oxidation or volatilization as mechanisms of penetration of steel.

### Mold Surface Committee Personnel

J. B. Caine, *Chairman*

J. A. Rassenfoss	D. C. Williams
H. W. Dietert	L. B. Osborn
Charles Locke	E. C. Troy
H. M. Kraner	K. J. Jacobson

### DISCUSSION

*Chairman:* C. W. BRIGGS, Steel Founders' Society, Cleveland  
*Co-Chairman:* WERNER FINSTER, American Chain & Cable Co., Reading, Pa.

H. H. BLOSJO (*Written Discussion*):<sup>1</sup> The Committee is to be congratulated on the fine photomicrographs of the sand metal interface of a casting and mold. That the penetration of

<sup>1</sup> Minneapolis Electric Steel Castings Co., Minneapolis

metal into the sand voids is due to mechanical conditions seems a certainty. The conclusion, however, drawn as to the compactness of the sand in areas of penetration seems to be in error.

In Table I the average weight of sand found in nine samples was 34.1 percent and the average weight of steel was 65.9 percent. The specific gravity of quartz being 2.65 and of steel 7.6, the ratio of the volume of sand ( $V_s$ ) to the volume of metal ( $V_m$ ) would be as follows:

$$\frac{V_s}{V_m} = \frac{34.1}{65.9} \times \frac{7.6}{2.65} = \frac{12.88}{8.67} = 1.48$$

$$\text{Per cent sand by volume} = \frac{1.48}{1.48 + 1.00} \times 100 = 59.6 \text{ per cent}$$

$$\text{Per cent sand by volume} =$$

Assuming that the voids in the mold were all filled by steel, which seems quite logical, the mold would contain 59.6 percent sand and 40.4 percent voids.

A standard rammed sample under these conditions would weigh 59.6 percent of 272 grams or 162 grams. This represents a fair ram and represents without doubt a truer condition of the mold than that indicated by the Committee.

J. B. CAINE (*Author's Written Reply to Mr. Blosjo*): Mr. Blosjo brings up an important point. Although soft ramming and the resultant large voids between the sand grains must be one cause of adhering sand, the sand areas under study and showing extreme penetration have been rammed to about an average level. Then why did the metal penetrate into the sand? All of the specimens except the last were taken from in between the teeth of cast tooth gears, from cores and reentrant angles, areas surrounded on more than one side by metal, areas that are heated to a relatively high temperature. The last specimen was taken from the bottom of a high vertical surface, an area subjected to relatively high hydrostatic pressure. Every foundryman knows that these areas are most prone to adhering sand.

Figures 4 and 5 are photomicrographs of the sand grains extracted from two of the sands of Table I after dissolving the steel from the mass with dilute acid. Again, there seems to be no sign of fusion or chemical change, even in the relatively fusible cement sand. The grains are angular, if anything more angular than the original sand, due to fragmentation. Cracking and fragmentation seems to have been quite extensive, due to either heat shock or stresses set up due to shrinkage of the steel. The sand grains adhere together slightly after removal of the steel, but the mass can be broken up with your fingers.

The important question is now, how small must the voids be to prevent penetration in those areas heated to above the melting point of the metal surrounding them. How hard must these areas be rammed and what is the effect of hydrostatic pressure. Information along these lines should point the way to minimizing adhering sand and the practical problems associated with it.

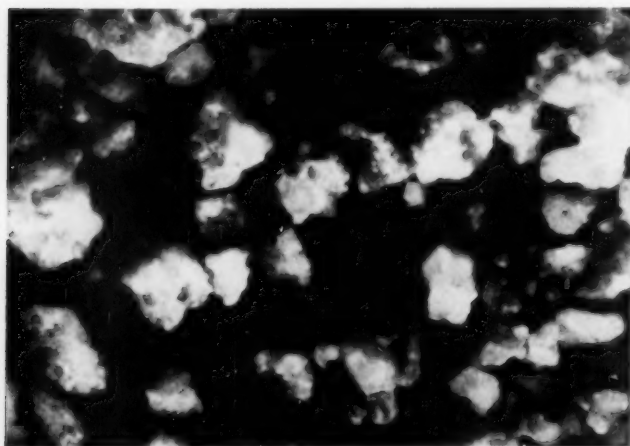


Fig. 4. Sand Grains Extracted from Penetrated Mass of Cement Sand. 25X. Oblique Illumination.

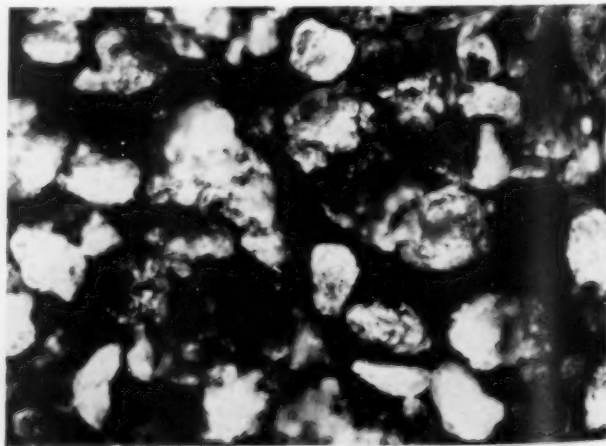


Fig. 5. Sand Grains Extracted from Penetrated Mass of Molding Sand. 25X. Oblique Illumination.



J. B. CAINE

H. H. FAIRFIELD:<sup>2</sup> My observation is that penetration decreases in severity with temperature. If you can get your temperature low enough the phenomenon almost disappears.

MR. CAINE: I grant you that metal temperature has a decided influence on penetration. If you will pour the metal very hot you will increase penetration. I have seen some very smooth castings poured with very hot metal, and I have seen cases of extreme penetration with cold metal. I just do not know. I think that we have variables that are not under control. They are at least variables that we do not appreciate, let us put it that way; and I think one of them is hydrostatic head. Nobody has ever studied that, and I think it is very important.

A. J. KIESLER:<sup>3</sup> With reference to a hydrostatic head and temperature, we performed an experiment using a standard 2-in. core around which we cast Wood's metal at approximately 170 F under 10 lb. of pressure. Upon examining the casting very closely a faint outline of the core could be seen. The metal penetrated completely through the core. The core used for this experiment was of the identical mixture and physical properties as the core used in casting the steam chest of our turbine castings.

MR. CAINE: Although you are penetrating the core at 170 F perhaps the amount of superheat was as great as if steel be poured at 2850 F. Do you know the degree of superheat in your experiment?

MR. KIESLER: The melting temperature of Wood's metal is approximately 160 F.

MR. CAINE: Do you know what the surface tension of Wood's metal is? Is it very low in regard to a higher melting point material such as iron or steel?

MR. KIESLER: I do not know.

MR. CAINE: Surface tension probably has an effect, but there is no information available.

J. A. RASSEFOSS:<sup>4</sup> What was the permeability of the core tested?

MR. KIESLER: The permeability of the core tested was 160.

H. M. KRANER:<sup>5</sup> I very much favor this matter of temperature which has been discussed, and the resulting viscosity and surface tension of metal resulting from variations in temperature. I would like to call attention to the fact that when we hold post mortems on hearths of blast furnaces, we find metal penetrated very thoroughly and completely into the brick so that penetration with static pressure is quite readily understood. I think we do not give enough consideration to the effect of the oxides which develop in castings because you cannot get metal to penetrate a piece of carbon or coke, for example. You cannot get it to penetrate a piece of magnesite either. The reason for this is that it will not wet those materials. If you get a slag on a silica mold surface you may get some wetting due to that slag. In other words, manganese oxide or iron oxide may react with surface. The mineral fayalite is  $2\text{FeO} \cdot \text{SiO}_2$ . This melts at a low temperature. Sand might go into solution with the manganese analogue of fayalite. We found by studying these mold surfaces with a binocular microscope that some of these crystals were not quartz. With a needle we could transfer them to a petrographic microscope and determine definitely that they were fayalite.

The whole thing indicates that there was some liquid developed on the mold surface prior to the penetration of the metal and that liquid slags may have something to do with penetration. It also indicates to me that there was no oxidation of the materials there, because you cannot get fayalite to form under oxidizing conditions. Fayalite is only formed under reducing conditions.

MR. CAINE: Were these sand surfaces obtained from the casting after it cooled to room temperature?

MR. KRANER: Yes.

MR. CAINE: The question then is when did fayalite form? If it formed after the steel had solidified that is one thing. If it had formed before the steel had solidified it is something else. I believe that oxides formed after the steel is solidified. I have taken a specimen of a sand similar to the ones that we are studying, immersed it into liquid steel, brought it out, and found a trace of a liquid constituent at the surface in contact with the steel. But the amount of iron present in the surface that had been in contact with only liquid steel was very low. The chemical analysis ran less than one per cent FeO. The reason I can say nothing more certain is that mixed up in between the grains were very minute droplets of iron. So I do not know how much of that FeO was present as Fe and how much was present as an iron silicate. No crystals could be seen. There was simply an orange-colored liquid phase in between the sand grains. Could it be that fayalite was formed after the metal had solidified? When the sand is in contact with steel until the casting has cooled to room temperature, we find as much as 15-20 per cent FeO. Fayalite which formed after the steel had solidified could have no bearing on penetration.

MR. KRANER: I would say the oxide is formed during the pouring of the metal. Anyone who has obtained samples of the "gunk" as we call it, which comes to the top of an ingot knows, that it forms during the pouring of that ingot. I think the same thing happens in the pouring of a steel casting.

Fayalite must form on cooling, because it melts at around 2200 F and could not exist as a crystal above that temperature. Fayalite must be maintained under reducing conditions until cooled otherwise you would get cristobalite and magnetite.

G. F. WATSON:<sup>6</sup> We have done considerable work on this problem. With the formation of the oxide, working with manganese steel and using the immersion method which I think Mr. Caine developed, we get manganese silicates formed in about 2 sec. and they are not formed after the metal solidified. We have isolated them.

We are also working on penetrated metal itself. Under the microscope we find that metal will go through a channel as fine as about 0.0004 in., and that is back some distance from the sand metal interface where you would expect the metal to be quite cool—cold as a matter of fact.

We do not know why it goes through there. I am inclined to believe that the static pressure is not the reason, because in heat resisting alloys (pouring drive shafts) we pour them on end with about a 16-ft head and we have no more penetration than we have in ordinary castings with heads of 15 in. or 10 in. I am inclined to believe that it is either pressure back of the metal, as from gas, or a suction caused by oxidation of the core oil. We do not know, we are hoping to find that out.

<sup>2</sup> Wm. Kennedy & Sons, Owen Sound, Ontario, Canada

<sup>3</sup> General Electric, Schenectady, N. Y.

<sup>4</sup> American Steel Foundries, East Chicago, Ind.

<sup>5</sup> Bethlehem Steel Co., Bethlehem, Pa.

<sup>6</sup> American Brake Shoe Co., Mahwah, N. J.

# RELATION OF CUPOLA RESEARCH PROGRESS IN CAST IRON DEVELOPMENT

Report of Chairman, A.F.S. Cupola Research Committee

R. G. McElwee \*

THERE APPEARS TO BE little hope of alleviation of the current critical situation on foundry raw material, especially coke and pig iron. This has influenced the A.F.S. Cupola Research Committee to concentrate their research efforts toward establishment of such practices as will give the foundryman maximum results with presently available materials. Such a program can and does contemplate studies of cupola melting conditions which can be of considerable assistance to the operator when and if better materials are available again.

While a study of fuels has been announced as a major study of the group, it is obvious that any contributing factor to better and more efficient operation must be included in the investigation.

Previous reports have indicated that the United States Bureau of Mines as well as the Canadian Bureau of Mines have offered to and are contributing to the investigation of cupola fuel. Ford Motor Co. has given coke of different coking cycles for study and production and laboratory reports have been made which will be published in AMERICAN FOUNDRYMAN.

## Coke Quality Test

It is the hope and aim of the committee to prepare a recommended laboratory practice for the evaluation of coke which will permit the operator at reasonable cost and in minimum time to assign values to each shipment of coke which will enable him to differentiate between lots on a basis of the amount necessary of each lot to attain reasonably constant melting practice. This is being done today by a rather costly and cumbersome method.

In the contacts which our staff have made to collect data it has become increasingly apparent that certain factors of control have sufficient bearing on results to justify inclusion in our study.

One of the outstanding factors which has been so

established is the control of slag analysis. Records have indicated that a fluid slag of correct lime-silica ratio can contribute much to the success of carbon recovery and iron structure. Records have been kept of good, indifferent and bad cupola operations in the field and there has been found a remarkable trend toward eutectics of lower melting point in slags as quality of operation improves.

## Slag Fluidity

Whether this is cause or effect is important but not vital. What has appeared significant is the improvement of certain operations when slag analysis was altered to give greater fluidity and approach a predetermined lime-silica ratio. Reports have been released on this phase of the investigation and the work continues. See "Slag Control Is Important In Cupola Operation," AMERICAN FOUNDRYMAN, p. 109, May, 1948.

Correct placing of the charge, careful attention to blast pressure-tuyere opening ratios and other factors within the control of the operator are being studied and reported for his benefit.

The outstanding fact which spurs the study and intrigues the participants is the fact that we still have with us some remarkably well operated cupolas, and certainly, we should be able to find and transmit to others the secret of their success.

The CUPOLA HANDBOOK supply is practically exhausted. In spite of the remarkable success which this book has had there are a number of chapters which demand revision. Steps will be taken to complete this revision on the same basis as was used in the original publication, i.e. the assignment of a subject to a man or men best qualified to handle the particular subject. This work will progress as rapidly as is practical.

The success of this undertaking, the research and the book depends largely on the broad participation of the members of the Society. Your cooperation is therefore solicited for both phases.

\* Vanadium Corporation of America, Detroit

# STRESS RELIEF OF GRAY CAST IRON

By  
J. H. Schaum \*

## ABSTRACT

*This report describes a number of experiments on the stress relief of gray cast iron. One set of experiments consisted of making relaxation tests and using the rate of relaxation as a means of evaluating stress relief. These studies showed that (1) the rate of stress relief is most rapid during the first hour at temperature, (2) rate of stress relief is very slow below 1000 F but increases as the temperature is raised, and (3) initial stress and alloy composition have an important influence on the rate of stress reduction by heat treatment. Observations of heat treatments on highly stressed cast wheels revealed that (1) indoor aging and low temperature (600 F) heat treatment are ineffective for stress relieving, (2) furnace cooling after heat treatment produces lower residual casting stresses than air cooling, and (3) relaxation tests agree closely with results of stress relief heat treatments of experimental castings.*

## Introduction

WITH INCREASED USE of gray cast iron in highly stressed mechanical parts, more attention is being directed toward adequate stress relief before the castings enter into service.

Internal stresses are generated in castings by differences in rate of contraction of various sections due to temperature gradients set up during cooling. Such temperature differences usually arise from variations in cross section and distance of metal from heads and gates. The resulting stresses are often large enough to cause cracks which may or may not be detected before the castings go into service, but in most instances the stresses are not large enough to cause the casting to fail until the added increment of a normal service load is applied. Internal stresses may also lead to distortion during machining or in service since the removal of metal by machining disturbs the balance of the internal stresses and causes distortion to take place in such a way as to restore this balance. The relief of these internal forces on heating is generally considered a creep phenomenon in which the stress is reduced through plastic flow.

This report describes the results of a series of stress

relief tests made on the Naval Research Laboratory relaxation machine and on a large number of cast wheels containing high stresses.

**Statement of Problem.** The objective of this work was to develop fundamental information on the stress relief of gray cast iron.

**Known Facts and Theoretical Consideration.** The past studies of stress relief heat treatments for cast iron have resulted in a series of conflicting opinions and observations. The extent of this confusion will be readily appreciated from the following summary of a number of the papers written on this subject.

Prior to 1915 practically no investigations on the heat treatment of cast iron were reported. In the year 1915, however, an anonymous author reported that casting strains in small hardware parts could be relieved by heating for 8 hr at a dull red heat (1100 F), followed by very slow cooling for about one day.<sup>1</sup> Castings so treated were softened due to precipitation of some of the combined carbon as graphite.

In 1917, L. M. Sherwin<sup>2</sup> found that equivalent strain removal was obtained by heating a series of low-silicon cast irons at either 1100 F or 400 F for 24 hr followed by slow cooling.

According to C. J. Wiltshire<sup>3</sup> casting strains were completely eliminated by heating slowly to 700 F, holding for 7 hr, and cooling to 300 F in 20 hr.

Harper and MacPherran<sup>4</sup> relieved casting strains by heating at 1150 F for 1 hr. With longer holding times the strength and hardness decreased materially.

Rolfe<sup>5</sup> heated iron in the temperature range of 750 F to 1830 F for 1 hr and cooled slowly in hot sand. He concluded that casting strains could be removed by heating to 1110 F for 1 hr followed by slow cooling. Tensile strength was reduced only 6.5 per cent and hardness 2.5 per cent.

J. W. Bolton<sup>6</sup> observed that strains could be removed by heating slowly to 700-1000 F and furnace cooling. In a later paper,<sup>7</sup> he recommended 700-800 F.

Grotts<sup>8</sup> recommended a 600 F anneal for quick strain removal, and J. A. Capp<sup>9</sup> advised that iron castings be stress relieved by heating to 930-1020 F for a period of 4 to 10 hr.

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Benson and Allison<sup>10</sup> found complete stress relief after a 6-hr soak at 1110 F but cautioned that increase in dimension due to oxidation commences at 1020 F and becomes serious at 1110 F.

Hurst<sup>11</sup> recommended a stress relief temperature of 750 F. He stated that pearlite decomposition commenced at 840 F and cautioned against heating above 930 F.

According to Morken<sup>12</sup> a normal treatment for relief of internal stresses consists of holding 4 hr at 900 F. Because of the graphitizing tendency of silicon, he stated that it is advisable to use lower temperatures for high-silicon iron and higher temperatures for low-silicon irons.

Le Thomas<sup>13</sup> stated that the French Admiralty specification for stress free cast iron requires a "prolonged heating to a temperature of the order of 930 F, but lower than 1155 F, followed by extremely slow cooling. The total duration will be from 24 to 48 hours."

In a recent publication, P. H. Russell<sup>14</sup> arrived at the following conclusions in his research on the subject of stress relief:

1. Exposure for four months to varying atmospheric conditions and temperatures reduced internal casting stresses about 15 per cent.
2. Heat treatment at 750 F was ineffective.
3. Temperatures in excess of 975 F were required to relieve 50 per cent of the internal stresses.
4. Internal stresses were not completely relieved at 1110 F.
5. The physical properties of the irons were not impaired by treatment up to 1110 F, except, that some falling off in transverse strength was occasionally observed.

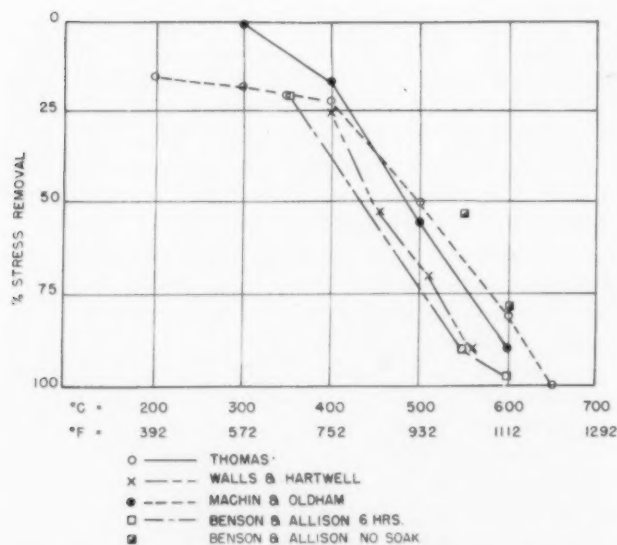


Fig. 1—Observations of several investigators on the effect of stress relief temperature on per cent stress removal.

Figure 1, taken from a report by P. H. Russell,<sup>14</sup> depicts graphically the observations of a few additional investigators on the effect of stress relief temperature on per cent stress removal. These excerpts

from numerous papers are indicative of the differences of opinion on proper stress relief heat treatments. Recommended temperatures range from 400 F to 1150 F and soaking time from 1 to 24 hr.

### Experimental Procedure

In the first phase of this work, the effect of time and temperature on the relief of internal stresses was studied with the aid of a special apparatus built at the Naval Research Laboratory, known as a relaxation machine. This machine, shown in Figs. 2 and 3, consists of three essential parts—the furnace, the loading mechanism, and the extensometer control system.

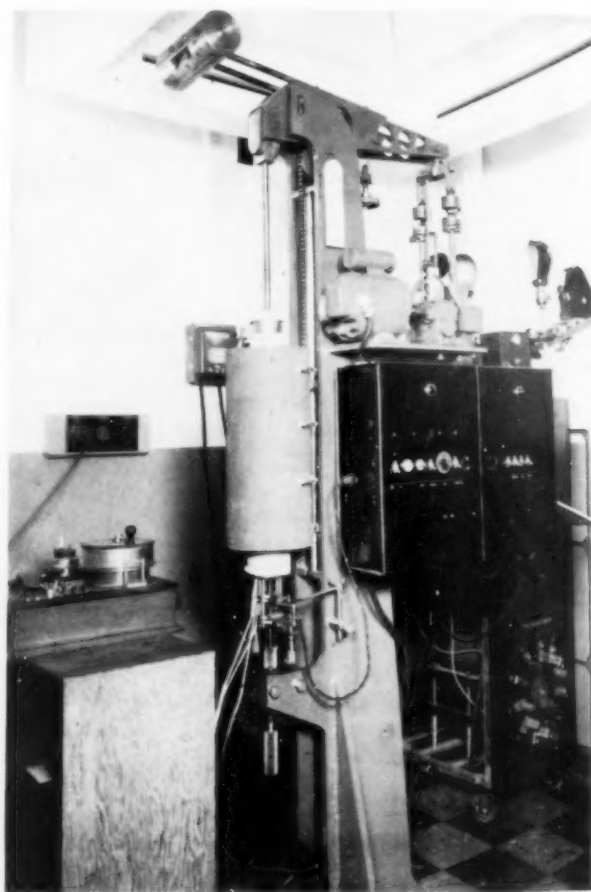
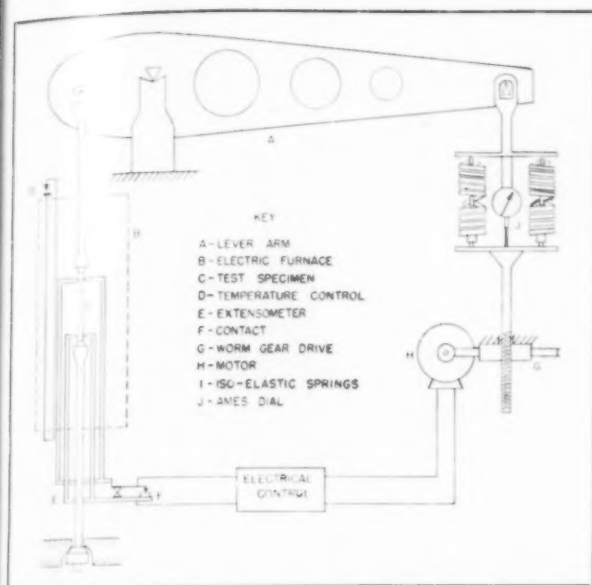


Fig. 2—Relaxation testing machine.

A test specimen, of the type shown in Fig. 4, is held by threaded grips in an electric furnace, B, and heated to any desired temperature. Temperature controller, D, maintains the specimen at constant temperature for the duration of the test. The specimen is next stretched a definite amount by applying a tensile stress through the lever arm, A, by means of G and H. The elongation is measured with extensometer, E, and the load with Ames dial, J. The machine is then set for automatic operation. As the specimen stretches plastically, due to the load and temperature imposed on it, the extensometer contact, F, closes and starts the motor, H, which reduces the applied stress just enough for the specimen to contract elastically and

Fig. 3—Sketch of relaxation machine.



This 4-spoked wheel casting was adopted because large stresses were present in the as-cast condition as a result of the thin spokes cooling faster than the heavy rim. Tensile stresses are first set up in the spokes tending to pull them away from the rim. Since cast iron will flow readily at low stress levels and high temperatures, the spokes stretch to relieve the stress. The spokes, being colder, reach room temperature and stop contracting before the rim, which continues to contract and exert a compressive stress on the spokes. The magnitude of stress retained in this wheel before and after heat treatment was determined by drilling reference marks approximately 2 in. apart on the rim and spoke and measuring the distance between these marks to within 1/10,000 of an inch with a Whittemore strain gage. The rim was sawed on both sides of the reference spoke so that both the rim and spoke

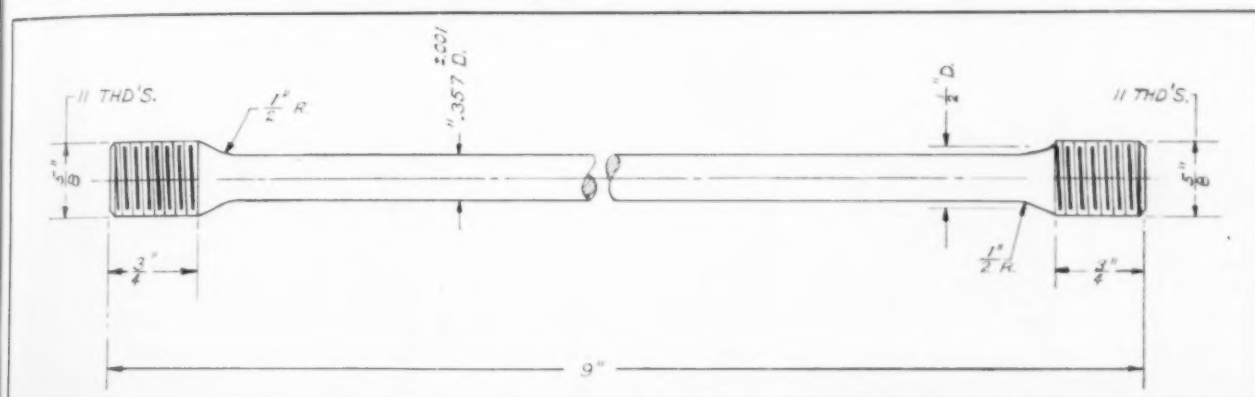


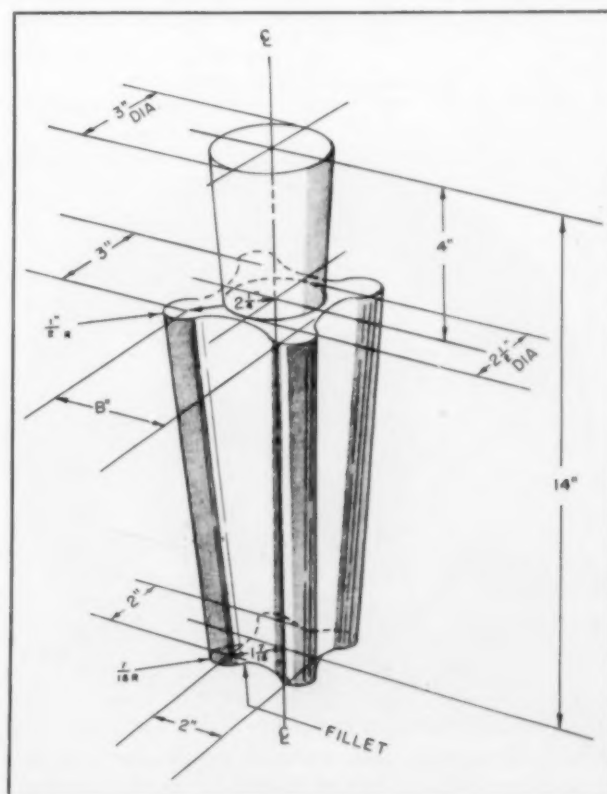
Fig. 4—Relaxation test piece.

open the extensometer contact, stopping the motor. The above cycle continues to be repeated and the load on the specimen is gradually reduced until a stress level is reached at which the rate of plastic flow is so slow that no perceptible change is observed for several hours. Thus the specimen is held at constant length throughout the test. A movie camera equipped with an automatic device takes single frame pictures of the load indicating dial and a clock underneath it so that a time versus load record is obtained. A detailed description of the relaxation equipment is contained in a paper entitled "Stress Relief of the Steel Casting" by E. A. Rominski and H. F. Taylor.<sup>15</sup>

The relaxation specimens were made from a cast ingot of special design developed at the Naval Research Laboratory (See Fig. 5). The four rounded corners were removed by sawing longitudinally and then machined as shown in Fig. 4.

Relaxation rates of these specimens were compared with stress reductions effected by heat treatments of the special type of casting shown in Fig. 6.

Fig. 5—Ingot from which relaxation test specimens were made.



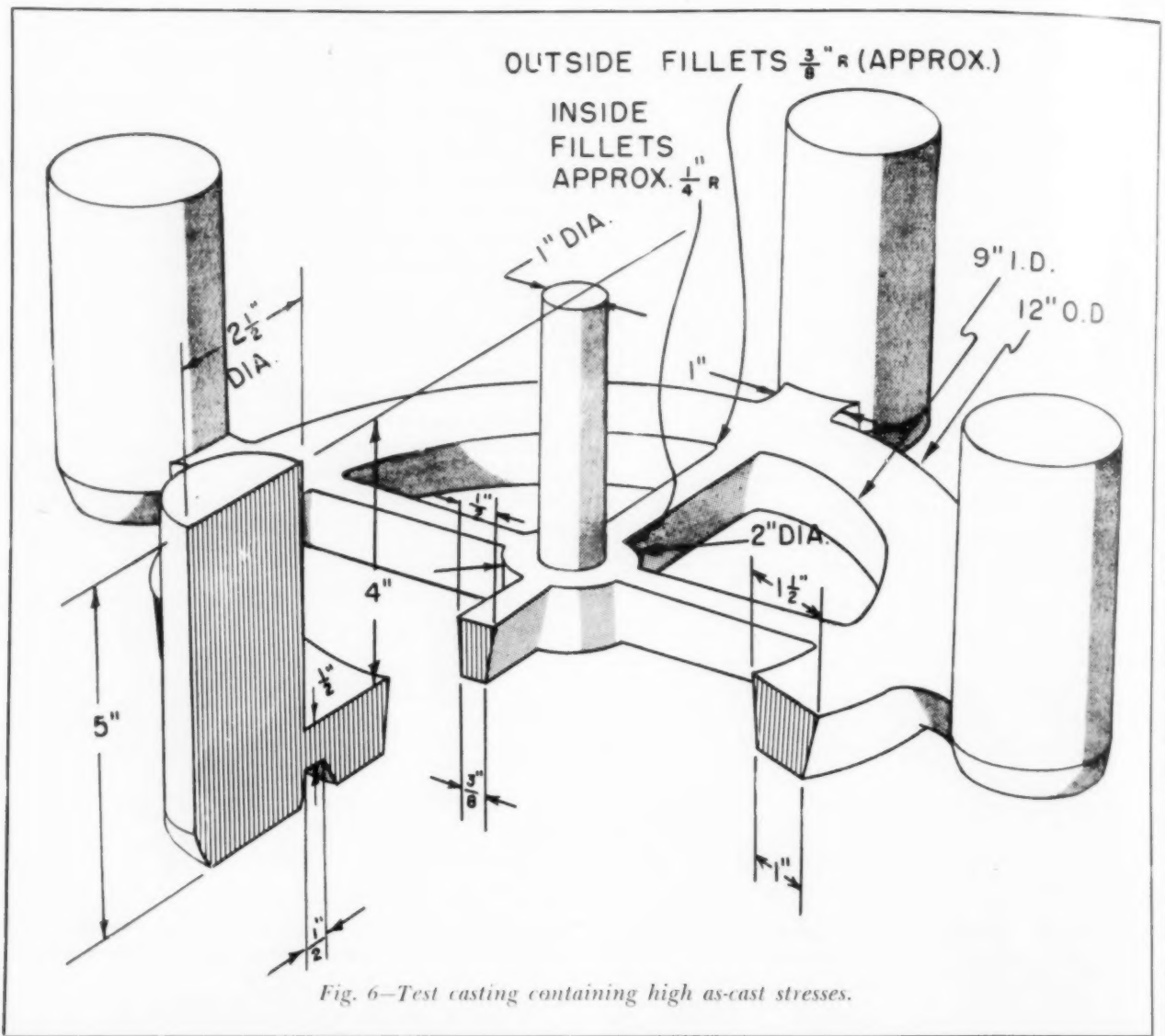


Fig. 6—Test casting containing high as-cast stresses.

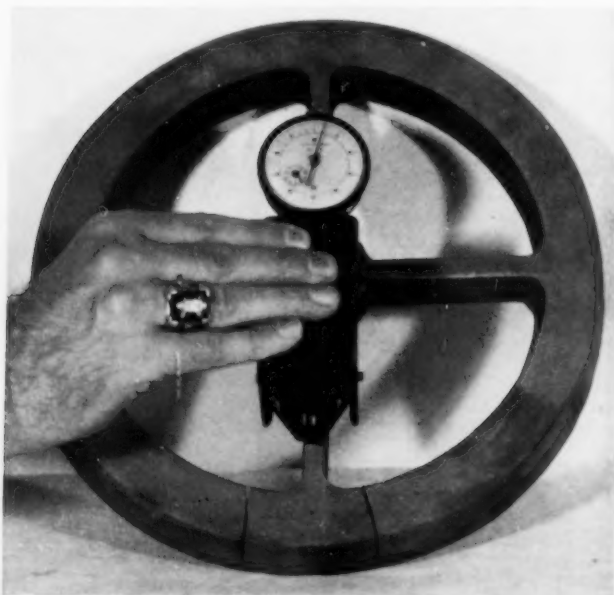


Fig. 7—Measuring strain in stress wheel after saw cuts were made.

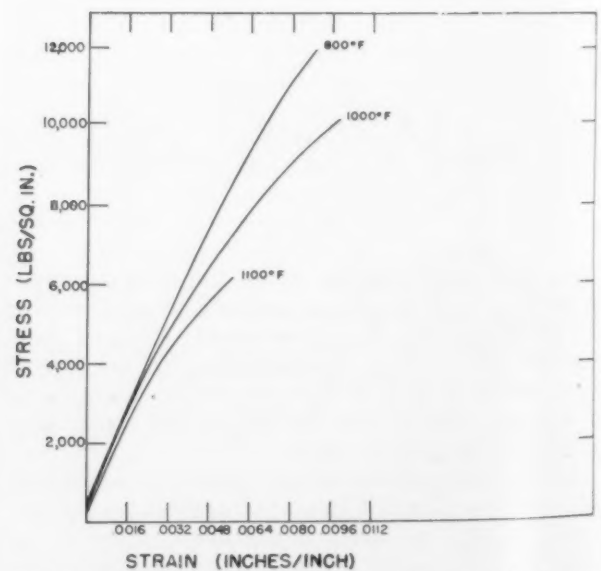


Fig. 8—Effect of temperature on stress-strain curves obtained when specimens were loaded in relaxation machine.



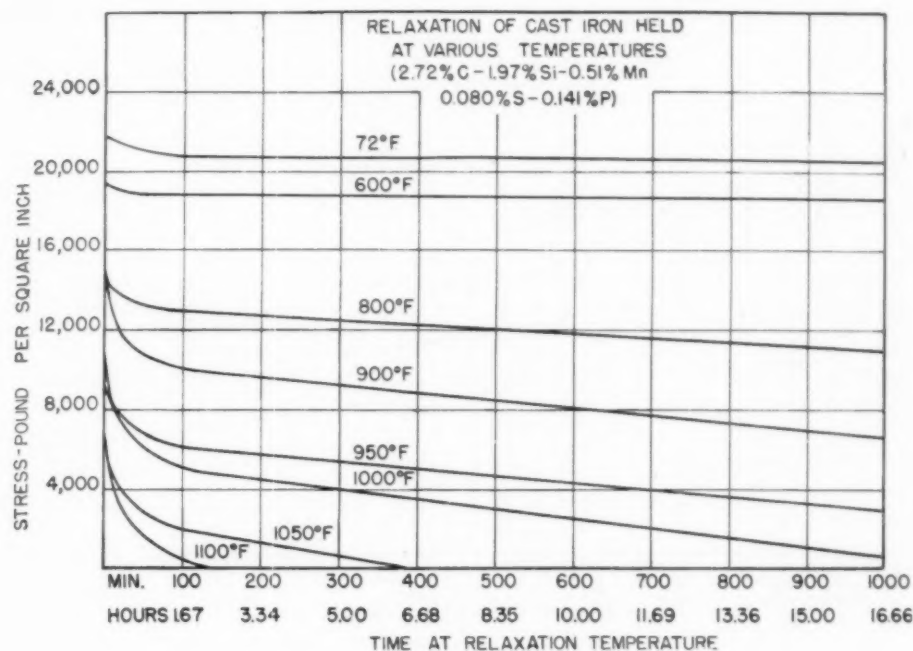


Fig. 9—Relaxation of cast iron held at various temperatures.

were free to move in such a manner as to relieve the internal stresses. The distance between the reference marks was then measured again and the net change indicated the degree of internal stress. Figure 7 shows the position of the reference marks, the saw cuts, and a measurement being made on the spoke with the gage. Since the expansion was much greater across the saw cuts in the rim than along the spoke, the former measurements were used in tracing the removal of stress by heat treatment.

The first experimental work consisted of a series of relaxation tests on a heat poured at the Naval Research Laboratory to determine the general effect of time and temperature. This was followed by relaxation tests on bars of different chemistry supplied by the Gray Iron Founders' Society to find the influence of chemistry. The analysis of these bars are given in Table 1. Stress relief studies were then made on some

stress-wheel castings poured at the Naval Research Laboratory to determine what effect the cooling rate from the heat treating temperature might have, and then stress-wheels supplied by the Gray Iron Founders' Society and poured from the same heats as the relaxation bars were heat treated using the cooling rate that was found to introduce the minimum of stress.

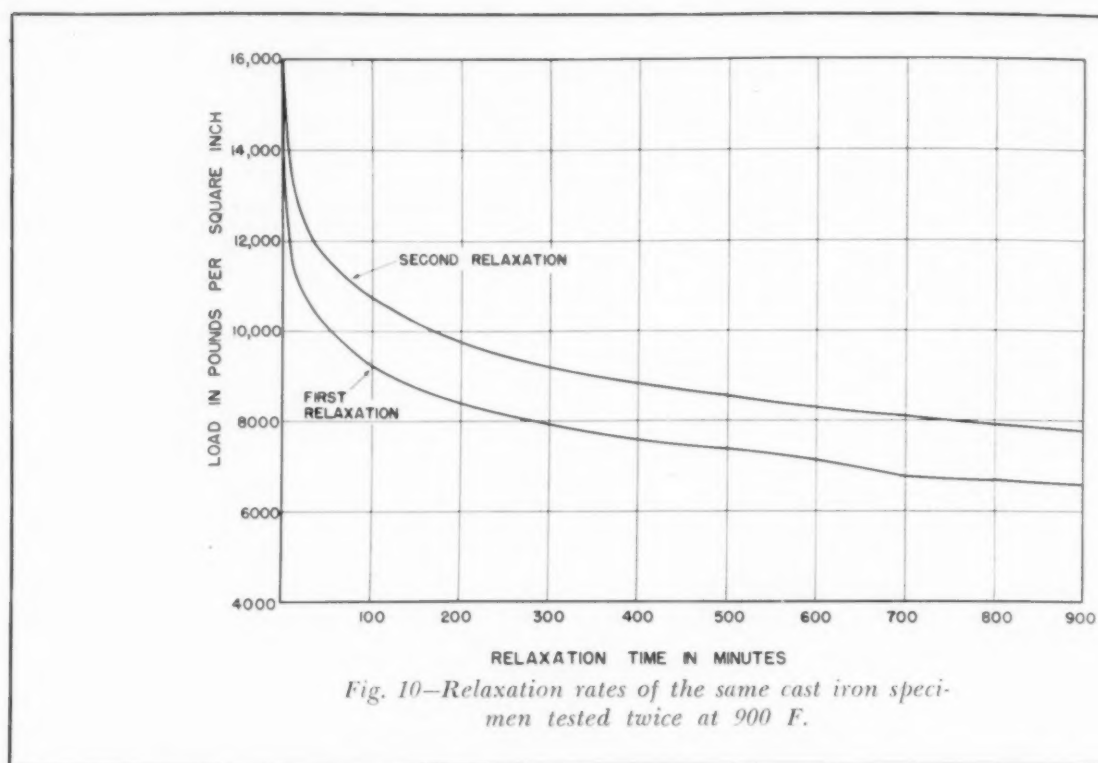
### Discussion of Experiments

**Relaxation Tests.** An investigation of effects of temperature and time on relaxation was conducted on some Naval Research Laboratory experimental foundry sand cast ingots containing 2.72 per cent C-1.97 per cent Si-0.51 per cent Mn-0.080 per cent S-0.141 per cent P. Relaxation specimens were loaded to a value which produced between 0.2-0.4 per cent strain on the 6-in. gage length. As would be expected, the higher the testing temperature, the lower was the applied stress required to produce this amount of strain. Typical loading curves for three different temperatures are given in Fig. 8. As the temperature increases the strain for any given stress increases; since gray iron has no definite yield point but instead a small plastic strain component at all stress levels, this strain represents both elastic and plastic deformation.

Duplicate specimens were tested at temperatures ranging from 72 F to 1100 F. The effect of testing temperature on the rate at which the applied stress was reduced is shown in Fig. 9. Low temperature treatments had only a slight effect on lowering of stresses; 16 hr at room temperature had practically no effect. Treatment at 900 F only reduced the stress from 14,000 to 10,000 psi in 100 min, but when the

TABLE 1—CHEMICAL ANALYSIS OF GROUPS OF RELAXATION SPECIMENS AND STRESS WHEELS

Group Number	C	Si	Mn	S	P	Ni	Cr.	Mo
1	3.21	1.74	0.70	0.094	0.111	1.05	0.15	0.53
2	3.82	2.19	0.85	0.044	0.115	0.10	0.47	0.42
3	3.58	2.24	0.64	0.092	0.140	0.10	0.08	0.04
4	3.41	2.33	0.62	0.124	0.244	0.10	0.03	0.04
5	3.60	2.49	0.64	0.076	0.378	—	—	—
6	3.51	2.15	0.69	0.113	0.138	0.25	0.25	0.04
7	3.29	2.18	0.97	0.135	0.112	0.25	0.25	0.05
8	3.16	1.70	1.13	0.109	0.064	0.10	0.10	0.03
9	3.00	1.65	0.92	0.086	0.057	1.44	0.28	0.50
10	3.47	1.90	0.89	0.089	0.097	0.10	0.20	0.04
11	3.01	2.36	0.70	0.095	0.344	0.20	0.05	0.04

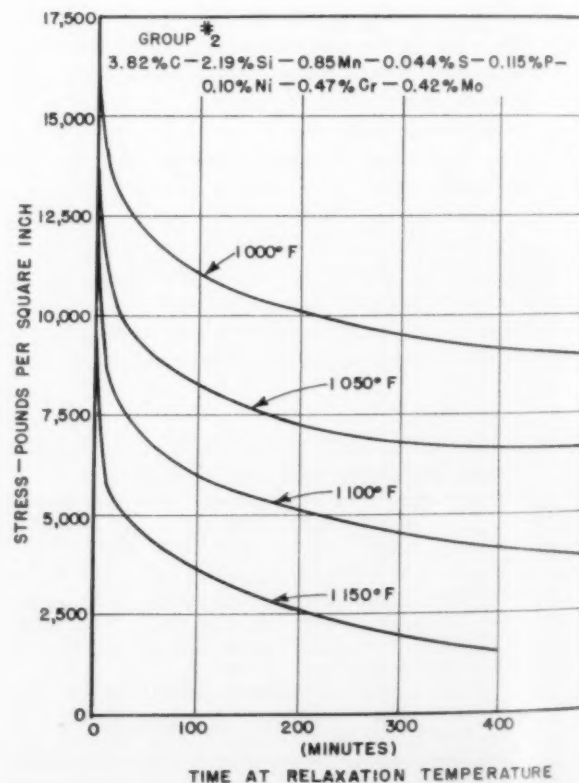
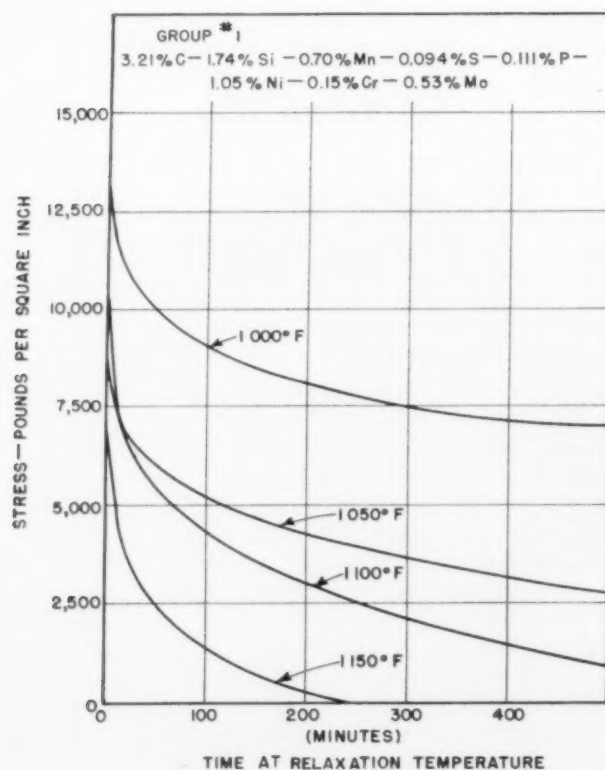


testing temperature was raised to 1100 F, essentially complete relaxation took place in approximately 1½ hr. In all cases the rate of stress relief was greatest during the first hour at temperature, and then decreased as the time at temperature increased.

Of special interest was a test in which a specimen

was reloaded to the original stress level and given a second relaxation test. As may be seen in Fig. 10, the second relaxation rate was slower than the first indicating a decreased rate of plastic flow. Thus only one relaxation test could be made on each specimen.

Since these tests showed the ineffectiveness of low



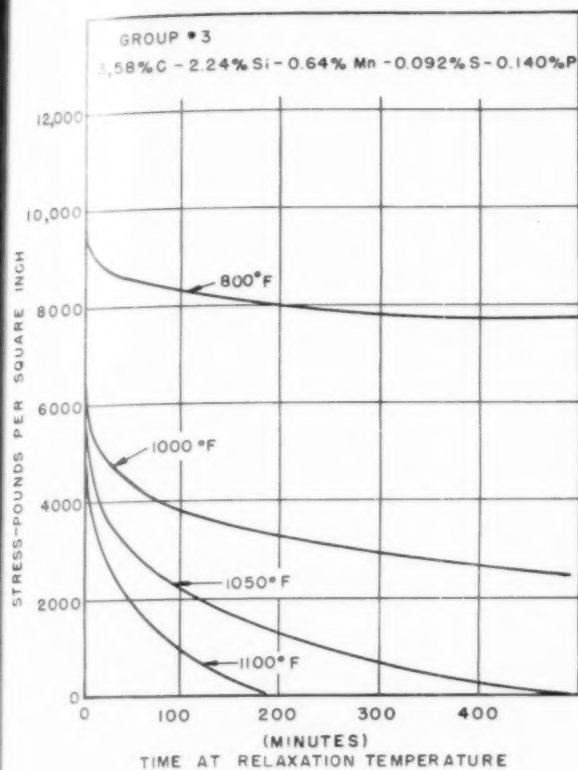


Fig. 13

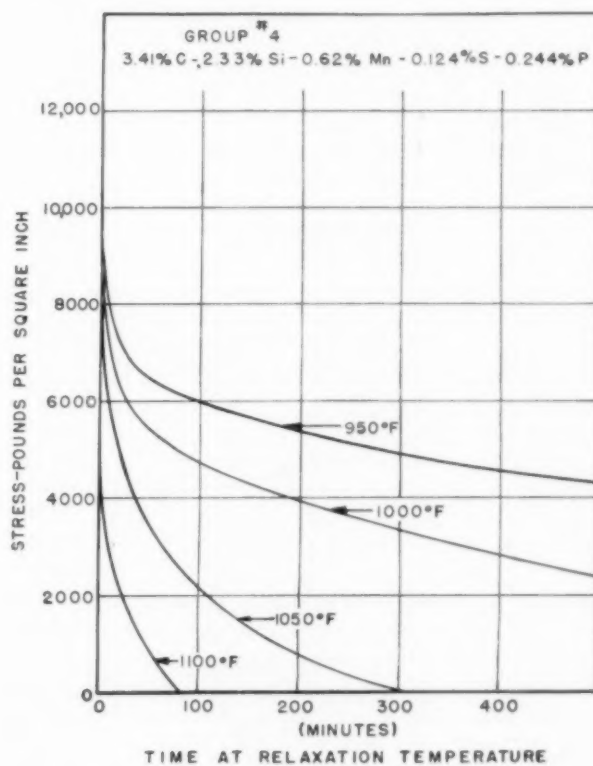


Fig. 14

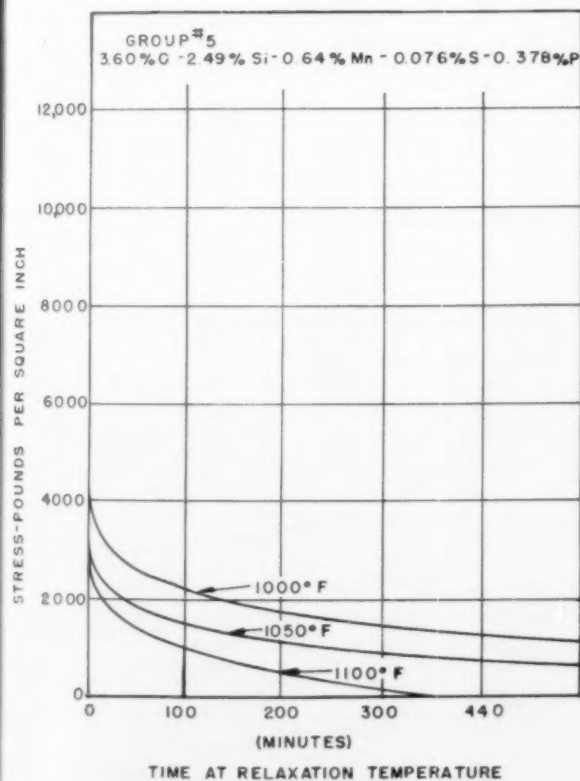


Fig. 15

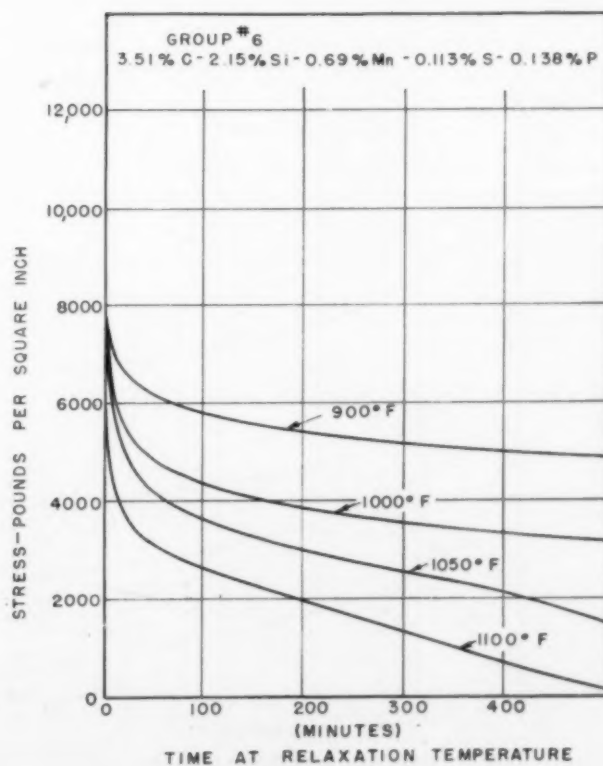


Fig. 16



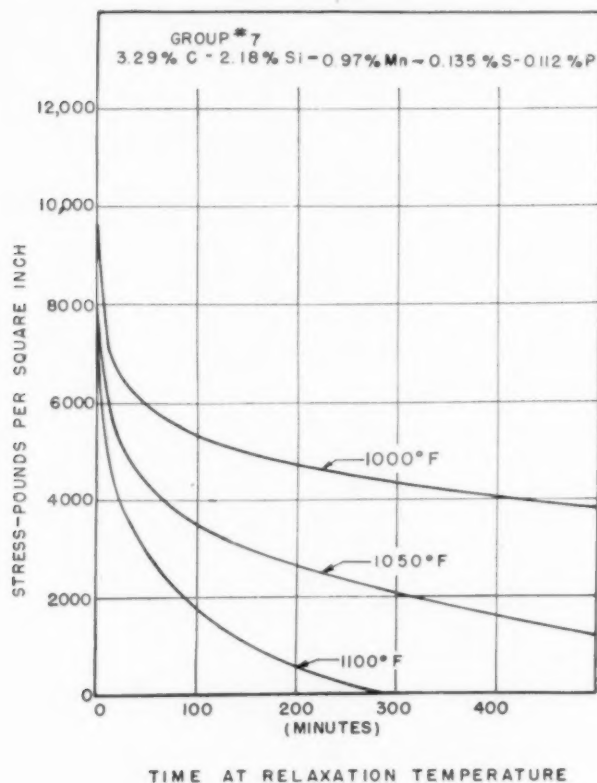


Fig. 17

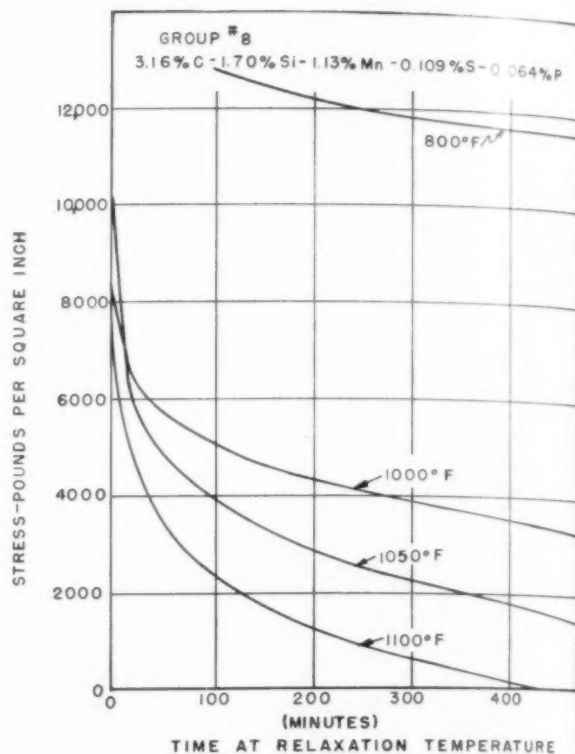


Fig. 18

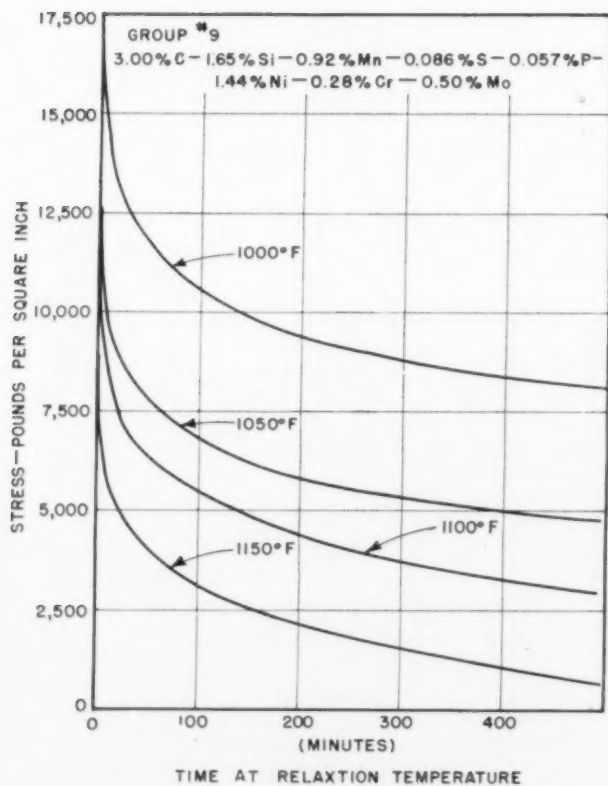


Fig. 19

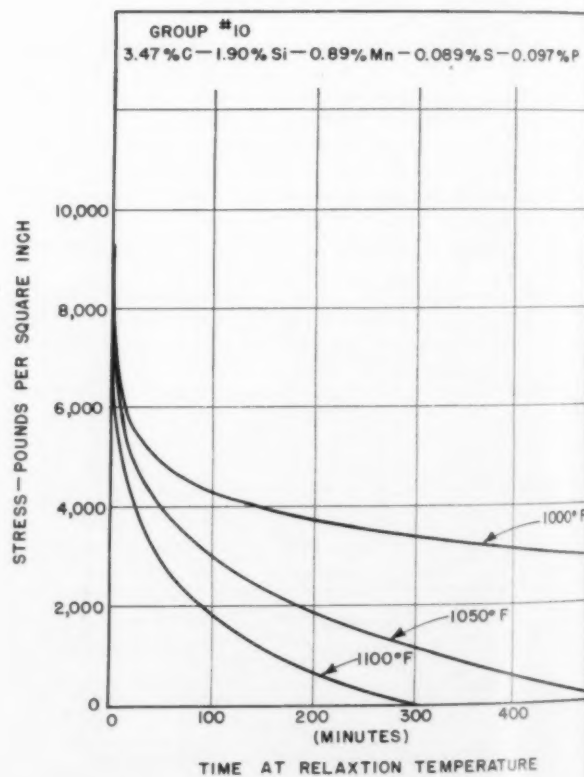


Fig. 20

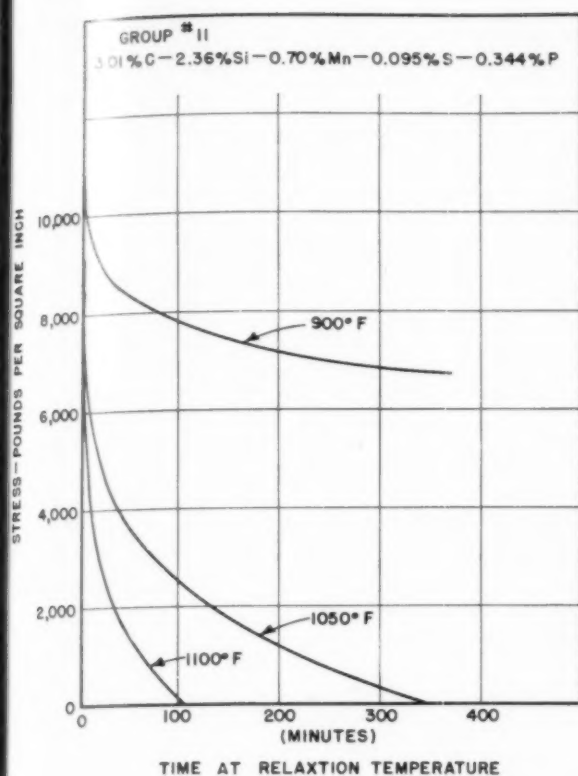


Fig. 21

temperature heat treatment of stress relief, the commercial sand cast ingot specimens were tested in the temperature range of 800 F-1150 F. Figures 11 to 21 show the relaxation rates for the 11 groups of commercial cast irons at various temperatures. The following general observations may be drawn from these graphs:

1. By comparing the relaxation curves of the alloy irons—Groups 1, 2 and 9 (Figs. 11, 12, and 19)—with the plain carbon irons (Figs. 13-18, 20 and 21), it is apparent that the former require higher temperatures

and/or longer holding times to reach the same stress level or the same percentage of stress relief.

2. No correlation was found between relaxation rate and equivalent carbon.

3. Figure 20 shows that if duplicate specimens have the same initial applied stress, the relaxation rate and completeness increase with testing temperature.

4. In Fig. 22, three of the specimens were loaded to different stress levels but tested at the same temperature, i.e., 1050 F. Comparing these three curves it is evident that for any given temperature the higher the initial stress the faster the relaxation rate but the longer the time required to reach any specific residual stress level.

Figure 23 has been constructed from data contained in Figs. 11 to 21. The various points on the graph show the original applied stress and the stress remaining after 2 hr at the various testing temperatures. The two curves bound the maximum and minimum points. Stress is lowered slowly and incompletely at temperatures below 1000 F, but rapidly and more completely at higher temperatures. The spread of data results from the differences in chemical analyses and in initial maximum stress. Since the modulus of elasticity varies for each alloy, there is unavoidable variation in the ultimate load for each test when loading to 0.2-0.4 per cent strain on the 6-in. gage length. Nevertheless, the general trend of an increasing rate of stress relief at higher temperatures is clearly shown.

The relaxation data are summarized in a different manner in Fig. 24. The points on this graph represent the initial stress and that remaining after the specimens were held at 1050 F for the indicated time. The reasons for the spread of data have been previously given. The curves of maximum and minimum stress show the rapid initial rate of stress relief and the leveling off after the first hour at temperature.

#### Stress Wheel Tests

Before conducting stress relief heat treatments on the limited number of stress wheels received from the

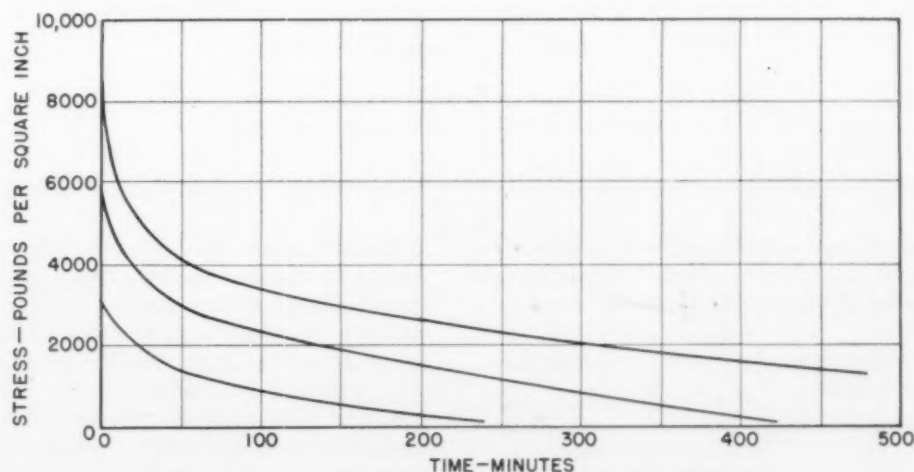


Fig. 22—Relaxation rate of gray cast iron at 1050 F showing effect of initial stress.

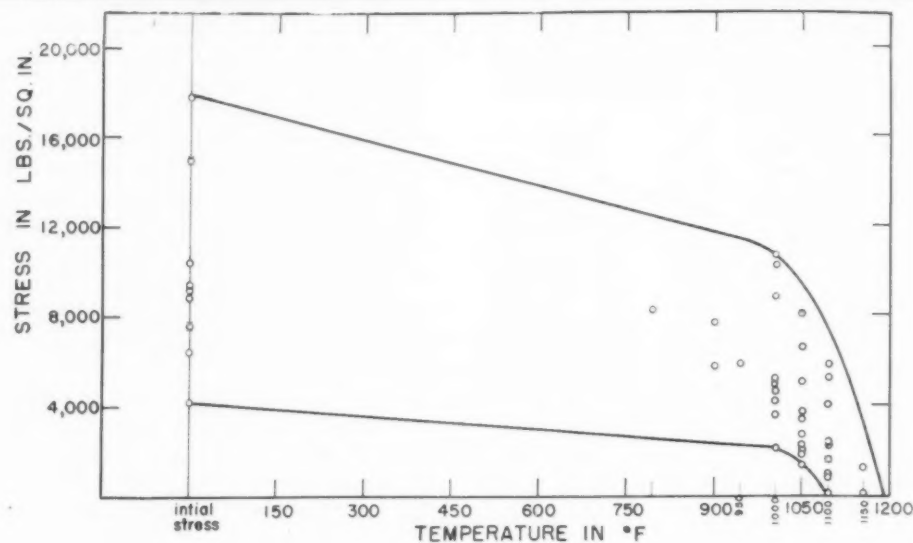


Fig. 23—Residual stress in relaxation specimens at the end of 2 hr at the indicated temperature.

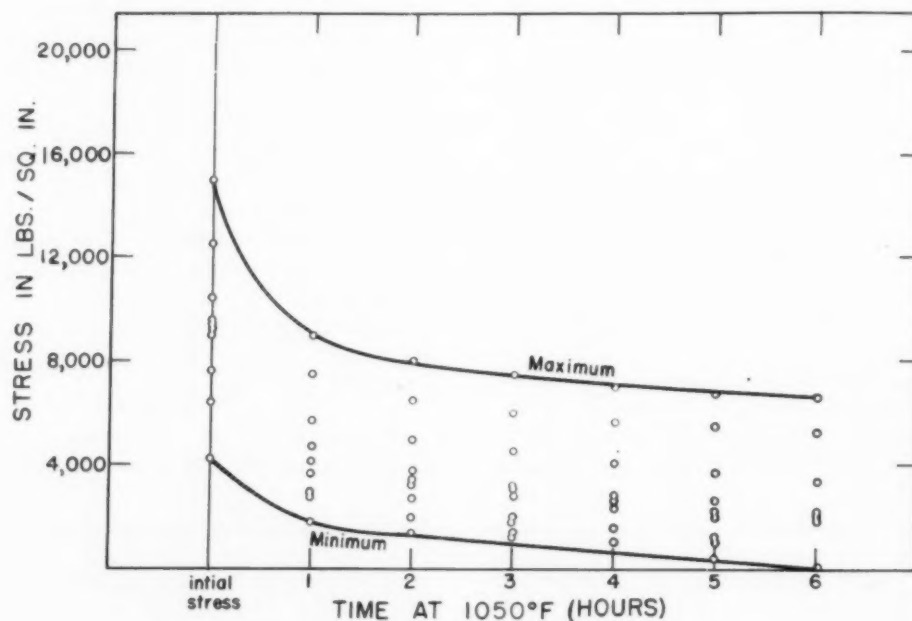


Fig. 24—Residual stress in relaxation specimens heat treated at 1050 F for the indicated time.

commercial foundries, an experiment was conducted on Naval Research Laboratory wheels to determine the effect of furnace vs air cooling from the stress relieving temperature.

The following strain gage measurements were obtained in this experiment:

Heat Treatment of Wheel	Rim Strain* (In.)	
	Air Cooled	Furnace Cooled
1 Hr at 1050 F	0.0089	0.0034
24 Hr at 1050 F	0.0052	0.0019

\* As-cast rim strain—0.0183

The difference between the residual stresses in air cooled and furnace cooled castings results from the unequal cooling rates existing between heavy and thin sections when air cooled; the slower furnace cooling keeps all parts of the casting at approximately the same temperature during cooling. It is interesting to note that 1 hr at 1050 F followed by furnace cooling results in lower residual stress than 24 hr at temperature followed by air cooling. Thus air cooling may introduce considerable additional stresses. For this reason furnace cooling was adopted for all commercial stress wheels studied.



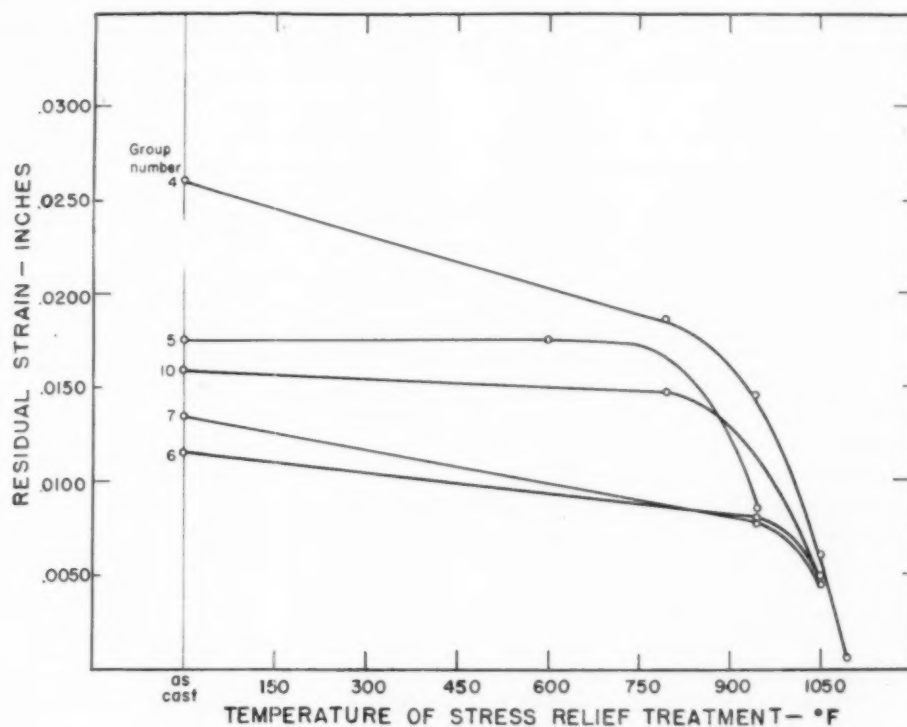


Fig. 25—Strain reduction effected by holding industry wheels at varying temperatures for 2 hr and furnace cooling.

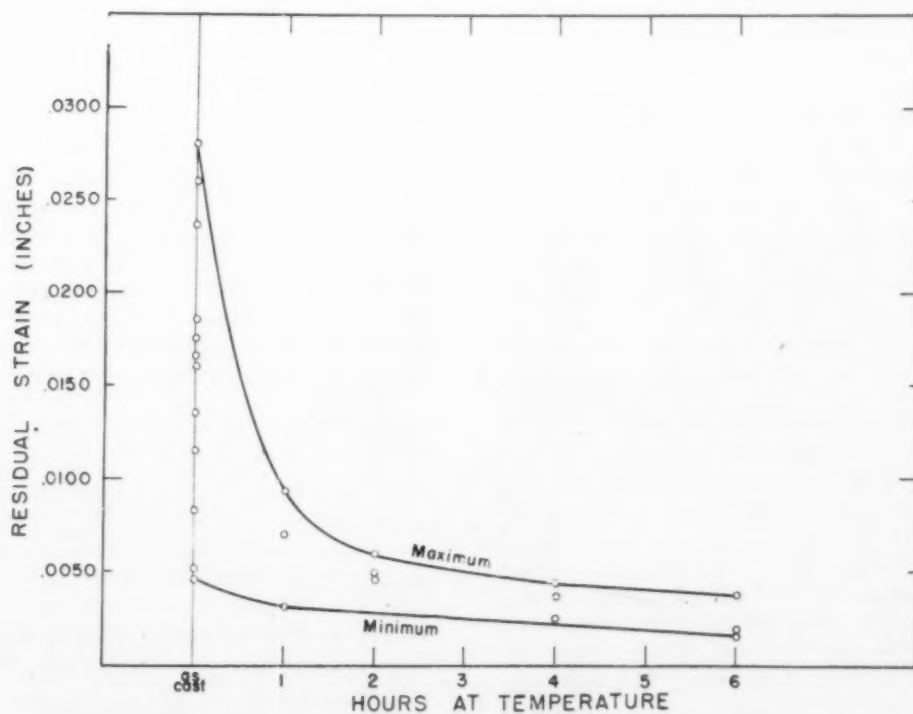


Fig. 26—Strain reduction effected by holding industry wheels at 1050 F for varying lengths of time followed by furnace cooling.

On the commercial wheels, studies were made to determine:

1. The strain reduction effected by two hours at various temperatures, and
2. The strain reduction effected by varying lengths of time at 1050 F.

The procedure followed in treating the wheels was to place them in a furnace at room temperature, heat to the desired maximum, hold for a specified period of time, and then furnace cool.

Although a casting may be held at the maximum heat treating temperature for only 1 hr, during the heating and cooling cycle the temperature of the casting is sufficient to effect further stress relief. Therefore the extent of stress relief in a casting held 1 hr at temperature will be slightly greater than obtained in a specimen held at similar temperature and time in the relaxation machine.

The data on effect of temperature variation at constant time are shown in Fig. 25. They are in agreement with the relaxation data in Fig. 23. The stresses in the rim are reduced only slightly by 2 hr heat treatment below 800 F. However, 2 hr. at temperatures above 950 F produce substantial reductions of stress.

In a test to show the ineffectiveness of long stress relief treatment at low temperatures, stress wheels of two different metal compositions were held at 600 F for 24 hr. Wheel A had an as-cast rim strain of 0.0174 in. and a final strain after heat treatment of 0.0168 in. Wheel B tested 0.0187 in. originally and its duplicate 0.0157 in. after treatment. Thus appreciable stress relief is not obtained at 600 F for one day.

The data on effect of time variation at constant temperature are summarized in Fig. 26. The temperature of 1050 F was selected because preceding experiments had indicated considerable stress relief action at this temperature. A marked reduction of internal stresses occurred during the first and second hours at temperature, after which the rate decreased rapidly in a way similar to the relaxation data given in Fig. 24. The spread of data has been previously explained.

The results of tests on the stress wheel castings appear to agree closely with those from the relaxation tests. The advantage of the relaxation test for obtaining stress relief data is apparent since one such test gives a complete stress-time curve while many wheel castings are required to yield the same information.

#### Stress Relieving By Other Methods

Since many foundries claim to have stress relieved castings by aging them outside for periods ranging from six months to a year, several tests were made in an attempt to evaluate this practice. In the first test, four stress wheels were poured from the same ladle of metal having a composition of 3.52 C, 1.93 Si, 0.46 Mn, 0.130 S, 0.148 P. One wheel was sawed the following day and the others were allowed to age inside the foundry for varying periods of time. The following table summarizes the effect of time on stress reductions:

Aging Time	Strain Observed in Rim by Sawing
1 day	0.0088 in.
3 months	0.0088 in.
1½ years	0.0089 in.

Indoor aging obviously did not reduce the as-cast stress in these wheels.

The effect of outdoor aging has been said to be due to the alternate heating and cooling of the casting. A rapid check on this theory was made by placing a stress wheel in a dry-ice box at 0 F for 2 hr and then transferring it to an oven at 220 F for 2 hr. After four complete cycles of alternate heating and cooling, the rim was sawed and 0.0191 in. of strain was measured. Before this treatment, a duplicate wheel showed 0.0209 in. of strain. Thus only a small amount of stress reduction was obtained by this treatment.

Besides the temperature changes which castings are subjected to in outdoor aging, the possibility of corrosion relieving a part of the stresses locked in the outer skin of the casting has been suggested. To determine the effectiveness of such corrosion, a rapid test was conducted by completely immersing a stress wheel in a 6 per cent solution of sulfuric acid for 20 min at a temperature of 165 F. Sawing of the wheel revealed a residual strain of 0.0150 in. compared with the as-cast strain of 0.0164 in. From this observation, corrosion might contribute slightly toward relief of casting stresses.

The manner in which stresses are generated in the test wheel were described earlier in this report. In addition to the stress induced by differential cooling, some may result from the resistance that the sand mold offers to the contraction of the rim as it cools. The possible magnitude of this effect was determined by ramming a pattern in the space between the spokes and rim so that relief cavities were produced with approximately ½-in. side wall of sand between the outer periphery of the cavities and the inner contour of the wheel. Two wheels were poured with and two without these cavities. The strain amounted to 0.0218 and 0.0222 in. on the two wheels without cavities and 0.0211 and 0.0177 in. on the wheels with relief cavities. The difference is not great but does show a trend in the expected direction. Another way to reduce the restricting effect of the sand is to shake out the castings shortly after pouring. That this practice might be helpful is shown by the data in the following table which lists the strain measurements made on five wheels poured from the same ladle of metal and shaken out at various time intervals.

Casting Number	Time Interval	Rim Strain
1	5 Min	0.0035
2	10 Min	0.0041
3	20 Min	0.0084
4	1 Hr 20 Min	0.0114
5	13 Hr 36 Min	0.0114

Up to about the first hour after pouring the strain increased with time, after which it remained constant, indicating that the resistance offered by the sand may increase the degree of internal stress.

#### Conclusions

##### A. Relaxation tests indicate

1. If the initial stress and the composition of specimens are the same, the relaxation rate and degree of stress reduction increases as the temperature is increased.

2. Rate of stress reduction is slow below 1000 F.
3. Rate of stress reduction is most rapid during the first hour at temperature and decreases as the time at temperature increases.
4. For any given temperature, the higher the initial stress the faster the initial relaxation rate but the longer the time required to reach any specific residual stress level.
5. Alloy irons require higher temperatures and/or longer holding times to reach the same stress level as plain carbon irons.

#### B. Stress wheel tests indicate

1. Two hours at temperatures below 800 F are ineffective whereas above 950 F substantial stress reductions are effected in the same time.
2. The greatest reduction of internal stresses occurs during the first hour at temperature.
3. Heat treatment at 600 F for 24 hr shows practically no stress reduction.
4. Air cooling after stress relief heat treatment may introduce considerable additional stresses which do not develop with furnace cooling.

#### C. Other Tests Indicated

1. Indoor aging is an ineffective method for stress relieving.
2. Alternate heating and cooling of stress wheels between 0 F and 220 F produces only a very slight stress reduction.
3. Corrosion of stress wheels with a mineral acid results in a very slight relief of internal stress.
4. The resistance of the mold sand may hinder the contraction of a casting in such a way as to materially increase the internal stress.

D. The relaxation machine gives a good indication of the stress-relief characteristics of gray cast iron and provides this information more easily than can be obtained from actual castings.

#### Acknowledgment

The author wishes to gratefully acknowledge the assistance of F. A. Brandt and W. O. Baker in conducting the relaxation tests and interpreting the data.

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#### DISCUSSION

Chairman: H. BORNSTEIN, John Deere & Co., Moline, Ill.

Co-Chairman: D. A. PAULL, Sealed Power Corp., Muskegon, Mich.

R. A. FLINN:<sup>1</sup> I want to congratulate the author on presenting data that we needed for a long time. Many of us felt that the old stress relief temperature ranges were entirely too low and that revision was badly needed. This paper certainly provides the required data.

There are two points I should like to bring up. I do not mean to confuse the paper but there are heat treating temperatures of 400 to 600 F, for example, for martensitic white irons that are spoken of as stress relief temperatures. Also heat treaters of quenched alloy steels speak about stress relief at 300 to 400 F. I think we should enter for the record that these treatments accomplish microscopic stress relief in the martensite, and this paper covers the grosser types of stresses.

I know the author realizes the dangers of getting into structural changes along with this stress relief. It makes the data more complicated, but we have been doing some work recently, trying to balance stress relief temperature against change in combined carbon. I should like to recommend that the combined carbon on the specimens treated at 1000 F or above be checked. We recently have been dealing with two types of gray irons, one with approximately 3.40 per cent total carbon and about 2 per cent silicon. The combined carbon in that material begins to drop off rapidly at about 1050 F. If you go into the ferritic structure you can expect a more rapid relaxation rate. The other iron is a lower carbon iron with approximately 3 per cent total carbon and about 1.80 per cent silicon. In that case you do not get the change in combined carbon until you get to approximately 1100 F, but there again where you have a stronger iron you need a higher stress relief temperature.

I would like to discuss the role of alloys and show that alloyed irons require a longer time or a higher temperature, or both. In reviewing those data we should see where we can preserve the as-cast tensile strength in these alloyed materials and still bring about stress relief. That is, if you make a casting to a 50,000 psi tensile specification, and then stress relieve it, I feel that after stress relief it should still be a 50,000 psi tensile iron. Therefore, the role of alloys should be carefully considered.

CHAIRMAN BORNSTEIN: On one of our machines we had the problem of reducing the internal stress in the wheels. The wheel is a built up structure. The rim and the spokes are of steel and the gray iron hub is cast in. We found that we could not get adequate stress relief at the usual temperatures up to 1000 or 1050 F. The way we were able to solve that problem was to put localized heating on the spokes and get plastic deformation there to relieve the stress. In that case, instead of the type strain gage used here, we used SR-4 strain gages. We found it generally true that with closed structures, such as wheels, it is difficult to find adequate stress relief at lower temperatures. In this particular case we had to go up to fairly high temperatures with very short time intervals, heat a short section of the spoke and get plastic deformation there in order to get stress relief in the wheel.

MEMBER: Did the author say that after taking the castings out and cooling them they would still retain about three times the strain? If you cooled the castings in the furnace, to about what temperature would you cool them? I am thinking of 15-lb castings.

<sup>1</sup> American Brake Shoe Co., Mahwah, N. J.



Mr. SCHAUM: Due to limited time and castings we did not make any attempt to establish the optimum temperature for withdrawing castings from heat treating furnaces. In most instances the furnace was about 300 F when the castings were removed. However, in your case I would recommend that you establish a practice by experimenting with some test castings. Results may vary with each alloy and with the nature of the casting. I imagine with certain castings that have no changes in cross sections you might be able to air cool them from 1050 F, but with castings that are likely to set up stresses during air cooling, furnace cooling would be advisable.

I realize that in industry you do not like to tie up your heat treating furnaces with slow furnace cooling operations, but in our experiments we did furnace cool to about 300 F.

W. J. BUCHE:<sup>2</sup> In the commercial test wheels referred to in this paper, was a specified shake-out time interval followed by the producers of the castings?

Mr. SCHAUM: No, there was not. We took them as we received them, in the hopes that they were all shaken out at approximately the same time.

Mr. BUCHE: What brings that question to mind is your statement that following heat treatment of the wheels the air quenching put back into the casting stresses that would have been relieved with a slow furnace cool. In another example you mentioned that the magnitude of casting stresses would be the same in one casting shaken out 5 min after pouring as would be found in the same casting if it had been shaken out 1½ hr later. Would it not be true that the 5-min shake-out would build up stresses due to air quenching that would not have resulted if the casting were left to cool slowly in the sand?

Mr. SCHAUM: That is a good point. We had only our experimental data to substantiate that observation. When we shook the castings out they were very hot. I suppose they were at least 1800 F, and they had to pass through a phase transformation certainly before they got down into a zone similar to our heat treating temperature ranges. Perhaps the explanation lies in the fact that in one case you are coming through a phase transformation and in the other case you are not. That is only a thought on it.

H. K. McGRATH (*Written Discussion*):<sup>3</sup> In Fig. 10, the difference was shown between the stress curve covering an initial heat treatment and one also of a second heat treatment. We do not clearly understand why the second heat treatment should produce a somewhat parallel curve to the first treatment if the

first treatment was a full stress relief. It would seem to us that the second curve should approach the appearance of a straight line. And along this same line of inquiry, were any more than the two heat treatments given? By that, we mean, do you have other curves for a third and fourth or fifth treatment?

Near the end of the discussion, a description was given of tests wherein the castings were taken out of the sand at various time sequences to determine the effect of such conduct upon the stresses set up within the individual castings. The question has arisen in our minds as to whether this difference in cooling rate had an appreciable effect upon the microstructure of the iron, and whether this difference of microstructure might reflect itself in the results produced and demonstrated.

It appears to us that the suggestion of R. A. Flinn that an investigation into the effects of the higher temperatures of stress relieving upon microstructure could greatly supplement the usefulness of the data presented.

Mr. SCHAUM (*Author's Closure*): I would like to thank Dr. Flinn and Mr. Bornstein for their pertinent remarks on this subject of stress relief.

Mr. McGrath has noted several parts of the report which should be clarified. In regard to his first question, I would like to emphasize that Fig. 10 summarizes the results of two consecutive relaxation tests on the same specimens. After the specimen had been held at constant length for 900 min at 900 F, it was reloaded to the original stress of 16,000 psi and the relaxation rate again measured. The ability of the cast iron to relax through the medium of plastic flow in this second test was reduced by the work hardening which took place during the first test. Repetition of the cycle a third and fourth time has revealed a continuation of the trend toward slower relaxation which indicates a reduction in plasticity.

Although we did not examine the microstructure of the castings which were shaken out of the sand at various time intervals after pouring, I am sure that as the castings were left in the sand for longer periods of time, the pearlite and graphite would become coarser. No work was done to evaluate the influence of microstructure on as-cast internal stresses. In this experiment it was theorized that early shake-out minimized the time interval during which stresses were generated in the castings by the resistance of the sand to contraction of the metal.

The author purposely kept the stress relieving temperatures as low as practical in order to minimize the influence of microstructural changes and accompanying alterations in physical properties.

<sup>2</sup> Brown Industries, Inc., Sandusky, Ohio

<sup>3</sup> Alten's Foundry & Machine Works, Inc., Lancaster, Ohio

# A STUDY OF FACTORS AFFECTING POURING RATES OF CASTINGS

By

J. G. Mezoff and H. E. Elliott

## ABSTRACT

Since one of the factors importantly affecting the quality of castings is their pouring rate, this work was undertaken to determine which factors of gating system design control the pouring rates of castings. While magnesium alloys were used in conducting the tests, many of the principles which were established would apply to the casting of any metal.

The effects on pouring rate of the following design factors were established: Sprue cross-sectional area, sprue cross-sectional shape, sprue length, sprue taper, sprue mouth design, and pouring basin depth. Also studied was the effect of pouring temperature. Results are in graphic form.

SELECTION OF THE PROPER POURING RATE for a given casting is important. Many quality factors may be adversely affected by the choice of a pouring rate which is either too fast or too slow. No less important than the selection of the proper pouring rate is the control of the rate for all molds made from the same pattern.

Work described in this paper was undertaken in an attempt to determine which factors of gating system design importantly affect the pouring rates of castings. Although the work was conducted using magnesium alloys, many principles were established which are applicable to any casting alloy. Since in practice most foundrymen control the pouring rate at the sprue, the present work centered attention chiefly on the effect of sprue and pouring cup design, although other foundry variables were also investigated.

## Experimental Work

In planning the work, the pattern described in Fig. 1 was designed. It consisted of a series of interchangeable sprues (A), a basin (B) which holds about 30 lb of AZ63 alloy (Al, 6; Zn, 3; Mn, 0.2 percent; remainder, Mg), a 4-in. diameter vent (C), and a boss (D). The sprue was molded in a position directly over the basin so that a gate or runner was not required. In this way the sprue itself controlled the pouring rate. A fabricated steel pouring basin, 5 in. deep and 10 x 5 in. in horizontal section was used. The sprue was located near one end of the basin.

For accurate timing an operation time recorder was

used. Copper wires with high-temperature insulation were connected to the recorder. The other ends of the wires were rammed-up in the mold in such a way that they were flush with the surface of the casting cavity, thus forming a point contact with the metal when it reached that position in the mold. In this way one contact was placed in the pouring basin, and the other in the boss on the casting cavity. The elapsed time between the moment the metal entered the pouring basin and the moment the casting cavity filled completely was measured. The pouring rate was calculated, using weight of complete casting minus that of vent.

The function of the vent was to minimize the "water-hammer" effect which tends to raise the cope of an unrisered mold. The lower contact was placed in a boss to avoid a possible premature contact caused by turbulence of the metal surface as it rose in the basin.

In all cases, unsuperheated commercial AZ63 alloy was used. This was melted in steel crucibles under crucible-type flux. Unless otherwise specified, the sprue length (cope height) in all cases was 12-in. and the pouring temperature 1400 F.

## Reproducibility of Pouring Rates

All of the data gathered in these studies are presented in graphical form. It was not considered necessary to present the voluminous data in tabular form; instead, the typical spread of values is indicated by plotting each test made as a single point (Figs. 2, 3, 4, 5, 6, 7 and 8). In all other graphs each point represents the average of data from at least three and in cases as many as 15 tests. In such cases, the spread of values obtained is not indicated.

A wide spread of values was observed in most of the test results. Under the best controls achieved, reproducibility of a result was only fair from one test to another. It is considered that the most likely cause of poor reproducibility lies in the pouring practice by which the metal is transferred from the crucible to the sprue. Pouring into one end of a rectangular pouring basin gives rise to considerable visible turbulence, especially at the beginning of the pour. Then too, it is difficult to duplicate exactly the effectiveness with which the pouring basin is flooded with metal at the beginning of a pour. Especially at the more rapid pouring rates were these problems troublesome. A

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H. E. Elliott, Metallurgist, The Dow Chemical Co., Bay City, Mich.

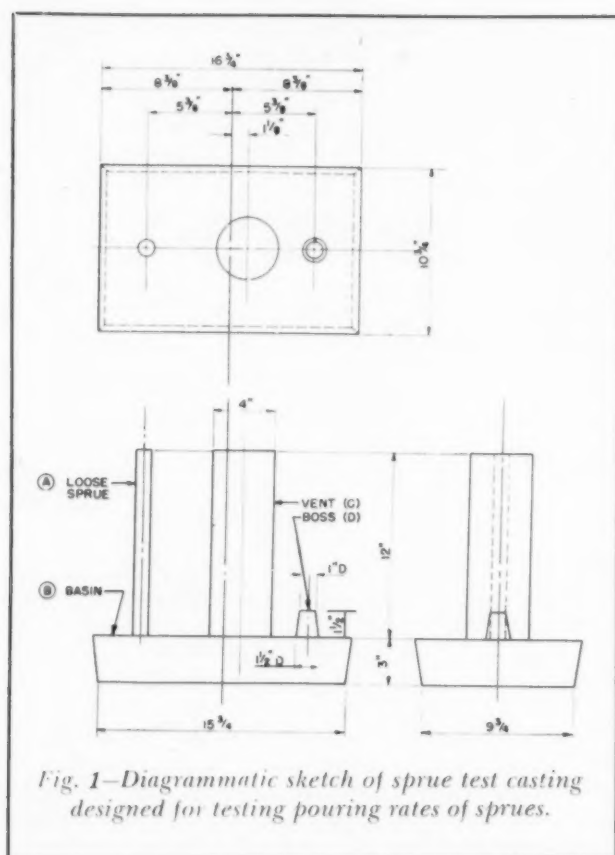


Fig. 1—Diagrammatic sketch of sprue test casting designed for testing pouring rates of sprues.

substantial part of the lack of reproducibility observed in these tests appears to be fundamentally a property of normal crucible pouring practice.

**Effect of Sprue Cross-Sectional Area on Pouring Rate**—The graph in Fig. 9 shows the effect of sprue cross-sectional area on the pouring rate of the test casting. These results were obtained using untapered sprues of various cross sections on the test casting, and measuring the resulting pouring rates. It is interesting to note that a straight-line relationship between sprue area and pouring rate, as has been suggested by other writers,<sup>2</sup> does not exist. As the sprue cross section increases, the area becomes less and less effective in delivering metal to the casting cavity. It is probable that the chief cause of this reduced effectiveness of area is the increased turbulence at the upper mouth of the sprue at the higher pouring rates. Another contributing factor may have been the greater difficulty with the larger sprue areas of flooding the pouring basin at the beginning of the pour.

It is also noticed in Fig. 9 that the round sprues poured faster than rectangular sprues of like cross-sectional area, and that slot sprues consisting of thin slots poured more slowly than those consisting of thick slots, total cross-sectional area being the same. In another paper<sup>2</sup> the authors have presented evidence that the restriction to flow offered by a round, untapered sprue is centered at its upper mouth, while with slot sprues, the sidewall resistance is also significant. It is apparent that, through some such mechanism, sprues having a large ratio of cross-sectional area to cross-sectional perimeter pour more rapidly than those having similar area but a lower ratio of area to perimeter. In Figs. 2 and 3 are plotted additional data showing

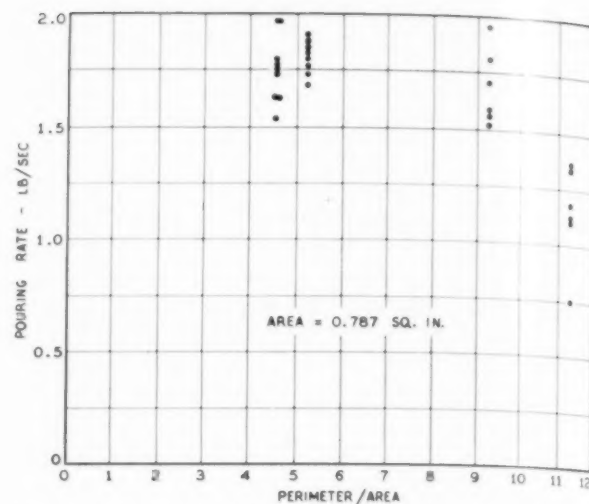


Fig. 2—Effect of perimeter/area ratio upon pouring rate of sprues having the same cross-sectional areas.

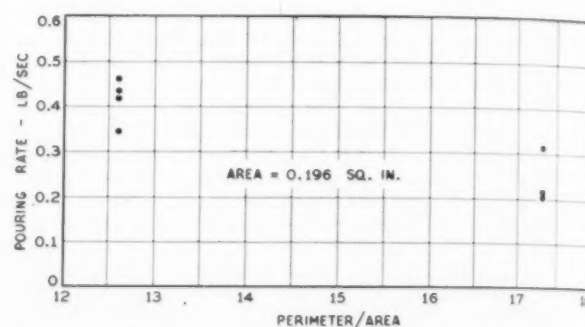


Fig. 3—Perimeter/area ratio effect upon pouring rate of sprues having the same cross-sectional areas.

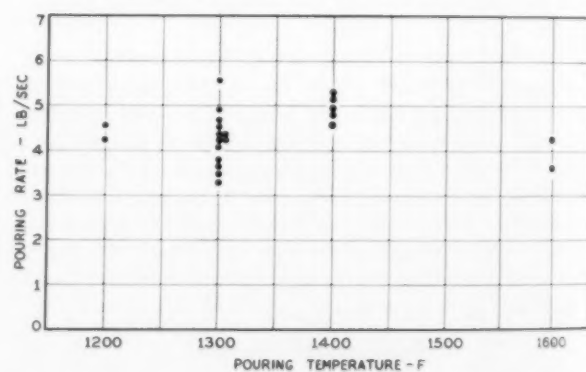
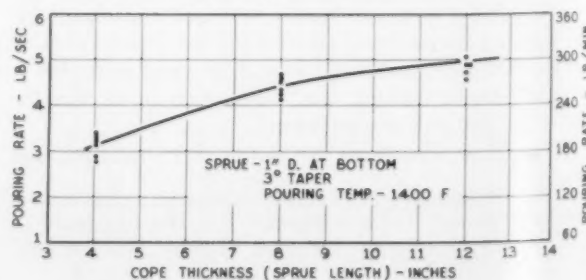


Fig. 4—Pouring temperature effect upon the pouring rates of sprues having diameters of 1 1/2 in.

Fig. 5—Sprue length effect upon the pouring rate of a tapered sprue placed small end down.





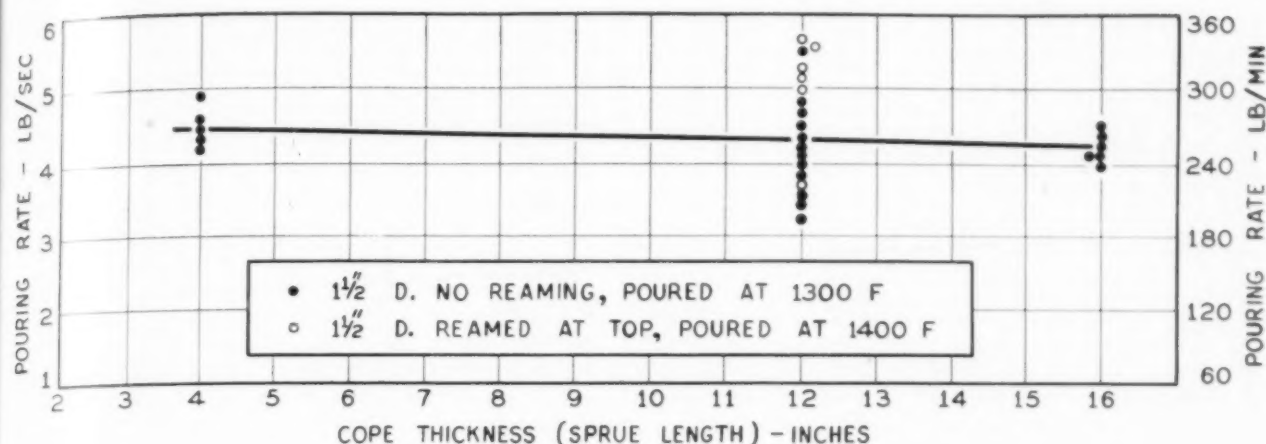


Fig. 6—Curve showing pouring rate of 1 1/2 in. diameter sprue as affected by the length of the sprue.

the effect of this area to perimeter ratio on the pouring rate for sprues of equal cross-sectional areas.

**Effect of Pouring Temperature on Pouring Rate**—Figures 4, 10 and 11 show the relationship between pouring temperature and pouring rate of various sprues. Figure 10 again shows the effect of the ratio of sprue area to sprue perimeter on pouring rate. A temperature range of around 1400 F appeared to give the most rapid pouring rate for a given sprue. A considerably higher or lower pouring temperature usually resulted in a slower rate of pour. The effect of lower pouring temperatures may be one merely of lower fluidity of the liquid metal. The effect of the higher temperatures in retarding pouring rates may be caused by increased evolution of gases from the mold material and increased reaction of these gases with the metal to form oxides. The effect of retardation of pouring rate at high pouring temperature was particularly marked when slot sprues were used. The entrapment of gases in the sprue under these conditions has been noted by the authors in another paper.<sup>2</sup>

**Sprue Length Effect Upon Pouring Rate**—Figures 5, 6, 7, 8 and 12 show (1) that the length of untapered round sprues does not affect their pouring rates; (2) the length of sprues tapering from a larger circular cross section at the top to a smaller circular cross section at the bottom has pronounced effect on pouring rate; and (3) that the length of slot sprues has a small but significant effect on pouring rate. These relation-

ships are understood by assuming the following premises: (1) the resistance to metal flow of an untapered round sprue is concentrated at the upper mouth; (2) the resistance to flow offered by a tapered round sprue is not concentrated at the upper mouth, but may be concentrated at the base if sufficient taper is present; and (3) the sidewall resistance to flow of a slot sprue is significant with respect to that offered at the sprue mouth. These same assumptions about sprues of various design were successful in explaining "aspiration" effects in other studies.<sup>2</sup>

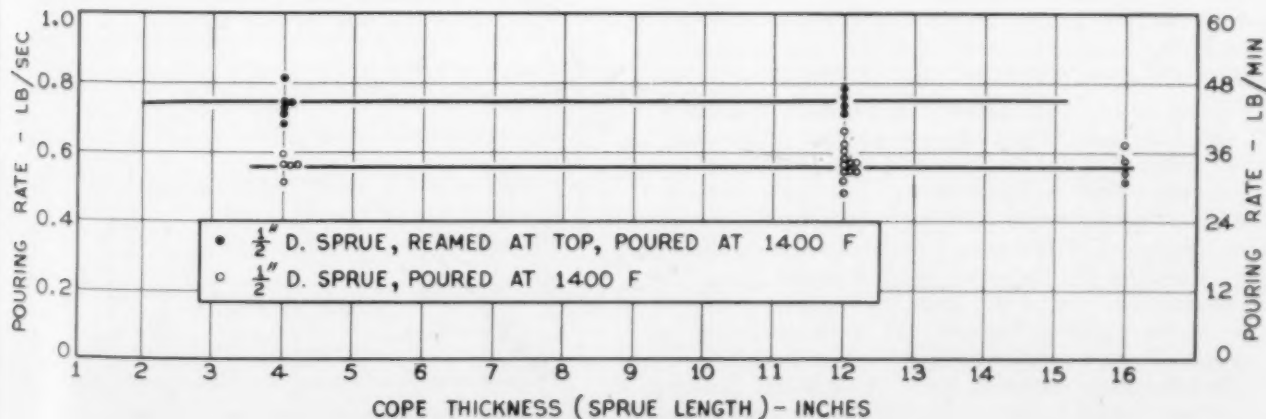
#### Resistance to Flow

With untapered sprues, the only significant resistance to flow appears to be at the upper mouth of the sprue. Resistance to flow offered by the sidewalls seems to be of a lesser magnitude not significant with respect to that of the upper mouth. Hence, the total length of the sprue has no bearing on the rate at which an untapered round sprue will deliver metal.

When the ratio of cross-sectional perimeter to cross-sectional area is high, as with a slot sprue, the sidewall resistance to flow becomes significant with respect to that of the sprue mouth. Hence, the length of a slot sprue, as well as its cross section, influences its pouring rate.

With tapered round sprues, the total resistance to flow will be the sum of component resistances offered by the upper mouth, the sidewalls and the opening at

Fig. 7—Graph showing the effect of the sprue length on the pouring rate of small sprues.



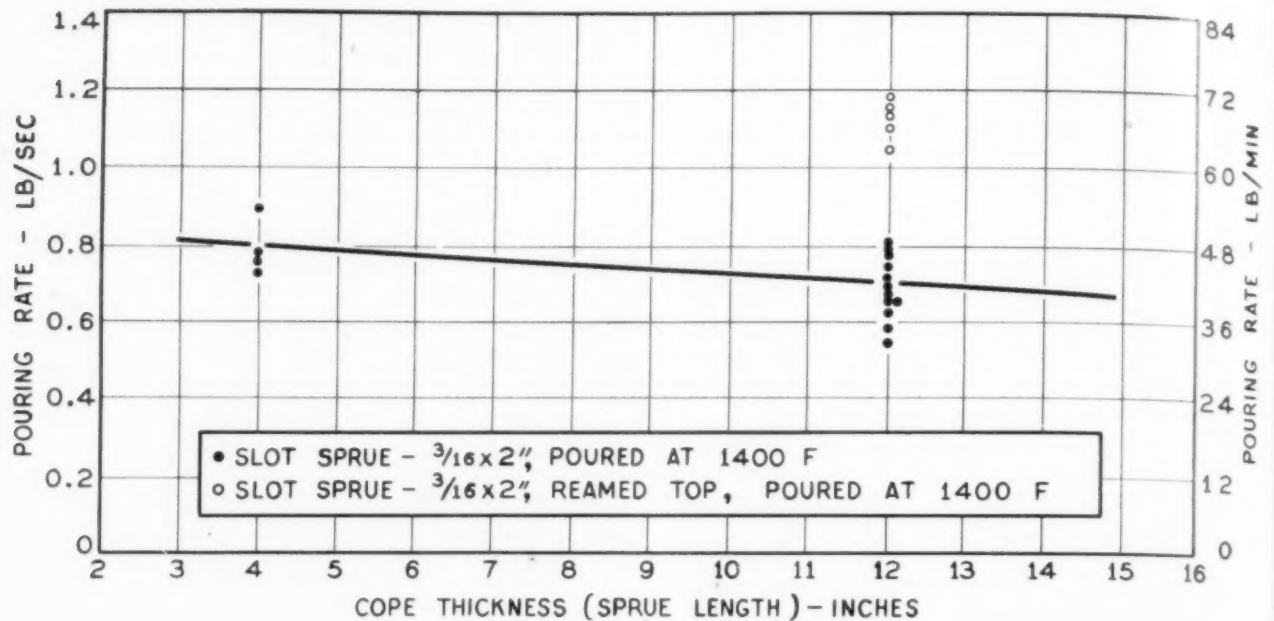


Fig. 8—Plotted test points show effect of sprue height on the pouring rate of a slotted sprue.

the bottom. If only a slight taper is present, the upper mouth may offer the only significant resistance. If taper is extreme, the size of the opening at the bottom end of the sprue may be the only significant factor controlling the rate of pour. At intermediate degrees of taper, all three components may be significant. In any case in which other components than the upper sprue mouth resistance are significant, the sprue length will affect the pouring rate.

**Influence of Metal Depth in Pouring Basin on Pouring Rate**—Figure 13 shows the effect of the depth of metal in the pouring basin on rate of pouring. It is seen that for the round, untapered sprues tested the depth of metal in the pouring basin profoundly affects the pouring rate. This is easily understood when it is considered that for such sprues, the only significant restriction to flow is offered at the sprue mouth; and that, therefore, the height of metal over the sprue mouth in the pouring cup is the driving force controlling the flow rate. Benkoe<sup>1</sup> has assumed the pouring rate of sprues to be a function of the total length of the sprue plus the height of the pouring basin. The data herein presented indicated that only the depth of metal in the pouring cup affects the rate of flow where untapered sprues are involved.

For sprues tapered with the large end up, it is to be expected that the metal depth in the pouring basin would not have so great an effect, since the sprue length would also be a component of the head of metal, provided that there is sufficient taper to the sprue.

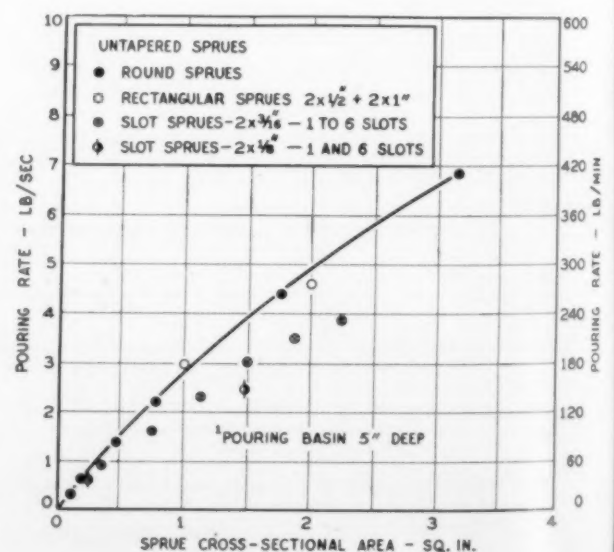
**Sprue Mouth Design and Pouring Rate**—Figures 6, 7 and 8 show the effect on pouring rate of "belling" the mouth of the sprue in such a way as to lower the resistance offered to metal flow at this point. The increase in pouring rate which resulted from reaming out the mouth of the sprue was greater than could be explained by the increased head of metal above the unreamed portion of the sprue. Therefore, a part of the increase in pouring rate had to be attributed to improved entry conditions at the sprue mouth.

Relationships between sprue design and pouring rate established in the foregoing experimental work

provide a basis for varying the pouring rate of castings at will by several alternate methods; and by delineating those factors of sprue design which importantly affect pouring rate provide a basis for preventing large variations in pouring rate from one mold to another from the same pattern.

Considering first untapered sprues of circular cross section, the most obvious factor of design affecting their pouring rates is their cross-sectional area. In arriving at a suitable pouring rate for a given casting, the foundryman varies the diameter of the sprue until the desired pouring rate is attained. The other two factors affecting the pouring rate of the sprue are the depth of metal in the pouring cup and the design of

Fig. 9—Pouring rate at temperature of 1400 F influenced by cross-sectional area of the sprue.



the upper mouth of the sprue. The total length of an untapered round sprue has no effect on the rate at which it will deliver metal to the casting (assuming that the gating is so designed that the pouring rate is controlled at the sprue).

### Pouring Rate Reproducibility

From these considerations, conclusions can be drawn as to which factors of sprue design must be most carefully controlled in order to secure maximum reproducibility of pouring rate from one casting to another. It would seem that three factors would be of first importance: (1) there should be a minimum of "rapping" of the sprue pattern to avoid variations in the effective sprue cross section; (2) the design of the upper mouth of the sprue should be as standard as possible from one mold to another; and (3) the pouring cup should be so chosen that it will be possible to control the depth of metal in it as closely as possible. More important than the absolute amount of variation in depth of metal in the pouring cup from one pour to another is the ratio between this absolute amount of variation and the total depth of the basin. The depth of metal in the basin is the "head" forcing metal to flow through the sprue mouth, and relative deviations in the amount of this head will produce corresponding relative deviations in the pouring rate.

It is to be mentioned that these relationships hold true only in the case of a permeable mold material. A round untapered sprue through an impermeable mold material might retain an unbroken column of metal in it throughout a pour due to the action of atmospheric pressure at the bottom of the sprue. If this happened, the total length of the sprue plus the depth of metal in the pouring basin would be the "head"

Fig. 10—Influence of pouring temperature and number of slots upon pouring rate of 3/16x2 in. slot sprues.

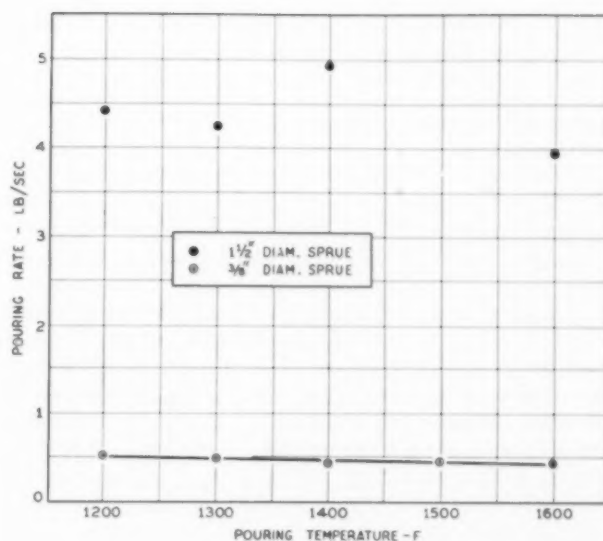
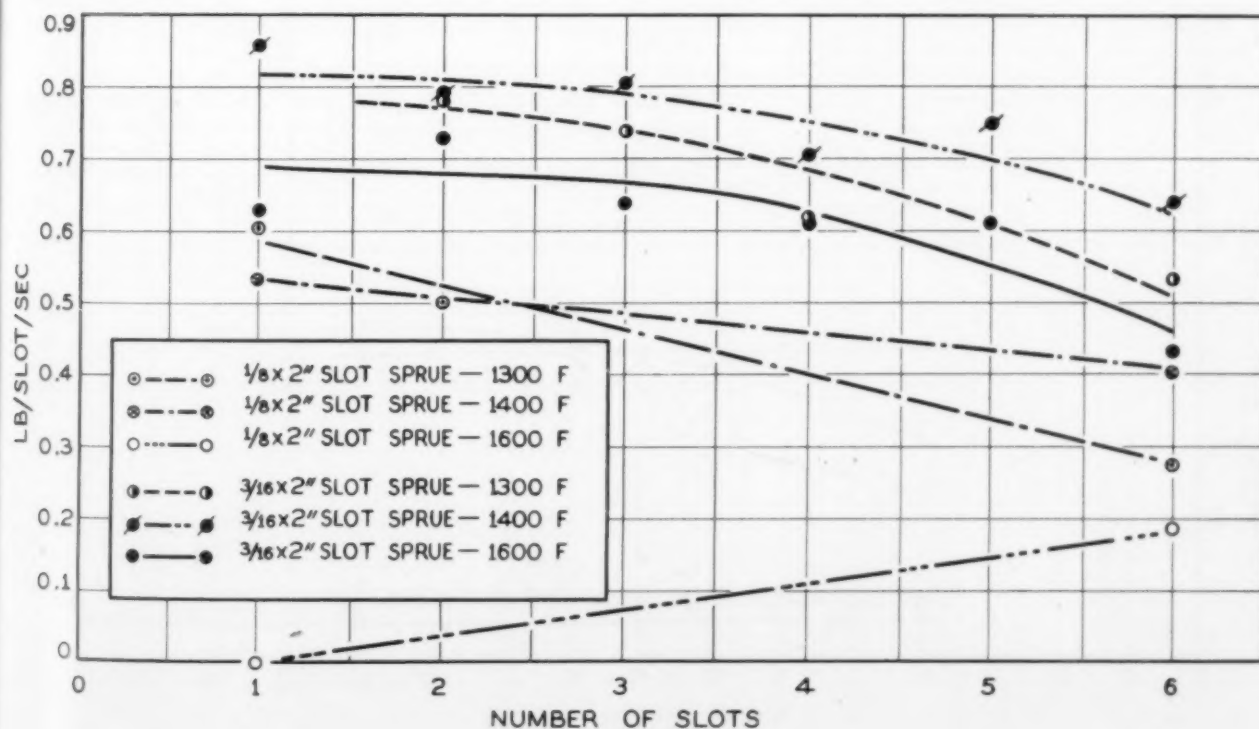


Fig. 11—Graph plotted to show the effect of pouring temperature upon the pouring rate of round sprues.

tending to force the liquid through the sprue system.

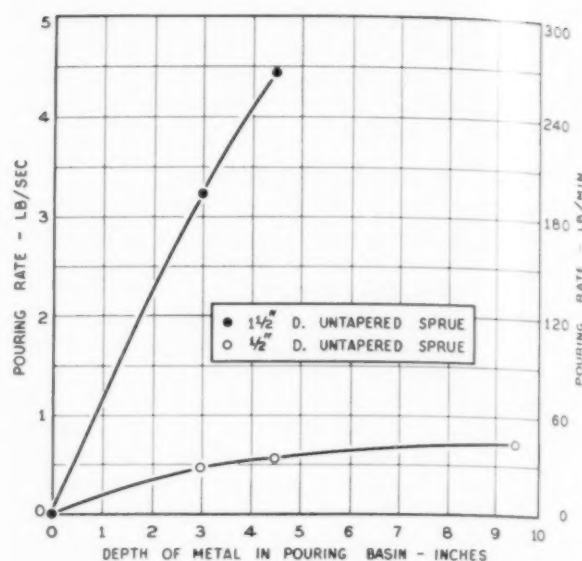
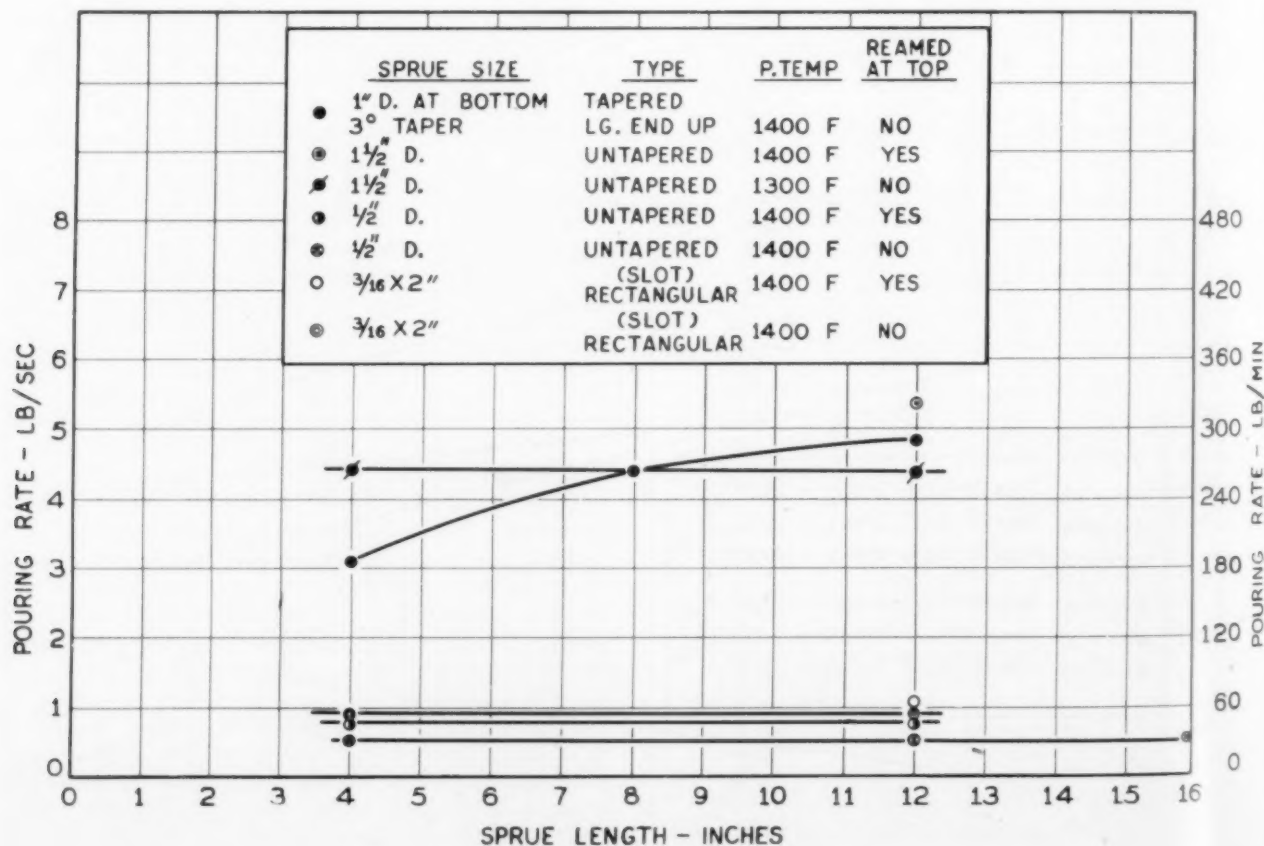
Neither do the same relationships appear to hold true in the case of tapered sprues with the small end down, nor with slot sprues. In both cases, the sprue length may be an important factor affecting the pouring rate of the sprue. Considering first tapered sprues, the degree of taper is of importance. If only a slight taper is imparted to the sprue, the upper mouth will remain the "bottleneck," and the sprue will behave exactly like an untapered one. As the degree of taper is increased, a point will be reached at which the size of the opening at the bottom of the sprue will become so small as to offer a significant resistance compared with that offered by the upper sprue mouth. At this



Fig. 13—Graph showing effect of pouring basin height upon pouring rate for sprue of 12-in. length.

point the sidewall resistance of the sprue may also be significant. As the degree of taper is further increased, the opening at the bottom will become so small that it will be the only significant resistance to flow in the system. At this stage, the "head" of metal forcing metal through this orifice will be the total length of the sprue plus the depth of metal in the pouring cup. Hence, for tapered sprues, the depth of metal in the pouring cup is unlikely to have as great an effect on the pouring rate as it does with untapered sprues. With sprues having a large amount of taper, the design of the upper mouth of the sprue is unlikely to be critical. It should be pointed out that the use of tapered sprues with a degree of taper sufficient to make the bottom opening the only significant source of flow resistance is likely to produce extremely large linear velocity of the metal through this orifice, and that due to this "spurting" such problems as turbulence with light metals or sand-wash with heavier metals would tend to be aggravated. The optimum amount of taper for a round sprue would be that degree of taper at which both the sprue mouth and the bottom opening offer significant resistance to metal flow, thus preventing "aspiration" effects<sup>2</sup> and also excessive velocity at the sprue base. Unfortunately, quantitative work describing this optimum degree of taper for various sprue lengths and pouring rates has not been done at the present time.

Fig. 12—Curves plotted to show effect on pouring rate of lengths of sprues of various sizes and types.



Slot sprues also act somewhat differently from simple untapered round sprues. While the resistance to flow at the upper mouth of slot sprues is significant (see Fig. 8), it is indicated that the sidewall resistance is also a factor affecting the pouring rate. This is concluded from the fact that the sprue length appears to have a small effect on the pouring rate of slot sprues.

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2. H. E. Elliott and J. G. Mezoff, "The Effect of Gating Design on Metal Flow Conditions in the Casting of Magnesium Alloys," *American Foundrymen's Association*, PREPRINT 48-6.

## DISCUSSION

*Chairman:* L. BROWN, Magnesium Fabricators Div., Bohn Aluminum & Brass Corp., Adrian, Mich.

*Co-Chairman:* A. T. RUPPE, Bendix Products Div., Bendix Aviation Corp., South Bend, Ind.

R. G. HARDY:<sup>1</sup> Has any thought been given to Reynolds' number in this study? In the work at the Naval Research Laboratory we found that if the Reynolds' number was below 20,000 we did not have skins, but when the Reynolds' number was higher than 20,000 we did have them.

MR. ELLIOTT: No, we did not go into that phase. It would be interesting to show the relationship between the formation of skins and the Reynolds' number.

MR. HARDY: Since you already have the velocity measured it would not be difficult to do.

MR. ELLIOTT: That is right.

WM. RUBIN:<sup>2</sup> Did you do any work on the effect of the molding conditions on the rate of pour, degree of ramming, moisture content, type of sand used and the sprue friction in pouring?

MR. ELLIOTT: I do not believe we touched on that at all. I cannot recall any tests that we made along those lines.

MR. RUBIN: Were the conditions the same in all the tests, the same moisture, sand and ram conditions?

MR. ELLIOTT: Yes, within fairly reasonable limits they were. We had a set-up in the experimental foundry such that the sand was controlled closely as far as its properties were concerned—moisture, permeability, and such; and the molds were rammed as uniformly as they could be by hand. Most of them were rammed by hand in these tests. Within the limits that we were able to control them, the conditions were the same. In other words, we did not consciously vary our molding conditions in an attempt to find the effect of such variations on pouring rate.

MR. RUBIN: How did you start your pour in order to choke up all the way through your test. I imagine that the control of this factor is the most important phase of the test.

MR. ELLIOTT: That depended upon the skill of our technician whom we trained to pour those molds. We instructed him to do it just as fast as he could—to flood the basin as quickly as he could and to keep it full during the pour. There was a human element. If you got another man who was not trained to pour a mold, then he would not do it as efficiently.

MR. RUBIN: How can an operator hit a 3-in. sprue?

MR. ELLIOTT: There was a pouring basin mounted over all these sprues. In general, that pouring basin was 10 in. long, 5 in. wide, and 5 in. deep.

MR. RUBIN: All he did was to keep the pouring basin full?

MR. ELLIOTT: Yes, the idea was to flood the basin, to get it clear full as soon as possible and to keep it that way. Of

course, that was easier to do with the small diameter sprues than it was with the larger sprues. I brought that point out as one of the reasons there was no linear relationship between sprue cross sectional area and pouring rate. At the faster pouring rates, for one thing, there was more turbulence in the pouring basin, and also it was harder to flood the basin and to keep it full in the case of the larger sprues. Both of these problems would be encountered in production pouring practice. So these effects are significant when the principles relating to the control of pouring rate are applied to production operations.

W. J. FINSETH:<sup>3</sup> Did you use a dry sand pouring basin to make the use of these basins practical in production?

MR. ELLIOTT: In these tests we used a welded boiler-plate construction.

MR. FINSETH: Did it have a green sand lining?

MR. ELLIOTT: No, we used no green sand lining in it. It was a bottomless basin. That is, the top surface of the cope mold acted as the bottom of the pouring basin. The inside of the basin was sprayed with what we call chill wash (a suspension of talc in alcohol with some rosin binder) to keep the metal from sticking.

G. L. LEE:<sup>4</sup> Did you use a high permeability or a low permeability sand?

MR. ELLIOTT: The permeability of the sand used was about 100.

MEMBER: Would it have similar effects on other types of metal besides magnesium?

MR. ELLIOTT: I would not expect a different behavior in other metals. You can reasonably explain most of these effects on a purely fluid flow basis. I do not think that most of the effects would be affected by the properties of the fluid that you happened to be using.

W. E. SICHA:<sup>5</sup> Do you have an explanation for the apparent discrepancy in Fig. 10 in which you plotted the effect on pouring rate of increasing the number of slot sprues. In that set of curves, the bottom one reversed the trend. Instead of having a reduced pouring rate per sprue with an increased number of sprues, the pouring rate increased in that one curve.

MR. ELLIOTT: The curve that Mr. Sicha refers to is the bottom one in Fig. 10 which shows the pouring rates of various numbers of  $\frac{1}{8}$ -in. x 2-in. slot sprues at 1600 F pouring temperature. Our results were very erratic with those  $\frac{1}{8}$ -in. slots at 1600 F. The point corresponding to a single slot actually shows a zero pouring rate, or an infinite pouring time. In other words, when we had only one of these  $\frac{1}{8}$ -in. slots pouring at 1600 F, the casting did not even run. The formation of gases and oxides in such thin slots at high pouring temperatures so erratically affects pouring rates that this curve is not to be relied on. That probably explains the discrepancy.

<sup>1</sup> Light Alloys Limited, Renfrew, Ontario, Canada

<sup>2</sup> International Nickel Co., Bayonne, N. J.

<sup>3</sup> Aluminum Co. of America, Cleveland

<sup>4</sup> Gibson & Kirk, Baltimore

<sup>5</sup> Consultant, Chicago

# A SURVEY OF SIEVE SERIES AND GRADE SCALES

By  
Robert E. Morey \*

## ABSTRACT

*In this paper the author made a survey of existing sieve series and grade scales for determining particle size of particulate materials. He discusses the good and the bad points of the various grade scales. On the basis of this study the author proposes a grading system which combines the best features of existing systems to produce a uniform system which may be acceptable to all who measure and control particle sizes in science and industry.*

PARTICLE SIZE and distribution are of great importance in many branches of science and industry. Particle distribution determines the packing characteristics of such materials as concrete aggregate where dense packing is sought, as well as filter sands where open packing is desired. The surface area per unit weight is important in controlling the rate of chemical reactions and in processes such as the bonding of molding sands where the sand surface must be coated with clay or oil. The study of particle size and distribution by many different industries and sciences, frequently without cognizance of the work done by related industries and sciences has led to many systems of measuring and describing the sizes of particles. The systems or tools usually used for measuring the larger particles are sieves or series of sieves and the arbitrary scales by which they are described are called "Grade Scales." Many of the branches of science and industry which are dependent on the measurement and control of particle size and distribution have originated their own sieve series and grade scales with the result that there are now many grade scales and a wide variety of sieves with different openings and wire diameters. The purpose of the present report is to examine the literature and collect the records of various grade scales and to study their good and bad points.

## General Discussion

Particulate materials are formed by natural or artificial means. Natural means include such processes as

weathering or disintegration of rocks, with subsequent leaching by rainwater, leaving grains of sand and other minerals. These grains may be transported and sorted by both wind and water and during this process receive some grinding. Grinding and polishing also result from glacial action. In any case the material as found in nature is the result of one or more processes which produce an aggregate containing grains with an almost infinite number of different sizes. It has been established that, if such material sorted by natural or artificial processes is sieved through several different sieves and the total amount coarser than each sieve is plotted against the size of the sieve opening, a curve results in which most of the data is crowded toward the fine-size end of the graph. By using a logarithmic instead of arithmetic size scale it is possible to compress the scale at the coarse end of the range and to expand it at the fine end of the range so that the data is well distributed over the graph. The resulting curve frequently resembles an elongated letter S and is known as a "cumulative curve." In order to avoid the necessity for using a logarithmic size scale some sieve series are made with a "built-in" logarithmic relation. That is, the relation between the sizes of openings in the various sieves are in geometric proportion. When a sample is tested in such a series and a cumulative graph is made, it is only necessary to use equal distances between the ordinates for the various sieves. Krumbein<sup>1</sup> states that the increased symmetry of the curve when using a logarithmic instead of an arithmetic size scale is ample justification of the procedure. In almost every case test results are plotted to a logarithmic size scale either by direct use of a logarithmic scale or by plotting successive sieves in a logarithmic series at equal intervals along the size scale. In either case the shape of the resulting curve is the same.

For every cumulative curve<sup>1</sup> there is a corresponding size frequency curve. The frequency curve is the first derivative of the cumulative curve and conversely the cumulative curve is the integral of the frequency curve. At any particular point along the size scale, the area under the frequency curve up to that point is proportional to the height of the ordinate on the cu-

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ulative curve. A cumulative curve may be plotted from an set of sieve data and an accurate size frequency curve may be obtained by graphically differentiating this curve as shown in Krumbein's work. Frequently, however, an attempt is made to plot a size frequency curve directly by using the amounts passing one sieve and retained on the next finer sieve and plotting them as a function of the size of the finer sieve. In this case, the amount of material passing one sieve and retained on another obviously depends on the accuracy of both series and the ratio between their sizes. A flaw in a sieve or any departure from a true logarithmic ratio in the sieves will cause a serious error in the shape of the curve. The sieves in the series must be closely spaced in order to permit a smooth curve to be drawn and they must extend in both directions so that the total sample is included in order to avoid inaccuracies at the ends of the curve. With so many sources of error to be avoided, it is not surprising therefore that frequency curves produce unsatisfactory results.

When a curve showing the size distribution of a sample has been produced, its position and shape give considerable information as to the physical character of the sample just from visual examination. More accurate description or evaluation may be obtained by utilizing certain coefficients or parameters derived from the curves which are described briefly below.

### Sieve Series and Grade Scales

Sieves, made of woven wire, silk, or perforated plates have been in use for a great many years for measuring the size of particulate materials. Nesting sieves are frequently used which are made so that a flange on the bottom of each sieve fits into the bottom of the sieve below. In use, the sieves are stacked in order by size with the finest sieve next to the pan at the bottom. The stack of sieves is then shaken by hand or by mechanical device until each particle reaches a sieve whose openings are too small to permit the particle to pass. Each of the sieves retains particles whose minimum projected area is larger than the square opening in the sieve but is smaller than the opening in the sieve above. The amount retained on each is weighed and plotted to form a distribution curve for the material. Very early it became apparent that a constant ratio was desirable between the sizes of consecutive sieves of a series, and various investigators have proposed sieve series with a geometric ratio between the openings of adjacent sieves. Some of the sieve series which have been proposed are described and discussed below. Some investigators have gone a step further and assigned grade or class names to parts of the range of particle sizes using grade names such as boulders, cobbles, gravel, sand, silt, dust, clay and colloid. These grades have in turn been subdivided into from two to five sub-classes using adjectives such as very coarse, coarse, medium, fine and very fine. These scales are called "grade scales." Such scales are, of course, arbitrary and almost every individual has a different idea as to the size limits of a particle which he would call by a name such as sand or gravel. There is, however, some agreement on the general ranges of

particle sizes covered by these grade names and it may eventually be possible for everyone to agree on a uniform system.

The first geometric sieve series is ascribed to Paul R. von Rittinger.<sup>2</sup> In 1867, he proposed that the openings in any sieve should be twice the area of the openings in the next smaller sieve. In a woven wire sieve with square holes the linear dimension along the side of the hole would then be the square root of two or 1.414 times that of the next finer sieve. This sieve series has become known as the square root of two series or the Rittinger ratio. It is actually a logarithmic series using a base of two instead of ten as in common logarithms. A sieve series based on the square root of two and a starting size of one millimeter is shown as Table 1.

TABLE 1—A SIEVE SERIES BASED ON THE SQUARE ROOT OF TWO RATIO

Linear Dimension of Opening, mm	Ratio of Areas
32.000	1024
22.624	512
16.000	256
11.312	128
8.000	64
5.656	32
4.000	16
2.828	8
2.000	4
1.414	2
1.000	1
0.707	$\frac{1}{2}$
0.500	$\frac{1}{4}$
0.354	$\frac{1}{8}$
0.250	$\frac{1}{16}$
0.177	$\frac{1}{32}$
0.125	$\frac{1}{64}$
0.088	$\frac{1}{128}$
0.062	$\frac{1}{256}$
0.044	$\frac{1}{512}$
0.031	$\frac{1}{1024}$

Orth<sup>3</sup> in 1875, published one of the first grade scales that the writer has seen. He gives limits for the size of the various grades as shown in Table 2. It is inter-

TABLE 2—ORTH'S GRADE SCALE

Size Limits, mm	Designation
Over 3	Gravel
3 to 1	Very coarse sand
1 to 0.5	Coarse sand
0.5 to 0.25	Medium sand
0.25 to 0.05	Fine sand
0.05 to 0.01	Dust
Less than 0.01	Finest dust

esting to note that this grade scale, usually with minor changes, occurs again and again in the literature. It is still extensively used (with some revisions) by the Department of Agriculture and soil scientists. This scale has no definite ratio between grades and the sieves were selected to give a large amount of practical information with a minimum of test work. Data from this series is most easily plotted as a cumulative curve using a logarithmic size scale. This type of grade scale is frequently used for "typing" soils. If, for instance, a particular soil has sand as its most abundant constituent and gravel is next in amount, the soil sample would be classed as a gravelly sand.

In 1898 Johan August Udden<sup>4</sup> published a paper on the mechanical composition of wind deposits. His data were presented in the form of histograms based on a new grade scale. He used a square opening, 1 mm in length along the sides, as a base and made the side of the openings of each successive sieve either twice or half the length of the previous sieve. Udden's scale with the grade names is shown in Table 3. The

TABLE 3—GRADE SCALE PROPOSED IN 1898 BY  
JOHAN AUGUST UDDEN

Size Limits, mm	Designation
8 to 4	Coarse gravel
4 to 2	Gravel
2 to 1	Fine Gravel
1 to 1/2	Coarse sand
1/2 to 1/4	Medium sand
1/4 to 1/8	Fine sand
1/8 to 1/16	Very fine sand
1/16 to 1/32	Coarse dust
1/32 to 1/64	Medium dust
1/64 to 1/128	Fine dust
1/128 to 1/256	Very fine dust

ratio between areas of openings is 4 and between linear dimensions is  $\sqrt[3]{4}$  or 2. This scale is widely used in this country by geologists.

A comparison of Tables 1 and 3 shows that alternate sieves from the hypothetical series shown in Table 1 will furnish the sieves necessary for Udden's Grade Scale. This is a true ratio or logarithmic scale and may be plotted with actual opening on a log scale of size or with the amounts retained on consecutive sieves on consecutive equally spaced ordinates. Udden was among the first to try to combine a grade scale with a logarithmic set of sieves. Sand, for instance, was subdivided into four classes called coarse, medium, fine and very fine sand and each fraction covered an equal distance along the log scale. In the "dust" grade the same adjectives were repeated to extend the scale toward the fine end. In the 1898 version only three classes were given for gravel but this was changed in 1914 when a revised edition was published.

In 1899 C. G. Hopkins<sup>5</sup> in an effort to get all users of sieves to agree on a uniform scale, devised a grade scale which is shown as Table 4. His scale is based on

TABLE 4—GRADE SCALES PROPOSED BY C. G. HOPKINS  
IN 1899

Based on $\sqrt[3]{10}$ (3.16) and $\sqrt[4]{10}$ (1.777)		
$\sqrt[3]{10}$ Series	$\sqrt[4]{10}$ Series	Designation
Over 1 mm	Over 1 mm	Gravel
1 mm to 0.32 mm	1 mm to 0.56 mm	Coarse sand
	0.56 mm to 0.32 mm	
0.32 mm to 0.10 mm	0.32 mm to 0.18 mm	Medium sand
	0.18 mm to 0.10 mm	
0.10 mm to 0.032 mm	0.10 mm to 0.056 mm	Fine sand
	0.056 mm to 0.032 mm	
0.032 mm to 0.010 mm	0.032 mm to 0.018 mm	Coarse silt
	0.018 mm to 0.010 mm	
0.010 mm to 0.0032 mm	0.010 mm to 0.0056 mm	Medium silt
	0.0056 mm to 0.0032 mm	
0.0032 mm to 0.001 mm	0.0032 mm to 0.0018 mm	Fine silt
	0.0018 mm to 0.001 mm	
Less than 0.001 mm	Less than 0.001 mm	Clay

a ratio equal to  $\sqrt[3]{10}$  or 3.16. For closer sizing he also suggested that additional points be used at the geo-

metric mean size between sizes in the regular series. This would produce a series with a ratio equal to the  $\sqrt[4]{10}$  or 1.778. Hopkins was one of the first to suggest a size as small as 0.001 mm as the separation between silt and clay. Hopkins scale is arranged so the main grade separations fall on the ends or at mid-points of logarithmic cycles when semi-log chart paper is used.

J. S. Diller<sup>6</sup> in 1902 described a grade scale (Table 5) which is almost identical with that of Orth.

TABLE 5—GRADE SCALE USED BY J. S. DILLER IN 1902

Size Limits, mm	Designation
Over 2	Gravel
2 to 1	Fine gravel
1 to 0.5	Coarse sand
0.5 to 0.25	Medium sand
0.25 to 0.10	Fine sand
0.10 to 0.05	Very fine sand
0.05 to 0.01	Silt
0.01 to 0.005	Finest silt
Less than 0.005	Clay

In 1905 Dr. Albert Atterberg<sup>7</sup> described a grade scale in which the grades were based on a ratio of 10. He selected grade names in common usage and applied them as nearly as possible to the sizes which geologists and others associate with the names. Atterberg's scale is shown as Table 6. The divisions be-

TABLE 6—GRADE SCALE PROPOSED BY DR. ALBERT  
ATTERBERG IN 1905

Grade Limits, mm	Name
2000 to 200	Blocks
200 to 20	Cobbles
20 to 2	Pebbles
2 to 0.2	Coarse sand
0.2 to 0.02	Fine sand
0.02 to 0.002	Silt
Below 0.002	Clay

tween the grades correspond with what he believed to be significant changes in the behavior of the material. For instance, he claimed that grains coarser than 2 mm will not hold water while grains smaller than this size tend to hold water due to capillarity. He therefore set the division between gravel and sand at 2 mm. The 0.002 mm size is about the upper limit of Brownian movement. Therefore particles larger than 0.002 mm were called silt while smaller particles were called clay. He proposed sub-grade limits at the geometric mean between the main grade limits. This scale has for its ratio  $\sqrt{10}$  or 3.16, like that of Hopkins,<sup>4</sup> but it is based on 2 mm instead of 1 as proposed by Hopkins.

Several references<sup>8,9,10,35</sup> describe a grade scale used by the Bureau of Soils but it is almost identical (see Table 7) with the scales proposed by Orth<sup>3</sup> and

TABLE 7—BUREAU OF SOILS GRADE SCALE

Size Limits, mm	Designation
2 to 1	Fine gravel
1 to 0.5	Coarse sand
0.5 to 0.25	Medium sand
0.25 to 0.10	Fine sand
0.10 to 0.05	Very fine sand
0.05 to 0.01	Silt
0.01 to 0.005	Fine silt
Less than 0.005	Clay

Diller.<sup>9</sup> One credits Whitney with developing the scale. This probably refers to Milton Whitney who was Chief of the Bureau of Soils in 1904.

The Institute of Mining and Metallurgy<sup>11</sup> adopted a sieve series in 1907 which is shown in Table 8. It is

TABLE 8—SIEVE SERIES ADOPTED BY THE INSTITUTE OF MINING AND METALLURGY IN 1907

Mesh	Opening, mm	Mesh	Opening, mm
5	2.540	50	0.254
8	1.574	60	0.211
10	1.270	70	0.180
12	1.056	80	0.157
16	0.792	90	0.139
20	0.635	100	0.127
25	0.508	120	0.107
30	0.421	150	0.084
35	0.416	200	0.063
40	0.317		

unusual in that there is no ratio or relation of any kind between successive sieves. It appears to have been based on a random collection of sieves which were being manufactured at that time. The sieves had very heavy wire so that the open area was only about 25 per cent of the area of the sieve. This scale is very difficult to plot because the sieves have odd sizes both in inches and millimeters and an arithmetic scale cannot be used because the sieves do not have a fixed ratio between the sizes of consecutive sieves.

Keilhach<sup>12</sup> in 1908 described a grade scale (Table 9) almost identical with that of Orth in 1875.

TABLE 9—KEILHACH'S GRADE SCALE

Size Limits, mm	Designation
Over 2	Gravel
2 to 1	Very coarse sand
1 to 0.5	Coarse sand
0.5 to 0.2	Medium sand
0.2 to 0.1	Fine sand
0.1 to 0.05	Superfine sand
0.05 to 0.01	Dust
Less than 0.01	Finest dust

The Tyler Standard Screen scale was brought out about 1910 and was described by G. A. Disbro<sup>13</sup> in 1913. This article shows the intelligent use of cumulative curves over 35 years ago. The Tyler scale is shown

TABLE 10—TYLER STANDARD SCREEN SCALE BASED ON 200 MESH WIRE CLOTH MADE FROM 0.0021 IN.

WIRE AND A $\sqrt{2}$ RATIO	
Mesh	Opening, mm
	26.67
	18.85
	13.33
	9.423
	6.680
	4.699
	3.327
	2.362
3	1.651
4	1.168
6	0.833
8	0.589
10	0.417
14	0.295
20	0.208
28	0.147
35	0.104
48	0.074
65	
100	
150	
200	

in Table 10. It is a logarithmic series using the square root of two ratio and the starting point selected was 200 mesh wire cloth woven from 0.0021 in. wire and having openings of 0.0029 in. or 74 microns.

Professor Robert H. Richards<sup>14</sup> and T. J. Hoover<sup>15</sup> were mentioned by Disbro as the originator of the  $\sqrt[4]{2}$  series (Table 11) and the  $\sqrt[3]{3}$  series (Table 12)

TABLE 11—SIEVE SERIES BASED ON THE

$\sqrt[4]{2}$  (1.189) RATIO

Suggested by Prof. Robert H. Richards

5660 microns	500 microns
4760 "	420 "
4000 "	350 "
3360 "	297 "
2830 "	250 "
2380 "	210 "
2000 "	177 "
1680 "	149 "
1410 "	125 "
1190 "	105 "
1000 "	88 "
840 "	74 "
710 "	62 "
590 "	53 "
	44 "
	37 "

TABLE 12—SIEVE SERIES BASED ON THE

$\sqrt[3]{3}$  (1.442) RATIO

Proposed by T. J. Hoover

27.00 mm	1.000 mm
18.80 "	0.693 "
13.02 "	0.480 "
9.00 "	0.333 "
6.26 "	0.231 "
4.34 "	0.160 "
3.00 "	0.111 "
2.08 "	0.077 "
1.442 "	0.053 "

respectively. This work was apparently done some time previous to 1913 but the original references have not been located. Tables 11 and 12 show sieve series based on these ratios. Comparing Tables 10 and 11 shows that the Tyler sieves agree very closely with alternate sieves from Table 11.

TABLE 13—GRABAU'S GRADE SCALE

Size Limits, mm	Designation
Over 150	Boulders
150 to 50	Cobbles
50 to 25	Very coarse gravel
25 to 5	Coarse Gravel
5 to 2.5	Fine gravel
2.5 to 1.0	Very coarse sand
1.0 to 0.5	Coarse sand
0.5 to 0.25	Medium sand
0.25 to 0.10	Fine sand
0.10 to 0.05	Superfine sand
0.05 to 0.01	Rock flour
0.01 to 0.005	Superfine rock flour
Less than 0.005	Clay size

Grabau<sup>16</sup> gives the grade scale shown in Table 13. It is similar to Orth's scale in 1875 except that it has been extended at the coarse end to include gravel in three subgrades as well as grade limits for cobbles and boulders. This scale has no definite ratio but is rela-



tively simple to plot as a cumulative curve with a logarithmic size scale because the grade limits are round numbers.

In 1914 Udden<sup>17</sup> published a paper in which he amplified his original scale by extending it in both the coarse and fine directions as shown in Table 14.

TABLE 14—UDDENS GRADE SCALE AS MODIFIED IN 1914

Size Limits, mm	Designation
256 to 128	Large boulders
128 to 64	Medium boulders
64 to 32	Small boulders
32 to 16	Very small boulders
16 to 8	Very coarse gravel
8 to 4	Coarse gravel
4 to 2	Gravel
2 to 1	Fine gravel
1 to 1/2	Coarse sand
1/2 to 1/4	Medium sand
1/4 to 1/8	Fine sand
1/8 to 1/16	Very fine sand
1/16 to 1/32	Coarse silt or dust
1/32 to 1/64	Medium silt or dust
1/64 to 1/128	Fine silt or dust
1/128 to 1/256	Very fine silt or dust
1/256 to 1/512	Coarse clay
1/512 to 1/1024	Medium clay
1/1024 to 1/2048	Fine clay

TABLE 15—U. S. SIEVE SERIES PROPOSED BY THE BUREAU OF STANDARDS IN 1919

This scale was adopted by ASTM, then dropped, revised and readopted in 1938. The revised scale is given below.

COARSE SERIES			
Sieve Designation, in.		Sieve Opening, mm	
4.21		107.6	
3 1/2		88.9	
3		76.2	
2 1/2		63.5	
2.12		53.8	
1 3/4		44.4	
1 1/2		38.1	
1 1/4		31.7	
1.06		26.9	
7/8		22.2	
3/4		19.1	
5/8		15.9	
0.530		13.4	
7/16		11.1	
3/8		9.52	
5/16		7.93	
0.265		6.73	
FINE SERIES			
Sieve Designation	Sieve Opening, microns	Sieve Designation	Sieve Opening, microns
No. 3 1/2	5660	No. 40	420
No. 4	4760	No. 45	350
No. 5	4000	No. 50	297
No. 6	3360	No. 60	250
No. 7	2830	No. 70	210
No. 8	2380	No. 80	177
No. 10	2000	No. 100	149
No. 12	1680	No. 120	125
No. 14	1410	No. 140	105
No. 16	1190	No. 170	88
No. 18	1000	No. 200	74
No. 20	840	No. 230	62
No. 25	710	No. 270	53
No. 30	590	No. 325	44
No. 35	500	No. 400	37

The added grades and subgrades maintain the excellent order introduced in his original work. The word silt was added to be used interchangeably with dust.

The "U. S. Sieve Series" was proposed in 1919 by the Bureau of Standards.<sup>18,19</sup> This series has been revised and is now a standard method of the American Society for Testing Materials. The scale, in its final form is shown as Table 15. All of the sizes shown are made in nesting sieves and provide the closest sizes which are available. They are suitable for very accurate work or for materials which are concentrated over a short range of particle sizes. For materials which are spread over a wide range of particle sizes or where high accuracy is not essential, it is common practice to use only every other sieve or every fourth sieve. This series is a logarithmic series based on the fourth root of two and progresses in both directions from a starting size of 1 millimeter.

Since the original Tyler Screen Scale (Table 10) appeared, intermediate sieves have been added between each of those in the original series to make a series for closer sizing. This makes it a fourth root of two series which has openings almost exactly the same as the U. S. Sieve Series. If one compares sieves from the two series having the same or very similar sizes of openings he will find in many cases that coarser wire was used in the Tyler Screen Scale than was used in the U. S. Sieve Series. The number of mesh per linear inch must therefore be less in the Tyler sieves. For this reason, Tyler sieves are usually marked with the mesh number of the Tyler Screen Scale as well as the "U. S. Equivalent Sieve Number" which is the mesh number of the U. S. sieve having the same size of opening. This condition is very confusing but could be remedied by discarding mesh numbers and referring to sieves by the size of the opening, preferably in microns, rather than to the mesh number.

The American Foundrymen's Society<sup>20</sup> has a sieve series shown in Table 16, which is based on the original Tyler Screen Scale.

TABLE 16—AMERICAN FOUNDRYMEN'S SOCIETY SIEVE SERIES

Sieve No.	Size in mm
6	3.327
10	1.651
20	0.833
28	0.590
35	0.416
48	0.294
65	0.208
100	0.147
150	0.105
200	0.074
270	0.053

This is a square root of two sieve series except that two sieves have been omitted near the coarse end. The missing sieves are the No. 8 (2.362 mm) and the No. 14 (1.168 mm). This Society also permits use of sieves from the U. S. Sieve Series which have openings approximately equivalent to those specified in the Tyler Screen Scale. Data cannot be plotted accurately on arithmetic coordinates because of the missing

screens unless all of the material is finer than the No. 14 sieve. It is better to use cumulative curves and semi-logarithmic paper however, because this is applicable to sands of all sizes.

Although the American Foundrymen's Society has never recognized any grade scale as such, it includes in its sand testing handbook<sup>20</sup> a series of definitions for gravel, sand, silt and clay from which an implied grade scale may be derived. The following definitions are taken from the 1938 edition of the *FOUNDRY SAND TESTING HANDBOOK*.

**"Gravel.** When used in connection with foundry practice the granular material resulting from the natural or artificial disintegration of rock and characterized by an appreciable proportion of particles that would pass a sieve with a one-half inch opening and remain on a No. 6 sieve."

**"Sand.** When used in connection with foundry practice the granular material resulting from the natural or artificial disintegration of a rock, a predominant proportion of which passes the No. 6 sieve and is retained on the No. 270 sieve. A sand may contain considerable quantities of larger or smaller particles, including clay."

**"Silt.** Very fine sand particles that pass a No. 270 mesh sieve, but which are not plastic or sticky when wet."

**"Fine silt.** Very fine sand particles less than 20 microns in diameter. This is included in AFS clay and by itself has very little plasticity or stickiness when wet."

**"Fines.** This term is used to designate those particles of the sand which are retained on the 200, 270 sieves and pan, as determined by the AFS fineness test." (Since the next coarser sieve in this series is the 140 mesh, "fines" includes all particles finer than 104 microns except where a separate test is made for AFS clay substance. In this case the clay is removed by decantation up to a size of 20 microns. In this case, apparently, fines includes the range of particles between 104 and 20 microns in size.)

**"Clay Substance.** That earthy portion of foundry sand which, when suspended in water, fails to settle one inch per minute, and which consists of particles less than 20 microns in diameter."

**"Colloids.** Colloidal Material. Finely divided material less than 0.5 micron in size, gelatinous, highly adsorbent and very sticky when moistened."

TABLE 17—A GRADE SCALE BASED ON A.F.S. DEFINITIONS

Size Limits, mm	Designation
12.7 to 3.36	Gravel
3.36 to 0.053	Sand
0.053 to 0	Silt
0.020 to 0	Fine silt
0.104 to 0	Fines
0.020 to 0	Clay
0.0005 to 0	Colloids

From these definitions the grade scale shown in Table 17 results. There is considerable overlapping of the grades particularly in the fine sizes. Some particle sizes for instance might be in any one of five grades.

It appears that a more exact definition of the size of the various grades would simplify this situation. The selection of 20 microns as the coarse end of the clay size range is much larger than in any of the other grade scales found.

A grade scale given by H. A. Baker<sup>21</sup> in 1920 is shown as Table 18. It is very similar to Orth's scale.

TABLE 18—GRADE SCALE USED BY BAKER IN 1920

Size Limits, mm	Designation
Over 2	Gravel
2 to 1	Very coarse sand
1 to 0.5	Coarse sand
0.5 to 0.25	Medium sand
0.25 to 0.10	Fine sand
0.10 to 0.05	Coarse silt
0.05 to 0.01	Silt
Less than 0.01	Clay

Having no ratio it should be plotted with cumulative per cent retained versus size on a log-scale.

J. Thoulet<sup>22</sup> described a grade scale (Table 19) in 1922 which seems to be original but does not follow any orderly system or ratio.

TABLE 19—GRADE SCALE USED BY THOULET ABOUT 1922

Size Limits, mm	Designation
Over 3	Fine gravel
3 to 0.89	Coarse sand
0.89 to 0.45	Medium sand
0.45 to 0.26	Fine sand
0.26 to 0.04	Very fine sand
Less than 0.04	Fine fine sand

In 1922 Wentworth<sup>23</sup> gathered a number of opinions from geologists and other workers in soils and sediments and proposed a revised scale which is known as Wentworth-Udden scale, or sometimes just as the Wentworth scale. It is shown in Table 20.

TABLE 20—WENTWORTH'S REVISION OF THE UDDEN GRADE SCALE

Size Limits, mm	Designation
Over 256	Boulders
256 to 64	Cobbles
64 to 4	Pebbles
4 to 2	Granules
2 to 1	Very coarse sand
1 to 1/2	Coarse sand
1/2 to 1/4	Medium sand
1/4 to 1/8	Fine sand
1/8 to 1/16	Very fine sand
1/16 to 1/256	Silt
Less than 1/256	Clay

Comparison of Table 20 with Tables 3 and 14 show that Wentworth's scale is a revision of the Udden scale but the orderly basis has been removed. He changed the four subgrades of sand to five, did not subdivide silt or clay at all. The gravel and boulder grades were completely rearranged.

G. W. Robinson<sup>24</sup> in 1924 and W. W. Rubey<sup>25</sup> in 1930 discussed settling velocities and preferred this term for the expression of particle size instead of diameter or "effective diameter." Rubey's scale is shown

TABLE 21—RUBEY'S GRADE SCALE BASED ON  
SETTLING VELOCITIES

Velocity Limits	Approximate Size Limits*	Designation
Greater than 3840 u/sec.	Over 200 u	Very fine sand
3840 u/sec. to 960 u/sec.	200 u to 100 u	Coarse silt
960 u/sec. to 240 u/sec.	100 u to 50 u	Medium silt
240 u/sec. to 60 u/sec.	50 u to 25 u	Fine silt
60 u/sec. to 15 u/sec.	25 u to 12 u	Very fine silt
15 u/sec. to 3.75 u/sec.	12 u to 6 u	Coarse clay
3.75 u/sec. to .9375 u/sec.	6 u to 3 u	Medium clay
Less than .9375 u/sec.	Less than 3 u	Fine clay

\*Assuming a specific gravity of 2.65, temperature of 67 F and approximately spherical shape.

as Table 21. By using settling velocities it is unnecessary to consider particle shape or density since no attempt is made to describe the particles by size.

Parker D. Trask<sup>26</sup> in 1932 used a scale of size limits for gravel, sand, silt, clay and colloidal clay which are shown as Table 22. In his work on oil-bearing sands,

TABLE 22—GRADE SCALE USED BY TRASK

Size Limits, mm	Designation
Over 1	Gravel
1 to 0.05	Sand
0.05 to 0.005	Silt
Less than 0.005	Clay
Less than 0.001	Colloid

Trask utilized statistical methods for the study of grain size, grain distribution or sorting, and skewness of distribution. His extremely simple methods made it possible to analyze and compare large numbers of sediments. Trask plotted his sieve analyses as cumulative curves using semi-logarithmic graph paper and from this determined the per cent of the sample in each grade of his scale. He then determined the "median size," "sorting coefficient" and "skewness coefficient" from the graph. The median size was determined by finding the point at which the cumulative curve crossed the 50 per cent retained line on the graph. The median size is usually very close to the statistical arithmetic mean size and it is much easier to determine and therefore serves as a very quick and reasonably accurate method of expressing the particle size of the sample. The sizes at which the cumulative curve crossed the 25 per cent and 75 per cent retained lines on the graph were then read from the size scale of the graph. These points are called the "quartiles." Dividing the larger quartile by the smaller and taking the square root of the quotient gives a number called the sorting coefficient. This is a number which expresses with a minimum of calculation, the distribution or spread of the sample. The sorting coefficient for uniform particles like a single size of ball bearings is 1.0 and the coefficient increases with increasing deviation from uniformity of size. The skewness coefficient was determined by dividing the product of the two quartiles by the square of the median size. This coefficient is useful for detecting abnormal samples. For normal samples the skewness is usually about 1.0 but may be more or less depending on the amount and direction in which the sample is skewed.

In 1932 the British Engineering Standards Association<sup>27</sup> adopted a screen scale which supersedes the Institute of Mining and Metallurgy Scale of 1907. The new scale is shown as Table 23. This series is almost identical with the U. S. Sieve Series and the

TABLE 23—SIEVE SERIES ADOPTED IN 1932 BY THE  
BRITISH ENGINEERING STANDARDS ASSOCIATION

Sieve Designation	Sieve Opening mm	Sieve Designation	Sieve Opening mm
5	3.360	36	0.422
6	2.810	44	0.353
7	2.410	52	0.294
8	2.056	60	0.251
10	1.676	72	0.211
12	1.403	85	0.178
14	1.202	100	0.152
16	1.002	120	0.124
18	0.854	150	0.104
22	0.698	170	0.089
25	0.599	200	0.076
30	0.506	240	0.067

Tyler Fine Series. The dimensions vary slightly, apparently to permit the use of wire gages which are used in England. This adds still another series of mesh numbers to a sieve series which is approximately the same as in the two series just mentioned.

In 1933 Wentworth<sup>28</sup> published an article in which he claimed that a relation exists between the method of transport of particles in water and the main grades as given by his scale. That is, gravel is the material carried or dragged by "traction," sand is the portion carried by "inertia suspension," silt is the part carried by "viscous suspension," and clay is material carried by "colloidal suspension."

#### Frequency Distribution Curves

In the study of statistics a large amount of information and theory has been developed in connection with the study of frequency distribution curves. The size frequency curve for sand and other particulate material has very nearly the same shape as a normal distribution curve if the individual percentages between any two sieves of a logarithmic sieve series are plotted as a function of size on a log scale. It is difficult to use statistical formulae, however, when size is plotted on a logarithmic scale. Krumbein<sup>29</sup> proposed a "transformation equation" for converting the Wentworth-Udden scale into an arithmetic scale which he called a "Phi scale." Any sieve, regardless of its opening size could be assigned a "Phi number" which represented its size. When the results of sieve tests were plotted they then had an arithmetic size scale but the curve has the same shape as if it had been plotted on a logarithmic size scale. By means of this technique the statistical mean size, standard deviation, skewness and kurtosis of any particulate distribution may be found. Krumbein's equation is

$$\phi = -\log_2 \xi$$

where  $\xi$  is the diameter in millimeters and  $\phi$  is the logarithmic expression for size. The phi values of the Wentworth grades are shown in Table 24.



TABLE 24—KRUMBEIN'S PHI SCALE

Wentworth Grades, mm	$\phi$ Value
32	-5
16	-4
8	-3
4	-2
2	-1
1	0
1/2	+1
1/4	+2
1/8	+3
1/16	+4
1/32	+5
1/64	+6
1/128	+7
1/256	+8
1/512	+9
1/1024	+10

In 1934 the International Society of Soil Science<sup>30</sup> adopted a grade scale based on Atterberg's classification which is widely used in Europe and is shown as Table 25. It has the advantage of having a logarithmic

TABLE 25—INTERNATIONAL GRADE SCALE ADOPTED IN 1934 BASED ON ATTERBERG'S SCALE

Size Limits, mm	Designation
Over 2	Gravel
2 to 0.2	Coarse sand
0.2 to 0.02	Fine sand
0.02 to 0.002	Silt
Less than 0.002	Clay

basis and each cycle corresponds to one grade. It is suitable only for rough grading because it lacks subdivisions into classes.

In 1937 Krumbein<sup>31</sup> proposed the equation  $Z = 0.301 - \log_{10} d$  for use with the Atterberg or International Scales, where  $d$  is the diameter in millimeters. This produces the Zeta scale which is shown in Table 26. The purpose of this method is to provide an arith-

TABLE 26—KRUMBEIN'S ZETA SCALE

Atterberg Grades, mm	Zeta Value
2000	-3
200	-2
20	-1
2	0
0.2	+1
0.02	+2
0.002	+3
0.0002	+4

metic size scale so that statistical methods may be employed as in the case of the Phi scale. In both the Phi and Zeta scales the numbers increase as the particle size decreases. As Krumbein pointed out, a graph may be constructed by plotting Tables 24 or 26 to suitable scales to convert actual sizes to either scale, but to find an exact value the formula must be used. This is somewhat tedious especially for the Phi scale where logarithms to the base of 2 are used. Logarithms to the base 2 are not commonly found in tables and must be obtained from tables of logarithms to some other base by means of equations for transforming logarithms from one base to another. The con-

stant 0.301 in the Zeta scale is to transfer the grade limit from 1 to 2 on the log cycle. If the grades were from 1 to 10, 10 to 100, etc., the constant could be eliminated.

C. A. Hogentogler<sup>32</sup> gives a classification used by soil scientists which is shown as Table 27. There are

TABLE 27—HOGENTOGLER'S GRADE SCALE

Size Limits, mm	Designation
Over 2	Gravel
2 to 0.42	Coarse sand
0.42 to 0.05	Fine sand
0.05 to 0.005	Silt
Less than 0.005	Clay
Less than 0.001	Colloids

no definite relations between the grades and they were selected to permit the rapid "typing" of soils with only 10, 40 and 270 mesh sieves.

Donald M. Burmister<sup>33,34</sup> in 1941 described a classification system which is shown as Table 28. It is based

TABLE 28—MODIFIED M.I.T. CLASSIFICATION SUGGESTED BY D. M. BURMISTER

Size Limits, mm	Designation
60 to 20	Coarse gravel
20 to 6	Medium gravel
6 to 2	Fine gravel
2 to 0.6	Coarse sand
0.6 to 0.2	Medium sand
0.2 to 0.06	Fine sand
0.06 to 0.02	Coarse silt
0.02 to 0.006	Medium silt
0.006 to 0.002	Fine silt
Less than 6 microns	Clay

on a classification used at the Massachusetts Institute of Technology which in turn uses the scale devised by Atterberg in 1905 with changes in the grade names. This scale has the advantage over the International scale in that the grades are subdivided into classes. Each grade, however, covers  $1\frac{1}{2}$  log cycles so the scale no longer has the agreement between grades and log cycles that is present in the International Scale.

### Summary of Existing Scales

This survey of sieve series and grade scales is by no means complete but it serves to illustrate the principles involved and methods which have been developed for particle size measurement.

In general the scales are based on the metric system, a large proportion of them using 1 millimeter as the starting point.

A random collection of sieves will do an excellent job of measuring the size of any sample provided they are of such a size that they split the sample in several places. If the total amount coarser than each sieve is plotted as a function of the size of the opening along a log scale a perfectly sound cumulative curve will be produced, regardless of sieve size or sieve spacing. This curve can be differentiated graphically<sup>29</sup> to produce the corresponding size frequency curve.

By introducing logarithmic sieve series which include all sieve series with geometric ratios of areas or width of squares, square root, cube root, fourth root,

and other exponential series, it became possible to plot the size frequency curve directly without first making a cumulative curve so long as the individual sieve percentages were plotted at equal steps along the size axis. Care must be used, however, to be sure that the ratio between the size of adjacent sieves is preserved because if extra sieves are inserted in the series or if any are omitted, serious errors in the curve will occur.

There have been recurring efforts to combine grade scales with sieve series so that a sample can be graded simply by sieving without first having to plot the data.

A substantial group favor grades which correspond to logarithmic cycles.

Grade scales with a logarithmic basis can be extended at either end as necessary to suit any particular purpose. Sieves are only a tool for measuring size in a particular range. Below about 40 microns it is impractical to use sieves and then other methods of measurement can be used. These include the pipette and hydrometer methods, the microscope, specific gravity and sedimentation balances, and devices such as the Wagner turbidimeter which measures light transmission through a dispersed suspension of the sample. At the large end of the scale particles too coarse for sieve could be measured with a meter stick or a measuring tape. Even the solar system could be "sieved" by using the astronomical telescope as the tool for making the measurements. In fact the system of measuring the brightness of stars by "star magnitudes" resembles the logarithmic grade scales very closely.

When closer sizing or description is required, grades can be split into two or more classes and sieve intervals can be split by the insertion of extra sieves in an existing sieve series. The size of the added sieves should be such that it is in geometric proportion between the two adjacent sieves.

As Krumbein pointed out, any sample may be represented as a cumulative curve and there is a corresponding size frequency curve. The cumulative curve is the integral of the size frequency curve and conversely the size frequency curve is the first derivative of the cumulative curve. At any particular size, the area under the size frequency curve is proportional to the height of the ordinate of the cumulative curve.

#### Proposed Grading System

At the risk of adding "one more grade scale" to the already large group the writer believes that a grade scale and sieve series can be made which will combine the best features of existing systems to produce a uniform system which may be acceptable to all who measure and control particle sizes in science and industry.

It is proposed that the system start at a size of 1 millimeter and progress in both directions as far as necessary with one logarithmic cycle per grade. This establishes a grade scale as shown in Table 29, Columns 1 and 2.

It is proposed that the common logarithm of the size in millimeters be used to form a "Delta" scale. This establishes the transformation equation

$$\Delta = \log_{10} d$$

where  $d$  is particle diameter or sieve opening. The

TABLE 29—A LOGARITHMIC GRADE SCALE FOR THE DESCRIPTION OF PARTICLE SIZES BASED ON THE METRIC SYSTEM

Size limits (d)	Designation	Logarithm of size ( $\Delta$ )
1000 mm to 100 mm	Boulders	+ 3 to + 2
100 mm to 10 mm	Cobbles	+ 2 to + 1
10 mm to 1 mm	Gravel	+ 1 to 0
1000 $\mu$ to 100 $\mu$	Sand	0 to - 1
100 $\mu$ to 10 $\mu$	Silt	- 1 to - 2
10 $\mu$ to 1 $\mu$	Dust	- 2 to - 3
1 $\mu$ to 0.1 $\mu$	Clay	- 3 to - 4
0.1 $\mu$ to 0.01 $\mu$	Colloidal Clay	- 4 to - 5
0.01 $\mu$ to 0.001 $\mu$	Colloids	- 5 to - 6

use of Delta units is shown in Table 29, Column 3, where they are given for the grade limits of the proposed grade scale.

It is proposed that each grade be subdivided into five classes by modifying adjectives. This subdivision for the gravel grade is shown in Table 30.

TABLE 30—METHODS OF SUBDIVIDING GRADES INTO CLASSES

Class Limits	$\Delta$ Limits	Diameter Limits mm
Very Coarse Gravel	1 to 0.8	10 to 6.30
Coarse Gravel	0.8 to 0.6	6.3 to 4.00
Medium Gravel	0.6 to 0.4	4.00 to 2.50
Fine Gravel	0.4 to 0.2	2.50 to 1.60
Very Fine Gravel	0.2 to 0.0	1.60 to 1.00

It is proposed that sieves be made (for the range of particle sizes which are amenable to sieving) dividing each grade into ten fractions whose sizes are to be in geometric proportion. This establishes the sieve series shown in Table 31.

In Column 2 are shown the exact geometric steps in millimeters which are found by taking the antilogarithms of the numbers shown in Column 1. Column 3 shows the same figures "rounded off" to produce a number which can be plotted more easily when semi-logarithmic chart paper is used. The amount of the "rounding" is less than 1 per cent in every case so either column could be used as the basis for a sieve series. Column 4 lists some existing sieves which are very close to or exactly on the nominal sieve opening. From this it can be seen that very few new sieves would have to be made to produce a set of sieves very close to the proposed series.

It is proposed that the sieves be marked with their opening in millimeters for sieves of 1 mm or over and in microns for sieves under 1 mm, that the logarithm of the size be included, and that mesh and wire diameter be omitted or in very small figures.

The obvious advantages of the proposed grade scale and sieve series are their simplicity and adaptability to the scientific measurement and control of particle sizes. An examination of Table 31 shows that it combines the systems which are now used most extensively. It includes all the points used in the Udden and Wentworth systems. It includes all the points used in the Atterberg and International systems. It includes all of the points used in the Hopkins scale except for the smallest subdivisions. It includes all

TABLE 31—PROPOSED SIEVE SERIES

Logarithmic Size	Exact Geometric Size Steps in Mm	Geometric Size Steps in Round Numbers	Existing U. S. Sieve Near Enough to be Acceptable
2.0	100.000	100	101.6 mm (4 in.) U. S. Series
1.9	79.433	80	
1.8	63.096	63	63.5 mm (2½ in.) U. S. Series
1.7	50.120	50	50.8 mm (2 in.) U. S. Series
1.6	39.811	40	
1.5	31.623	31.6	31.7 mm. (1¼ in.) U. S. Series
1.4	25.119	25.0	25.4 mm (1 in.) U. S. Series
1.3	19.953	20.0	19.5 mm (¾ in.) U. S. Series
1.2	15.849	16.0	16 mm (⅝ in.) 0.120 in. bronze wire (Tyler)
1.1	12.589	12.6	12.7 mm (½ in.) U. S. Series
1.0	10.000	10.0	10.03 mm (No. 2) 0.105 in. bronze wire (Tyler)
0.9	7.9433	8.0	7.93 mm (⅝ in.) U. S. Series
0.8	6.3096	6.3	6.35 mm (¼ in.) U. S. Series
0.7	5.0120	5.0	5.00 mm (No. 4½) 0.025 in. steel wire (Tyler)
0.6	3.9811	4.0	4 mm (No. 5) U. S. Series
0.5	3.1623	3.16	3.18 mm (No. 7) 0.018 in. bronze wire (Tyler)
0.4	2.5119	2.50	2.54 mm (No. 8) 0.025 in. bronze wire (Tyler)
0.3	1.9953	2.00	2 mm (No. 10) U. S. Series
0.2	1.5849	1.60	1.60 mm (No. 12) 0.020 in. bronze wire (Tyler)
0.1	1.2589	1.26	1.257 mm (No. 16) 0.013 in. steel wire (Tyler)
0.0	1.0000	1.00	1 mm (No. 18)
-0.1	0.79433	0.80	0.800 mm (No. 22) 0.014 in. bronze wire (Tyler)
-0.2	0.63096	0.63	0.627 mm (No. 24) 0.017 in. bronze wire (Tyler)
-0.3	0.50120	0.50	500 u (No. 35)
-0.4	0.39811	0.40	0.406 mm (No. 40) 0.009 in. bronze wire (Tyler)
-0.5	0.31623	0.316	0.310 mm (No. 45) 0.010 in. bronze wire (Tyler)
-0.6	0.25119	0.250	250 u (No. 60)
-0.7	0.19953	0.200	0.198 mm (No. 70) 0.0065 in. steel wire (Tyler)
-0.8	0.15849	0.160	0.155 mm (No. 90) 0.0050 in. steel wire (Tyler)
-0.9	0.12589	0.126	125 u (No. 120)
-1.0	0.10000	0.100	
-1.1	0.079433	0.080	0.0787 mm (No. 190) 0.0022 in. bronze wire (Tyler)
-1.2	0.063096	0.063	62 u (No. 230)
-1.3	0.050120	0.050	0.0508 mm (No. 280) 0.0016 in. bronze wire (Tyler)
-1.4	0.039811	0.040	

the points in the scales of J. S. Diller, K. Keilhach, H. A. Baker, P. D. Trask and the Bureau of Soils. It includes all points but one in both Orth's and Hogen-togler's scales.

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### DISCUSSION

*Chairman:* J. B. CAINE, Sawbrook Steel Castings Co., Lockland, Cincinnati.

*Co-Chairman:* WERNER FINSTER, American Chain & Cable Co., Reading, Pa.

A. I. KRYNITSKY:<sup>1</sup> When we are talking about fine silt, or some kind of silt, I think it is simply confusing.

MR. MOREY: Silt is something we do not know very much about. As we study those finer particle sizes, the surface area changes very rapidly with change in size. Surface area is something that is important to us when we have to coat that surface with clay, cereal binders and core oils.

MR. KRYNITSKY: We call silt everything that passes through the number 270 sieve; that is below 50 microns.

MR. MOREY: You can subdivide it into as many grades as you want then depending upon the use you have for it. If you have no use for it or if there is none of it in your sand, there is no use spending any time trying to subdivide it. In the fine Albany sands, for instance, most of the sand will fall in that grade and there the division of sizes in that sand is important.

MR. KRYNITSKY: I wish to repeat that all these subdivisions may lead to confusion. It is simple and practical to call "silt" the material passing through the number 270 sieve.

<sup>1</sup> National Bureau of Standards, Washington, D. C.

MR. MOREY: That is the definition of the American Foundrymen's Society.

A. C. ROSENCRAZ:<sup>2</sup> Have you been able to determine at what point the bond clay begins? I have an idea that while AFS clay may begin at 20 microns, I have dirt in my clay which is not a bonding constituent. If with the aid of the hydrometer I could determine at what point the bond begins, I could eliminate in measuring clay that amount of dirt which may be a larger size and yet has no bonding property in it.

MR. MOREY: The hydrometer will only determine particle size. It will not tell whether or not the particles are actually a plastic clay. It is well established, however, that true clay particles are much smaller than 20 microns.

MR. ROSENCRAZ: At what point then as far as micron size is concerned does clay bond appear?

MR. MOREY: I cannot say exactly when it begins, but it is very fine. It is probably less than one micron.

D. C. WILLIAMS:<sup>3</sup> I think it has been adequately determined that the dimensions of the unit clay particle is expressed in Angstrom units. Therefore when we speak of particles of clay measured in microns we are considering an aggregated mass. Further it should be remembered that any of the methods under discussion describe only particle sizes and does not indicate the chemical composition or the bonding qualities of the particles. The mechanical dispersion procedure used will influence the particle size distribution data obtained.

<sup>2</sup> Farm Tools, Inc., Evansville, Ind.

<sup>3</sup> Ohio State University, Columbus, Ohio

# MODERNIZATION OF THE "SMALL" FOUNDRY

By

Lester B. Knight\*

## ABSTRACT

No foundry can afford not to modernize and provide good working conditions under which the producers can produce maximum units with adequate return to labor, management and owners. Every foundry can modernize and reduce its costs and improve its products. Foundry modernization must be "tailored" to fit the particular plant under consideration.

Some of the factors which should be investigated and properly modernized to secure maximum results are: Sales methods and prices, organization and personnel, cost control, patterns, rigging, production equipment and methods, wage incentives, plant layout, material handling, good housekeeping, and mechanized or motorized facilities.

THERE ARE APPROXIMATELY 6,000 foundries in the United States and Canada. Of these, 2,550 have 25 or fewer employees, 1170 have 25 to 50 employees, 1000 have 50 to 100, 850 have 100 to 250 and only 475 have 250 or more employees.<sup>1</sup> Thus, it is apparent that most foundries are small industrial units. It should be noted, however, that the great bulk of the production comes from the latter two classes with approximately 1325 units employing 100 or more people.

The principles of efficient operation apply equally to those employing 25 or fewer or 100 or more employees. The foundry employing less than 50 people usually does not have substantial funds available for capital expenditures. However, it can be operated efficiently as the author will illustrate by actual case histories.

The foundry business, being a process industry with many variables, offers an opportunity to improve costs and product by better supervision. The personnel of the plant are the most important asset of the plant. One of the problems today is lack of properly trained supervision who know and do the job of supervision. During the past five years the foundry industry trained virtually no supervisors with adequate background. Meanwhile, demand for increased production, and the tremendous expansion of facilities has placed a burden on existing supervision. Supervision was spread so thin that virtually all plants can make substantial im-

provement by an intensive supervisor training program. It is increasingly evident that the facilities in a foundry are no better than the men operating them.

The demand for higher quality, closer tolerances, better finish, and lower costs, places on every foundry the need for modernization of methods, rigging, procedures and facilities. "How can we modernize our foundry?" is the question we hear most often. The increased cost of materials and labor, competition from other materials and methods of manufacture, has given great impetus to the desire to modernize. The great demand for castings at prices which insure a profit for the conservative foundry operator has made the funds available to carry out a modernization program.

It is not possible to give one answer for all foundries who are interested in modernization. Each program should be designed to fit the individual plant requirements and financial condition. Overly simplified, perhaps, the foundry problem consists first, of rigging to produce economically, and second, materials handling. There is much that any foundry may do to improve its operations and reduce its costs other than to spend large sums for capital equipment.

Modern selling methods employed to permit "sales engineering" of the proper casting for the particular job insures greater continuity of orders, customer satisfaction, and better profits. The old idea of selling castings for a conversion cost of metal at so much per pound should be replaced with the sale of an engineered casting to do the job at so much per piece.

## Patterns and Rigging

In too many foundries that the author has visited, the design of the casting and construction and design of the pattern, core box, and flask equipment is such that the castings are "made" in the Cleaning Room instead of the foundry. The "finishing" cost is much too high. Quality patterns designed to permit maximum yield and high unit production commensurate with the number of castings to be made from the pattern are a "must" in any modernization program. Consistently high scrap losses usually indicate faulty pattern and rigging of the job.

\* President, Lester B. Knight & Associates, Inc., Chicago.

<sup>1</sup> Penton Publishing Co. Survey, 1947.

Wherever economically practicable, patterns should be mounted on plates or boards, with gates and risers mounted or clearly indicated. In one steel foundry, if eight squeeze molds are to be made from the pattern, it is mounted and made on a machine. For larger cope and drag work in flasks 30 in. x 36 in. and larger, every pattern is mounted and rammed on a jolt or jolt-rollover machine. This not only insures lower costs but closer tolerances, better finish, less scrap and higher quality. Any machine operator may produce high quality, uniform castings, if properly instructed. It is considered that each foundry should analyze its operations and costs, and establish standards for mounting patterns. If the "engineering" is accomplished in the Pattern Shop, the need for rapidly vanishing skilled molders is greatly reduced.

Figure 1 shows examples of patterns for farm imple-

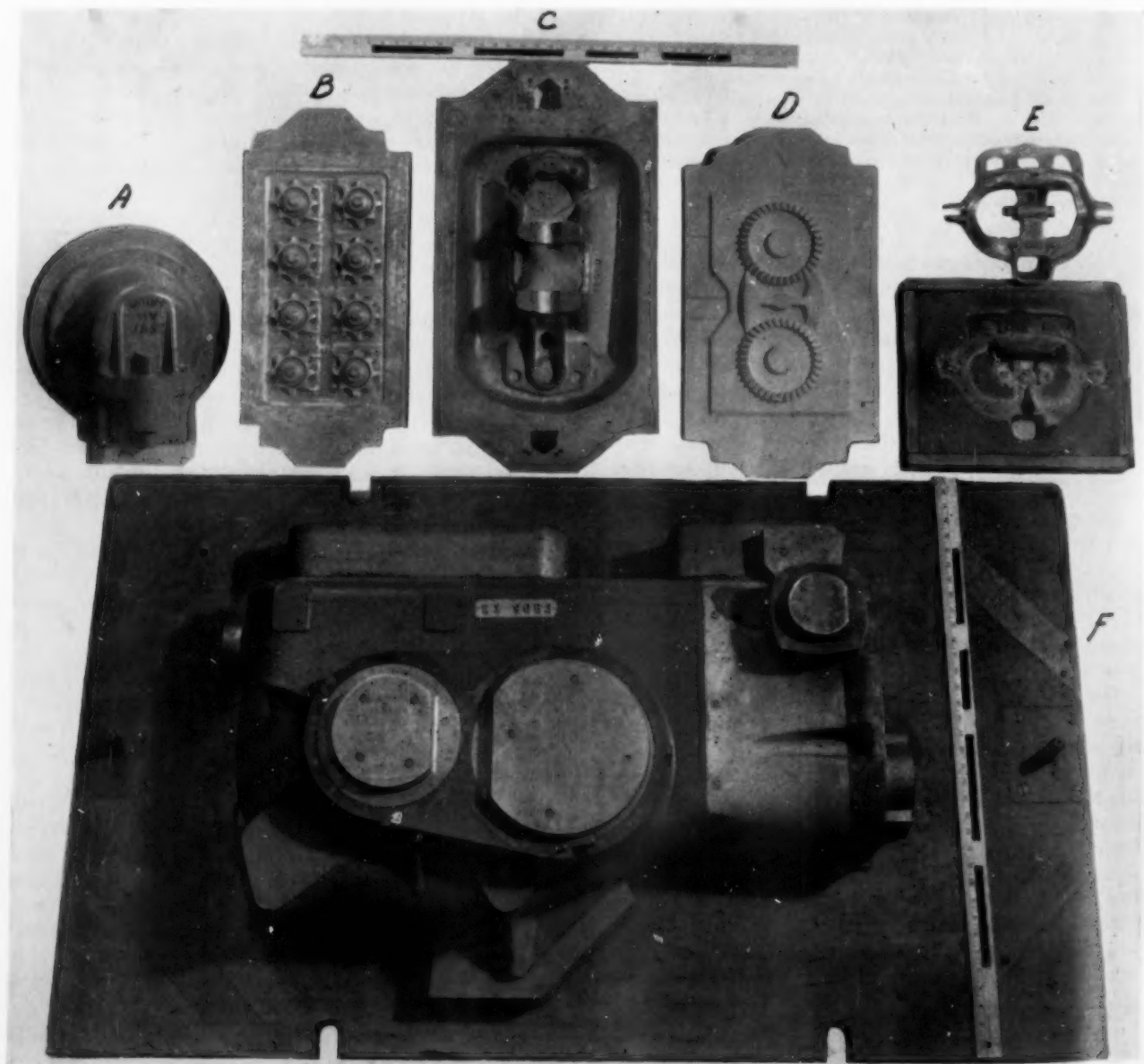
ment castings. These are shown because they are typical of the type of patterns in use today. When 850 patterns for this company were analyzed to establish the most modern rigging, it was found that by standardizing on flask sizes and with proper mounting, the resultant decrease in production costs would amortize the cost for patterns, core boxes, and flasks in approximately six months. Patterns, flasks, machines, etc., should be checked frequently because maximum unit production of high quality castings can be obtained only when the production equipment is in first class condition.

#### Personnel and Good Housekeeping

Foundries have found it increasingly difficult to secure and keep the quantity and quality of labor, supervision and management necessary to remain

Fig. 1—Photograph of typical patterns now in use for production of farm implement castings. Moderniza-

tion of rigging permits increased production per man-hour and improved quality of product.





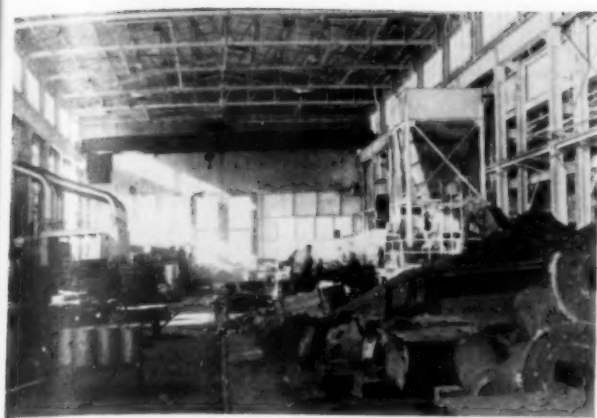


Fig. 2—Cleaning room after modernization. Note white paint increases level of light, which makes for better work and increased output per man-hour.

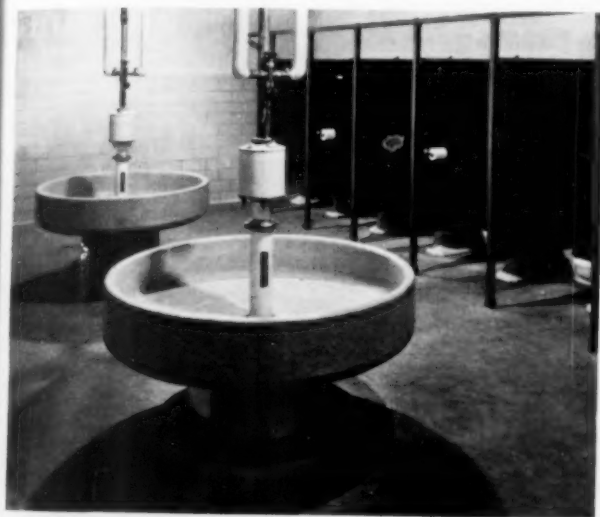


Fig. 3—A modern equipped wash room.



(Courtesy, American Laundry Machinery Co., Rochester, N. Y.)

Fig. 4—Molding floor in a jobbing foundry. Note good housekeeping.

competitive with other materials and processes. Part of the reason, of course, is that in many foundries the work is heavy, hot, and dirty. Wages have been low in many parts of the country, although currently, the foundry industry is paying wages well "in line" with other industry. Good housekeeping practices help to make the foundry a better place to work. Good housekeeping pays dividends because workmen react to their surroundings and a dirty, disorderly shop is not conducive to economical, high quality work. Orderly disorder is the general condition in a well kept jobbing type foundry. The production foundry—particularly if mechanized, lends itself to more orderly operation. Figure 2 is a photo of a castings cleaning room after modernization.

Adequate, clean personnel facilities are a necessary part of the modern foundry, (Fig. 3). Good lighting 20 to 40 ft.-candles in the work zones, will pay for itself quickly in less scrap and more production. Racks, shelves, and benches for tools, rigging and cores should be provided in the area most convenient for use. Adequate and orderly flask storage not only helps to keep the flasks in good condition, but also to reduce flask handing costs. The only successful good housekeeping programs are those where the individual workmen keep things orderly and "in the proper place" as they go along. It is not economically practical to "pick up and clean up" after everyone. Each foreman must police his department and over a period of time, good housekeeping practices become good habits and no excessive cost is involved. The author's experience shows that the clean, orderly plant usually is the low cost plant, making a profit. Figure 4 shows a photograph of a molding floor in a jobbing foundry.

### Organization

The organization should be simple, with clear lines of authority and responsibility so that each department head knows and can carry out his job. Foremen should cooperate with each other to make certain that scheduling will insure cores, flasks, patterns, etc., at the right station at the right time. All of the many variables must be controlled if high quality, low cost castings are to result.

The foundry must know and control its costs. The old tendency to include many costs in indirect costs (example: Cleaning and Finishing) results in unequal distribution of costs. In more than one instance, profitable, desirable work has been lost because selling prices became too high; and unprofitable work retained, at below cost figures, because of failure to properly allocate costs. If a wage incentive program is used—and it is strongly recommended—the same standard data used to pay for the work may be used to price the jobs for quotation and control costs. This system makes it quite simple to establish standard costs and greatly simplifies cost control by calling attention to those jobs being produced outside a reasonable range of the standard. Every foundry to stay in business, to pay an adequate return to labor, management and owners, must know its costs and sell its product at a profit. In one foundry, recent analysis of 30 jobs indicated that 26 of these jobs were produced at a loss or

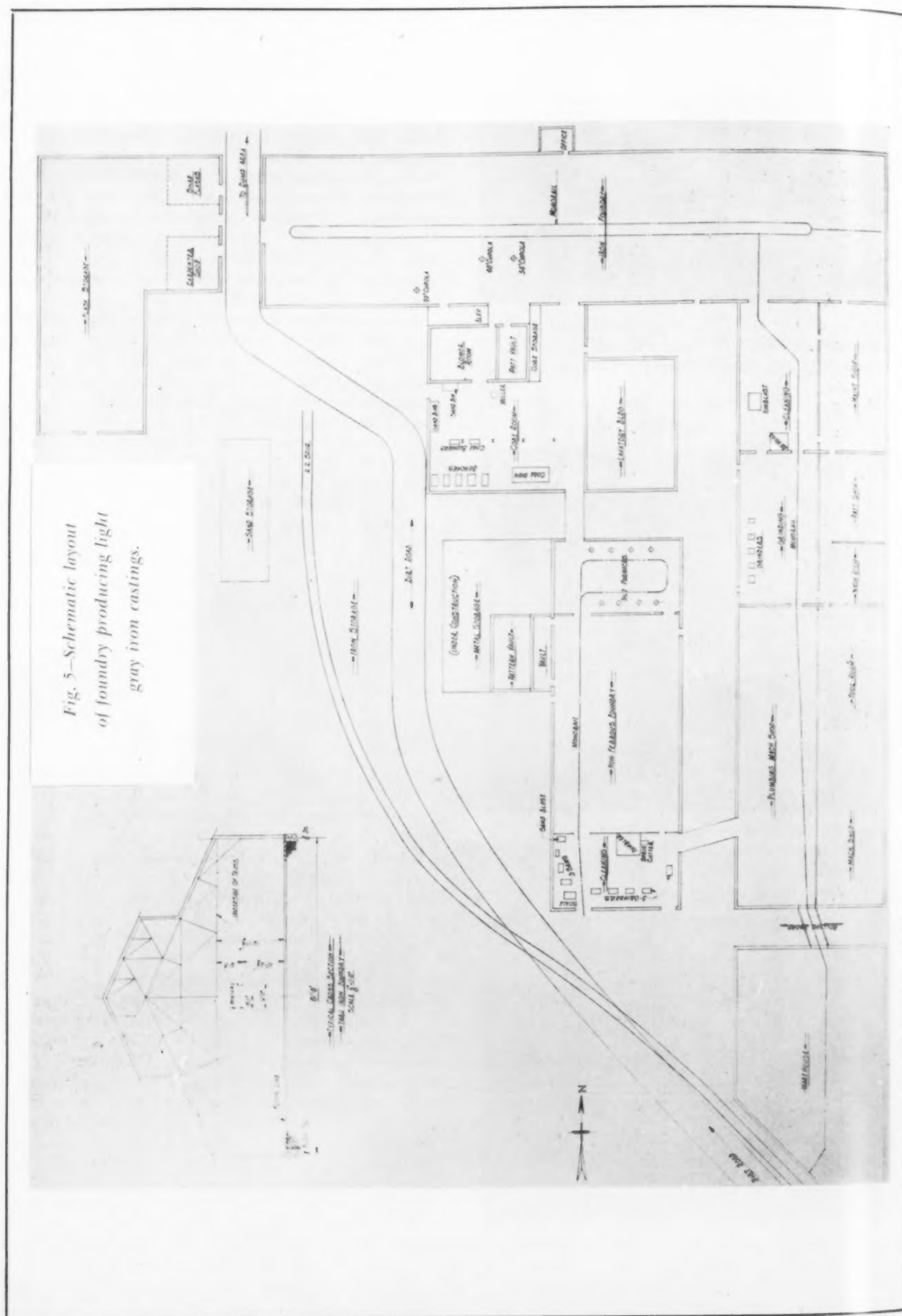
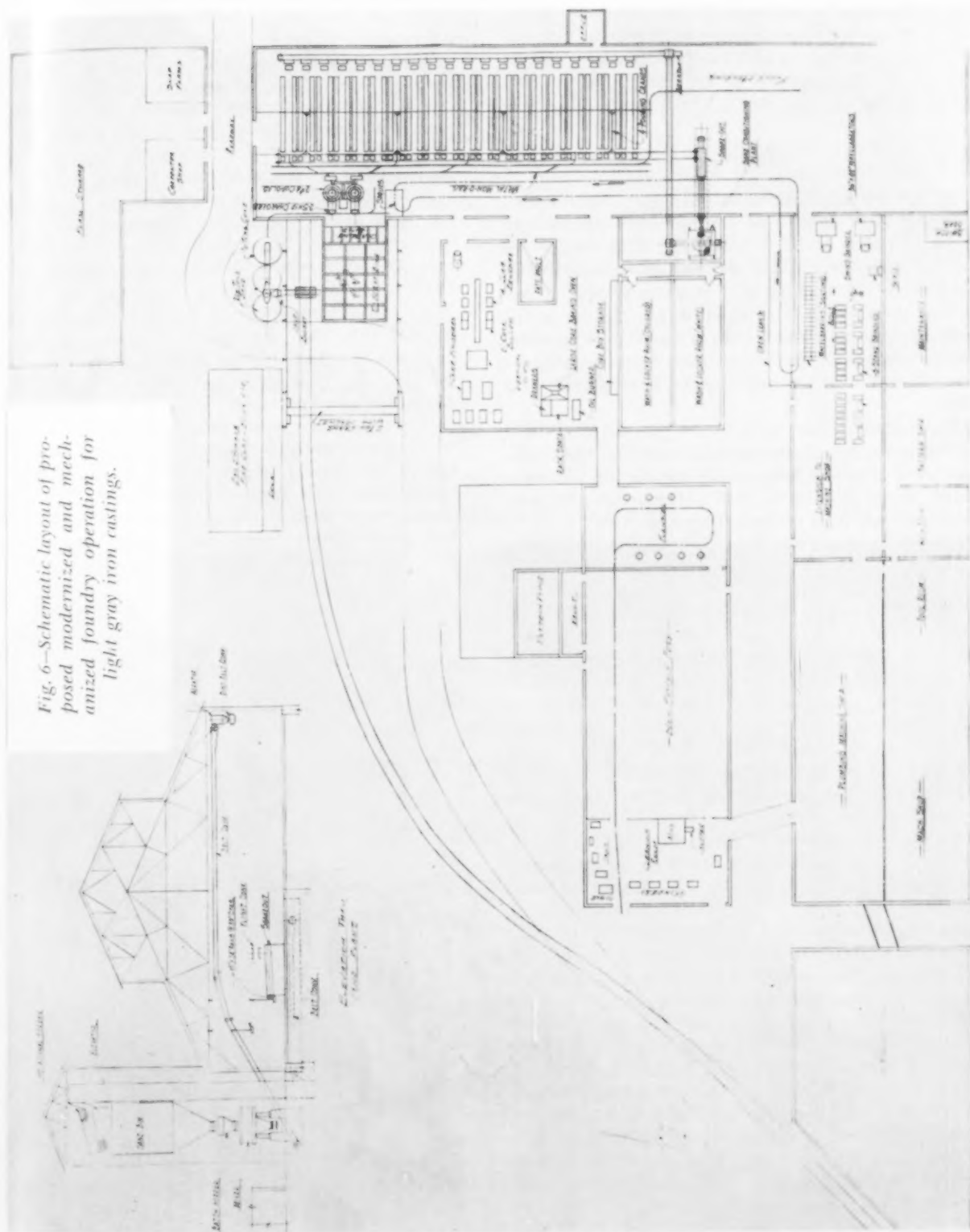


Fig. 6—Schematic layout of proposed modernized and mechanized foundry operation for light gray iron castings.





with less than reasonable profit. Prices were adjusted, realistically, and a wage incentive plan installed which resulted in higher earnings for the men of the company.

It should be emphasized again that no two foundries present the same problems or permit the same "treatment" to improve operating results. It may be interesting to check the figures and analysis on a few projects to illustrate one approach to the problem of foundry modernization and the results which can be obtained.

#### Case No. 1

Figure 5 is a schematic layout of a gray iron foundry producing light gray iron castings with an average weight of 1.5 lb each. The average daily melt is 10 tons (approximately) and the annual production is 1420 tons in 271 working days at a cost of \$201.10 per net ton. Average daily production per molder is 117 molds in 6½ hr with 1½ hr allowed for pouring. Average yield is 55 per cent. The payroll carries 93 men (26 molders). Annual profit is approximately \$5,000.

This foundry was anxious to "modernize" and mechanize to reduce costs and increase its profit. No funds were available for capital investment. Analysis showed many jobs being sold at less than cost and prices were adjusted realistically on all jobs.

In order to secure maximum production from existing facilities, it was necessary that the "producers" produce full time. Although the foundry had operated for 39 years with molders pouring their own molds and other departments (Core, Cleaning, etc.) furnishing men for shifting and dumping, it finally was agreed that piece rates (set by barter and with inequities for both labor and management) should be adjusted and a separate pouring, shifting, and dumping crew installed, with the molders molding a full 8 hr. Both supervision and molders said it "wouldn't work," and for a few weeks, it did not. Scrap was high, and an excessive number of men were required to shift and pour. Supervisors were taken to other foundries where similar operations could be observed and with proper training, the pouring crew was established.

The molder's production is nearly always controlled by the number of molds he can or wants to pour off, not by his capacity to produce with the production facilities. This was amply demonstrated by the fact that with the same number of molders, results in 1946 were approximately as follows: Average casting weight was 1.5 lb each; daily melt was approximately 31,000 lb, annual production was 2,011 tons in 220 working days, at a cost of \$198.77 per net ton. Average daily production per molder was 163 molds in 8 hr, an increase of 48 per cent, although molding time actually increased only 23 per cent. The payroll carried 98 men (26 molders). Annual profit was approximately \$50,000. Average yield was 57 per cent due partly to standardization of flask sizes, using larger flasks with more castings per mold and re-rigging for some cope and drag work on pin-lift machines. The foundry was faced with higher material costs and absorbed a 10-cent per hour wage increase.

In 1947 this same foundry produced 2,278 tons, in

219 working days, at a cost of \$226.00 per net ton. Average daily melt was approximately 39,000 lb. Production per molder increased to 170 molds per day. Payroll carried 104 men (30 molders). Annual profit was approximately \$55,000. Higher material costs of \$20.00 per ton and an 8-cent per hour wage increase are responsible for the average cost increase.

The substantial improvement in production with realistic sales prices resulted in profits which make possible progressive modernization of this plant to permit production from 20 to 60 tons of melt per day under the best possible conditions. Flexibility is required not only for wide range in total tonnage, but also in size and weight of castings to be produced. Figure 6 shows a schematic layout of the proposed mechanization of this plant. The cupolas are to be replaced with new equipment, relocated to permit efficient handling and charging of the raw material. With production up to 20 tons per day, the analysis showed that the yard crane facilities could not be economically justified, but at 40 or more tons of melt per day, the yard facilities become a "must" for efficient, low cost operation.

Skip-type chargers are proposed to load the cupolas which are provided for front slagging and with a forehearth for desulphurizing and to maintain virtually continuous metal from the cupola (Fig. 7). Molten metal was to be distributed by monorail metal carrier and poured from covered ladles mounted in pouring devices.



(Courtesy, Whiting Corp., Harvey, Ill.)

Fig. 7—Side dump, skip hoist charger for cupola.

A complete modern sand conditioning, sand and mold handling system utilizing roller conveyer was to provide required flexibility for handling the molds (Fig. 8). All sand was to be mulled and aerated and delivered by gravity from the molder's hoppers into the molds. Sand and castings, after pouring, are dumped onto conveyer (underground) for delivery to a shakeout to separate sand and castings. Castings are then carried in pans on a trolley conveyer for cooling, drying, (delivery of sprue to charging area) and delivery to the Cleaning Room.

The Core and Cleaning Rooms also were rearranged to permit increased unit and total production. Part of the Cleaning and Core Room rearrangement has been accomplished with substantially improved efficiency. Presently the new cupolas are being built and when placed in use, a new, realistic wage incentive based on standard data accumulated by a time study should permit further increase in unit and total production (approximately 16 per cent).

The mechanization program is expected to be completed by 1949. The estimated difference in cost based

on condition at the time of the survey for varying tonnages are as follows:

	1945 (Actual)	Estimated Future	
Production	1406 Tons	3300 Tons	4400 Tons
Cost per ton	\$201.10	\$163.60	\$158.70
Estimated amount available to pay for modernization		\$123,750	\$186,670

Based upon total modernization cost of \$300,000, the investment should be amortized in less than three years of operation. It should be emphasized again that the program is laid out to be installed progressively and that the initial improvements and economies were accomplished without major capital expenditures.

#### Case No. 2

Figure 9 shows a schematic layout of a jobbing gray iron foundry producing miscellaneous automobile and farm implement castings, ranging in weight from 1

cupolas with a fore-  
virtually  
Molten  
I carrier  
pouring

with ladle handling facility and also dumping hatches in foreground.

(Courtesy, National Engineering Co., Chicago)



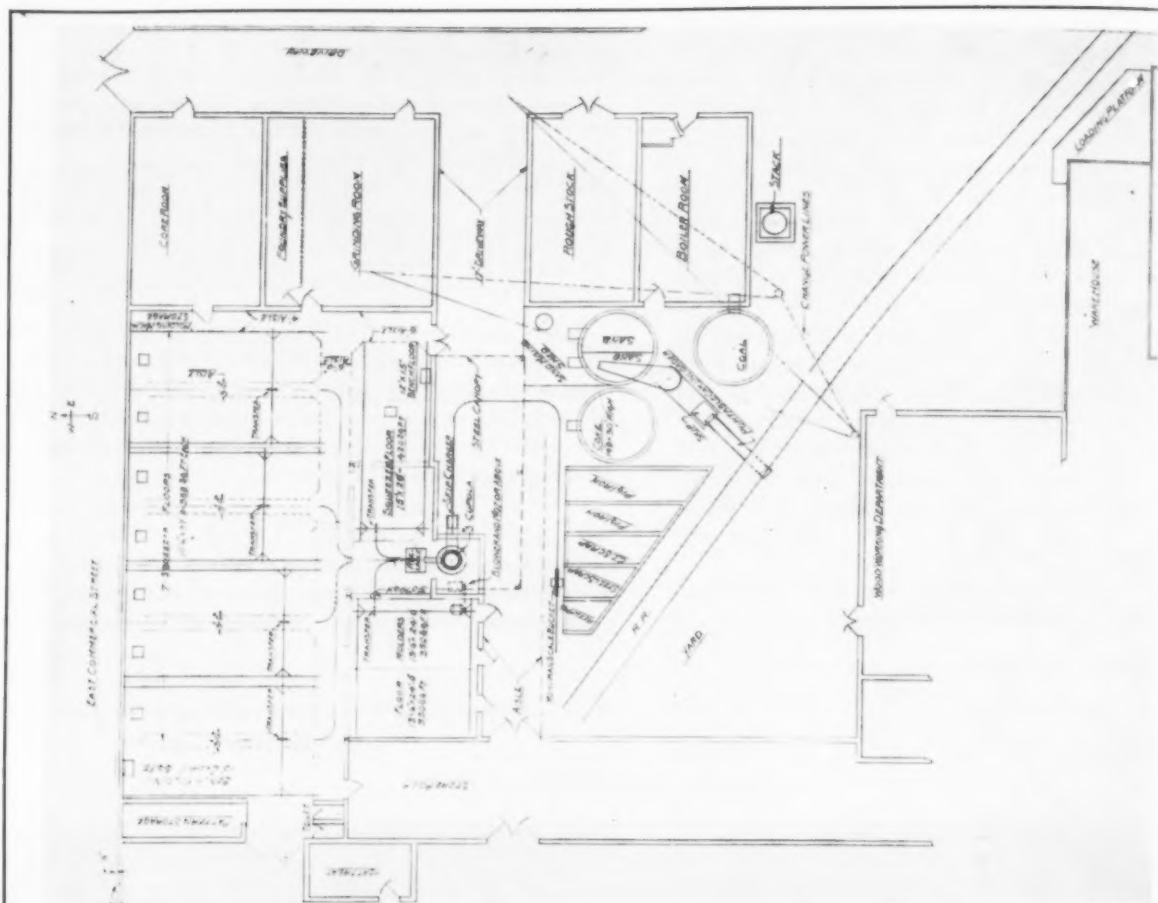


Fig. 10—Schematic layout of same gray iron foundry as in Fig. 9 after mechanization.

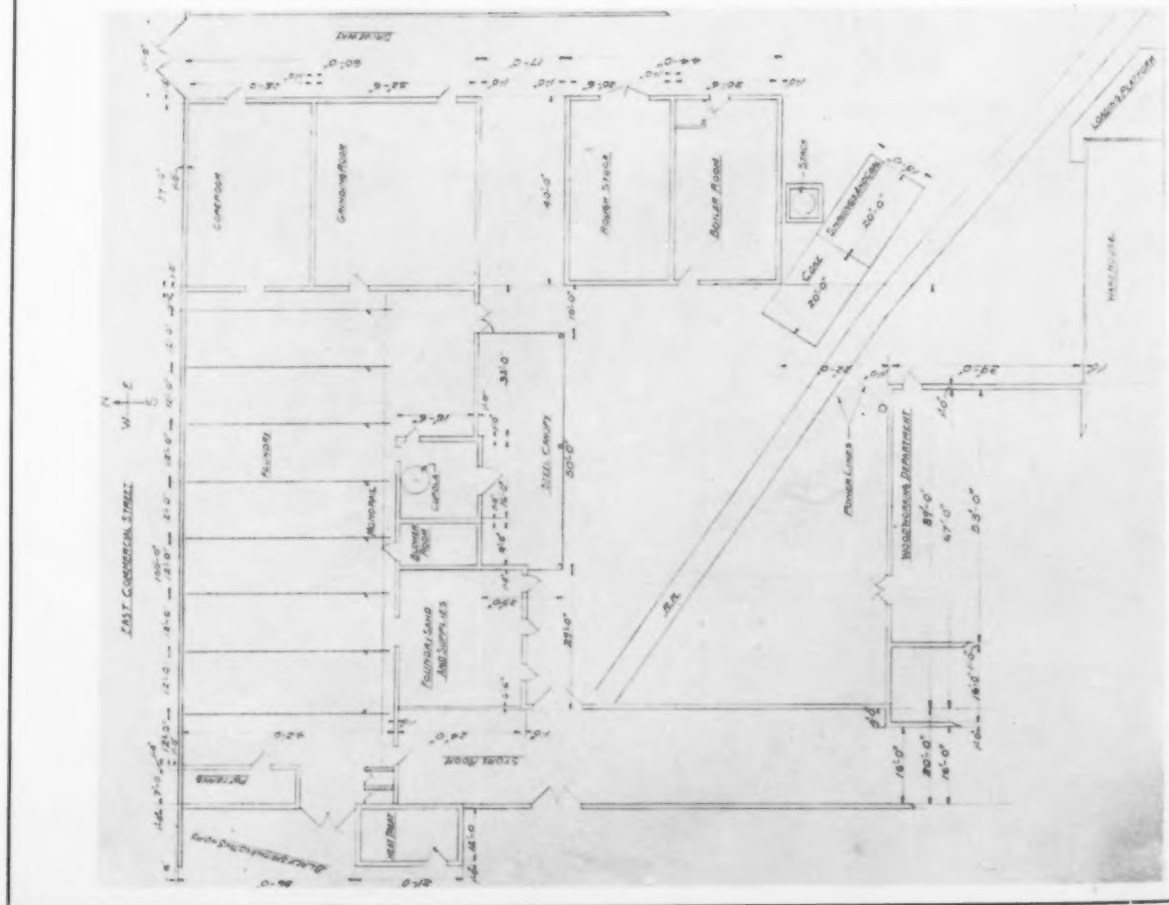


Fig. 9—Schematic layout of small gray iron foundry before modernization.



to 50 lb. At the time of the survey (July, 1946), this foundry was pouring 4 to 8 tons of melt every other day. On alternate days, the workers shook out the molds and cleaned the castings. Production was low and costs were high. Floor molders produced 5 to 7 molds per day, squeezer molders 10 molds per day. Yield was 71 per cent. Man hours were 86 per ton of good castings. Total area for foundry was 8300 sq ft. Total manning was 29 men.

Figure 10 shows a schematic layout of the plant as "modernized." The cupola has been relocated to permit approximately 38 per cent greater area in the foundry. A skip hoist charger and weigh bucket has been provided to facilitate charging the cupola. Bins for raw materials have been provided. Three additional molding floors were provided and all floors are now operated daily instead of every other day. A monorail distribution system has been provided to facilitate metal delivery and pouring off. The number of men required to operate the plant has been increased to 34. Piece rates on the molds resulted in increased unit and total production so that approximately 6 tons of good castings per day are produced. Man hours are reduced to 45 per ton. Total cost of the "modernization" was \$24,000. The above condition is not typical, by any means, but is one actual case. Figure 11 shows the molding floor with monorail pouring cranes.

### Case No. 3

Figure 12 shows a schematic layout for a Brass Foundry producing low pressure valves and some jobbing work. Figure 13 shows a photograph of a similar installation. Briefly, the unit consists of 12 molding stations with overhead sand delivery to half patterns

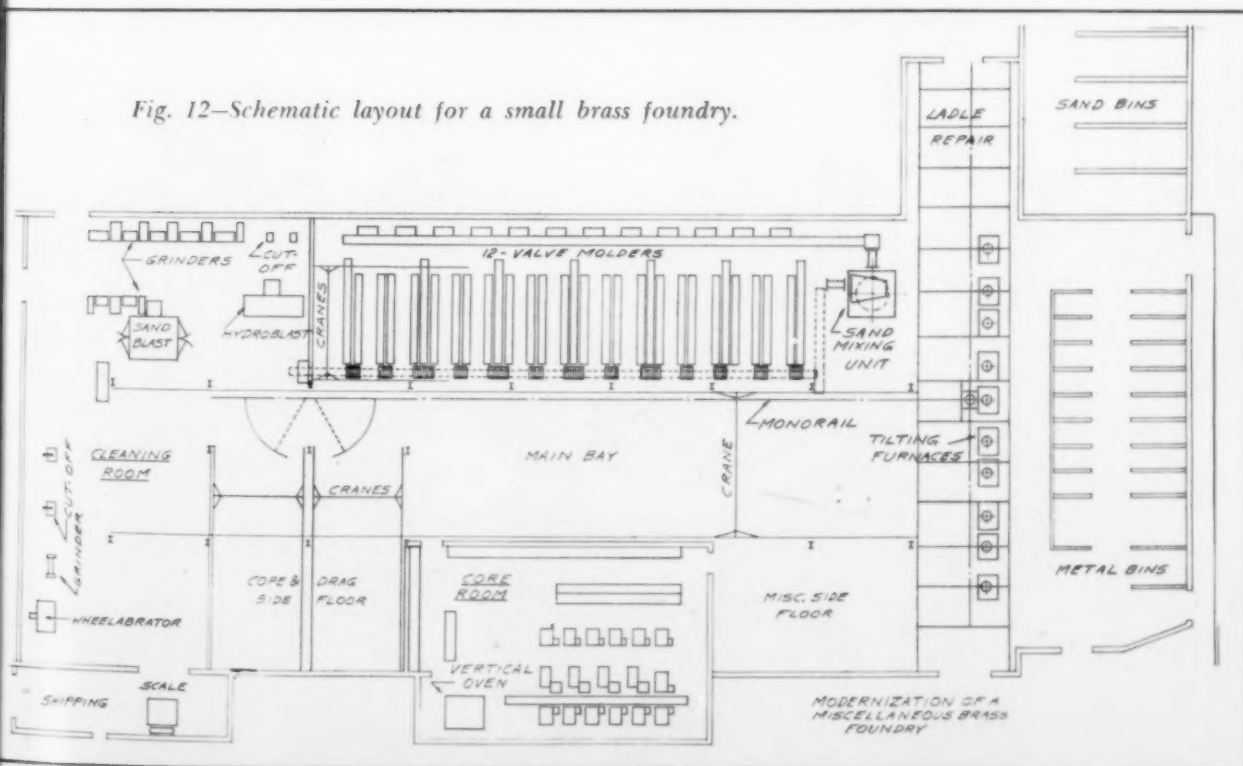


Fig. 11—Molding floor in small gray iron foundry. Note pouring monorail, ladle, and pouring device

mounted on pin-lift molding machines. Steel flasks are used with 3-in. drag and 4-in. cope (13 in. x 18 in.). Molders produce a total of approximately 250 molds in 8 hr averaging from 18 to 40 molds per hr, depending upon the number of cores to be set.

Before the mechanical unit was installed, molders produced 95 to 110 molds in 8 hr and poured off. Each mold averages approximately 15 lb of metal for a total daily requirement of 45,000 lb of metal. Metal is melted in gas-fired, nose-tilt crucibles with a capacity of 700 lb each. A melting and pouring crew of five men charge the furnaces, prepare the metal, and pour the molds. Covered pouring ladles, with skimming block, mounted in pouring devices are used for pouring, one man to a ladle. Under the "old" method, a

Fig. 12—Schematic layout for a small brass foundry.



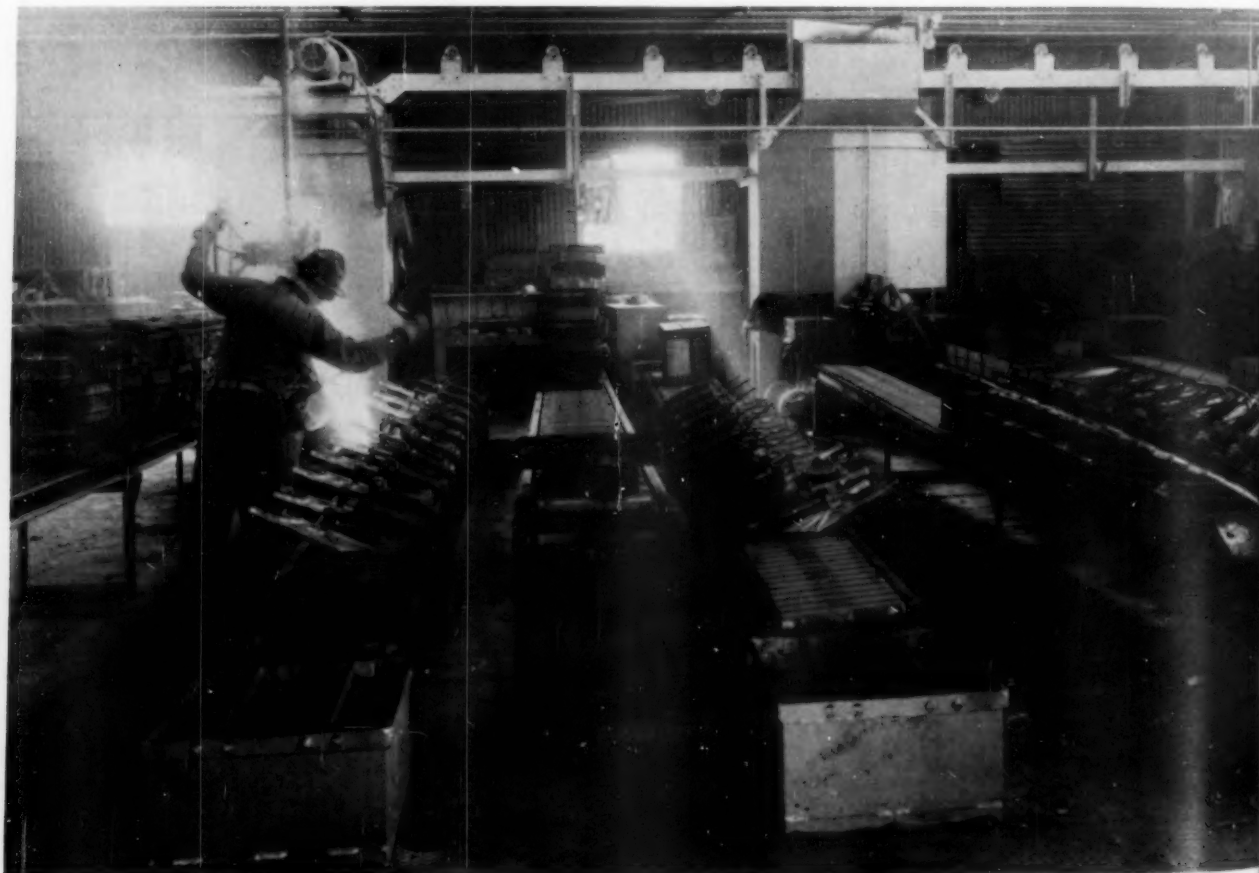
3-man pouring crew (2 on ladle, 1 on skimming) was used. After pouring the molds are dumped at the end of the roller conveyer and castings returned to the Cleaning Room by wheelbarrow. The sand is conditioned in a unit where all refuse is removed, the cleaned sand mulled, aerated, and distributed by overhead belt to the molder's hoppers.

The castings are cropped to remove gates, cleaned in a wet blast machine, rough inspected, sorted for grinding, weighed (weigh counted) and delivered to the grinders. Total manning in the foundry for the valve production is approximately 50 people, including supervision. All the heavy work has been eliminated. The furnaces are hooded and exhausted to reduce heat and eliminate fumes. General ventilation keeps the atmosphere clear.

#### "Small" Malleable Foundry

While this paper is to discuss modernization of the small foundry, the term "small" is a relative rather than a generic term, and in the United States a "small" malleable foundry is one producing approximately 5,000 tons of finished castings per year. Therefore, we have included in this paper the following discussion of a "small" malleable foundry.

*Fig. 13—A similar brass foundry installation. Note pouring monorail with one man pouring; also use of special weight device to eliminate shifting of weights.*



#### Case No. 4

Figure 14 shows a schematic layout for a malleable foundry producing castings with an average weight of 0.409 lb each. Approximately 40 per cent of the production consists of malleable chain links and fixtures, and the remainder constitutes miscellaneous jobbing castings. This plant operated for many years with a conventional type air furnace and molding machines or benches with which molds were made. The molds were placed on the floor and poured by the molders, with such help as was necessary being drawn from other departments. Under the old method of operation, the molders produced an average of approximately 115 molds per man per day. The design of the pattern and size of flask used was determined not on the basis of securing the maximum production per man-hour of effort, but on the basis of how much metal the molder was willing to pour at the end of the day.

When this foundry developed its modernization program, a comprehensive analysis disclosed that the castings "hit the floor" 10 to 12 times from the time the mold was shaken out until the castings were packed in the bags for shipment. In this foundry the castings were first cleaned in the hard condition by tumbling. After inspection, approximately 80 per cent of the work was ground in the hard condition before delivery to the Annealing Department. In the Annealing Department, periodic ovens were used on a 6-day cycle. Stacks of pots were hand packed, 4-high, and after annealing these stacks of pots were dumped over

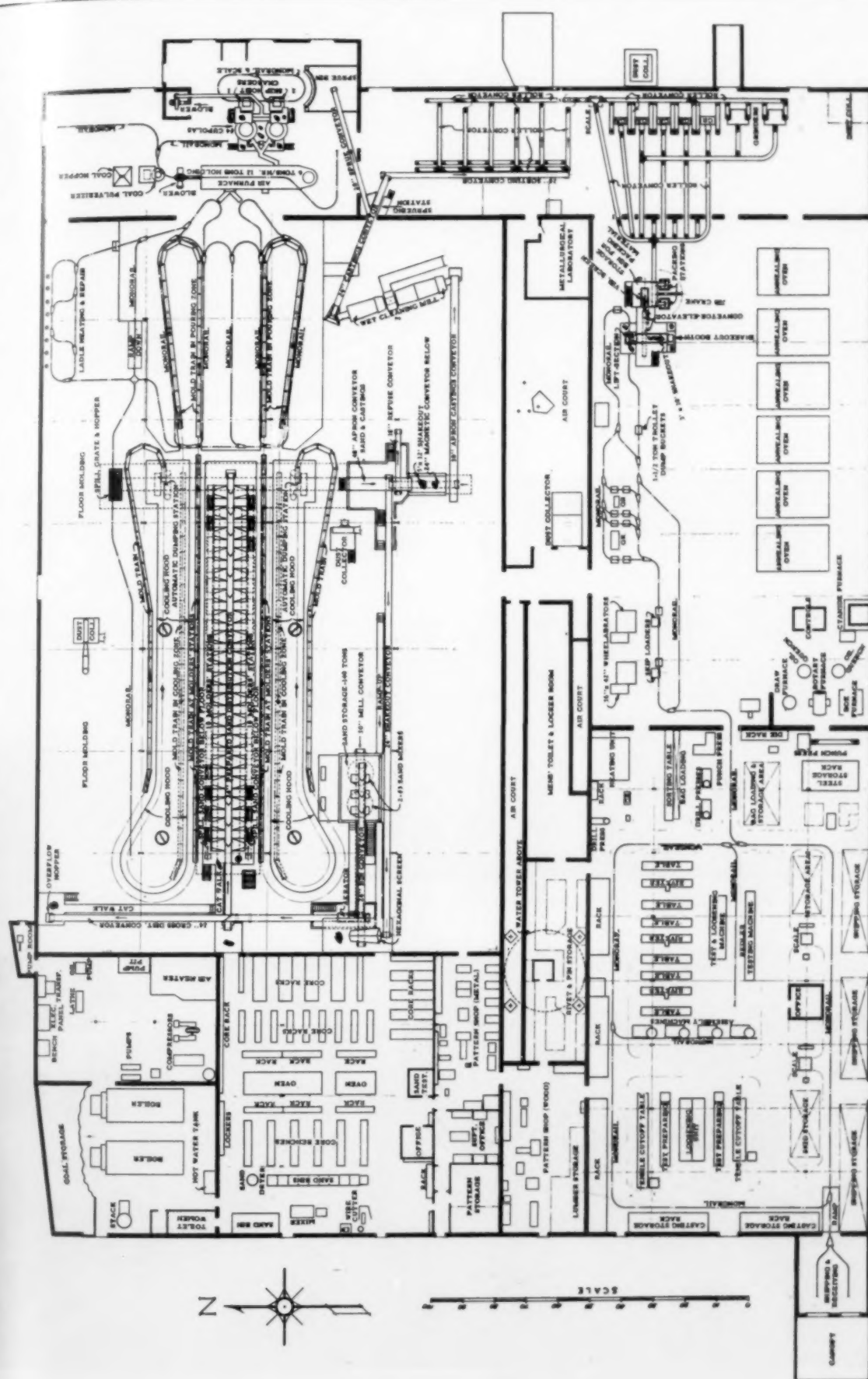
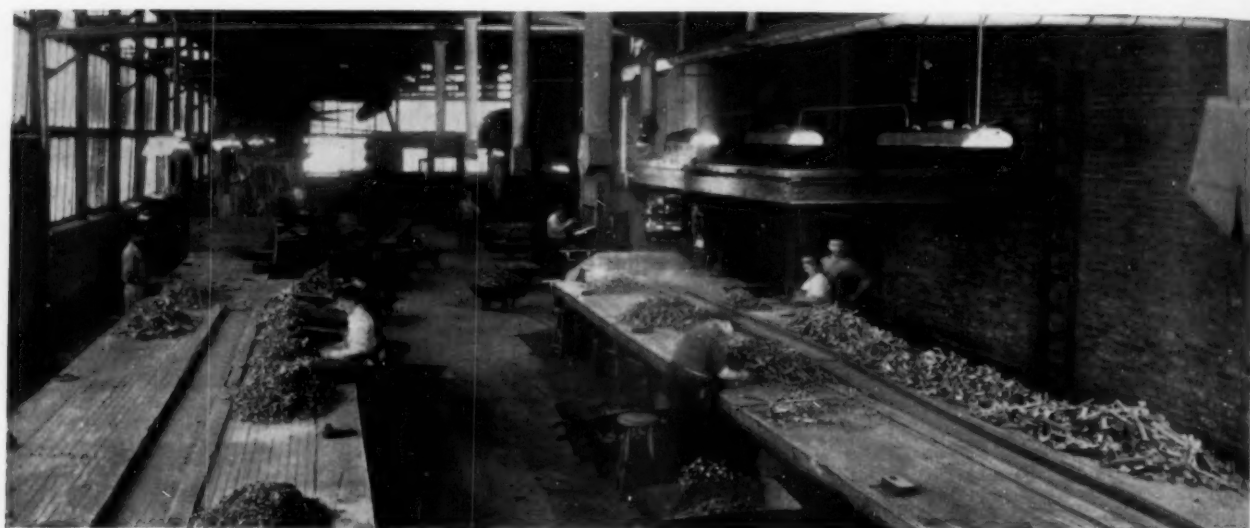


Fig. 14—Schematic layout of a modernized malleable iron foundry.





*Fig. 15A—Molding floor and batch furnaces before modernization.*



*Fig. 15B—Hard iron sorting with hard iron grinding in the background.*



*Fig. 15C—Annealing department. Note packing material and pots scattered on the floor.*



Fig. 16A (top)—One train of mold conveyer in the pour zone. Note in the background the two lines of molders' hoppers.

Fig. 16B (center)—Inspection and sorting prior to delivery to hard iron grinding and pot packing on roller conveyer.

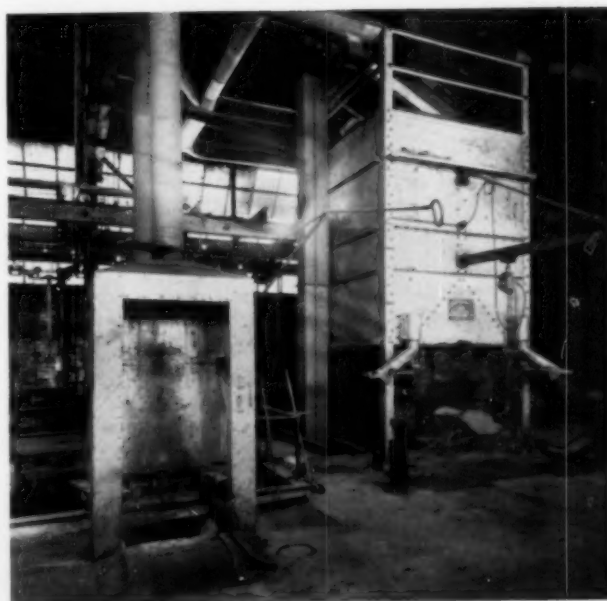


Fig. 16C (bottom)—Pot packing unit and pot dumping station which eliminates all packing material and casting on floor.

on the floor. The castings were forked out of the packing material with the usual result of a hot, dirty, laborious job in the Annealing Department. The annealing pot rings had a life of approximately 30 heats. The survey also disclosed that the total man-hours per ton of good castings shipped averaged approximately 115 to 120, based on shipments of approximately 325 tons of product per month.

The first step in the foundry modernization program was an analysis of all of the patterns and rigging for the job. This resulted in a redesign of many patterns, which permitted approximately a 10 per cent increase in the weight of castings per mold. Pattern analysis also established the fact that many molds that had been poured two-up, could, by redesign, be poured one-up with resultant increase in yield.

This particular foundry had a wide range of work to be made with some molds having as little as 6 lb of metal per mold and others as much as 30 lb of metal per mold. It was considered most satisfactory to design a unit which would permit pouring the mold while the mold was stationary so that it could be properly identified and poured in the proper manner. Figures 15 and 16 show conditions of the foundry modernization.

In Fig. 14 (schematic layout of modernized foundry) the foundry was mechanized by the installation of facilities which are now to be described.

#### Melting Department

The requirement of efficient production with metal requirements of 25 to 60 tons of melt per day resulted in decision to install one ventilated bottom, powdered coal-fired furnace with a rated bath capacity of approximately 12 tons of metal. The furnace was to be served with metal melted in a cupola lined to 38 in., which would give a normal melting rate of approximately 5 tons per hour. Two cupolas were provided to be used on alternate days to permit patching during the day time in order to insure the reconditioning of the cupola while melting supervision was available to inspect it. The cupolas are charged by means of two skip hoist chargers with a bucket carrying a 1200-lb batch. A batch weigh larry is mounted on a monorail to permit the make-up of the charges, each charge being approximately 600 lb with two charges making up a batch. The batch is discharged from the weigh larry into the bucket of the charger after which the starting button is pushed, and the bucket delivers the batch into the cupola and returns automatically to the loading position. Normal operation requires approximately 8 to 10 batches per hour. Two men ordinarily are able to charge the cupolas.

Normal practice is to start melting approximately 2 hr before molding and to "drop the bottom" approximately 2 hr before the end of the day. Fracture bars are taken periodically to control metallurgical results. Metal of the following composition is consistently produced: C-2.50-2.60, Si-1.15-1.20, Mn-0.52-0.60, S-0.10-0.12, P-0.16.

#### Molding

The molding unit consists of sand conditioning, sand and mold handling facilities. There are 36 mold-

ing stations, each provided with an overhead sand hopper, holding approximately 1½ tons of properly conditioned sand. A jolt squeeze type molding machine is located under each of the molding hoppers with a spill grate and belt under the machine to carry the spill sand back to the sand conditioning equipment. Each molder is provided with patterns, flasks (skip flasks), bottom boards, and cores, all in good working condition so that there are no interferences to the molder securing maximum production. It is the molder's job to produce molds and he is serviced with all requirements to aid him to produce the maximum number of molds during the working day. After the molds are made, they are placed on an "index" type of mold conveyer, consisting of 24 cars attached to a locomotive to make up a "train." The capacity of each train is the equivalent of 72, 12 in. x 18 in. molds or an average of four molds per molding station.

At this point, it should be noted that there are two separate units, each with 18 molders. Each unit has four separate "trains," each with 24 cars and a locomotive. Both units are served with sand from a common sand conditioning unit. Each train operates on a predetermined timed cycle and normal operating procedure is such that there are 85 to 90 cycles per 8-hr day, permitting an average productive capacity for each molding station of approximately 360 molds. Actually, some patterns permit faster production, others slower, so that the actual production in the unit varies from approximately 240 molds per molder per 8 hr to a maximum of 450 molds per molder in 8 hr.

#### Description of Cycle

Train No. 1 moves into the molding zone and stops at a predetermined position so that bottom boards are returned to the same molder on each cycle. Train No. 1 remains stationary for approximately 4½ min during which time the molders fill the trays with molds. At the end of the cycle, a whistle blows and Train No. 1 moves at approximately 120 ft per min into the pouring zone and stops in a predetermined position so that the same molds always come to the same position for pouring. This allows maximum identification of the molds and permits pouring with a minimum amount of pouring scrap.

In the meantime, while Train No. 1 was moving into the pouring zone, Train No. 2 moved into the molding zone. During the indexing period, the molders were busy making an additional mold to be ready to place it on the conveyer tray as soon as the train stopped. In the meantime, on conveyer No. 1 in the pouring zone, jackets and weights were placed and the molds poured from 250-lb covered ladles suspended in pouring devices from trolleys mounted on monorail. These monorail loops are positioned so that the pourer catches metal in the ladle at the spout from the air furnace and pours directly into the molds. This permits a minimum loss of temperature which is extremely important on this light class of work. Three jacket and weight shifters and three pourers for each unit are able to service the work from the 18 molding stations.



At the end of the second cycle, Train No. 1 moves out of the pouring zone into a cooling zone, while Train No. 2 moves into the pouring zone, and Train No. 3 moves into the molding zone. At the end of another cycle, Train No. 1 moves into the dumping zone and is "picked up" by a tractor drive and carried through the dumping zone at approximately 25 ft per min, where an automatic device tilts each mold conveyer tray, sliding the sand and castings onto a gathering steel apron conveyer below, while the bottom boards remain on the tray. The sand and castings from both units are delivered by the apron conveyer to a mechanical, vibrating shakeout. The shakeout separates the sand and castings, and delivers the castings onto an apron conveyer, which is operated by a time relay. This time relay also controls the cycling of a wet tumbling mill.

#### Sand Conditioning

Sand from the shakeout is delivered by belt conveyer over a magnetic pulley to remove magnetic refuse. Then by incline belt it goes to a screen to remove core butts and other refuse. The screen is totally enclosed and exhausted with fines control to permit control of the permeability of the sand without use of excessive amounts of new sand. The screened and cleaned sand is delivered to a storage bin provided with measuring hoppers to deliver the sand into batch type mullers where new sand, seacoal, clay and moisture are added. The mixers are equipped with cooling hoods so that cool sand may be delivered at the molding stations. After a mixing cycle of approximately 3 min, the sand is discharged onto a belt conveyer, which delivers it to an aerator. From the aerator it goes by belt conveyer to the molding stations where it is plowed off into the various molders' hoppers. One man at the mixer and one man at the plow-off stations permit the conditioning and delivery of the sand for the 36 molders. Sand for additional short run and experimental floor work also is conditioned in the unit and delivered to the floors by tractor from the hoppers provided to receive this sand.

#### Casting Cooling and Cleaning

The castings delivered from the shakeout to the apron conveyer are carried by the casting apron conveyer at a speed of approximately 2 ft per min for 12 min to permit the accumulation of a batch of castings on the apron. At the end of 13 min, the apron automatically speeds up to a speed of approximately 30 ft per min so that within 2 min the apron is emptied and the batch of castings is delivered to the wet tumbling mill. The castings are tumbled for 7 min during which time they are rough cleaned and many of the castings are automatically sprued. At the end of the 7-min cleaning cycle, the wet mill automatically reverses itself and during the next 6 min delivers the castings to an inclined conveyer. This conveyer carries them past a sprueing station where the sprue is placed on a conveyer for delivery to the cupola charge make-up area. The castings, after the sprue has been removed, are delivered to the sorting conveyer where women sort the castings into pans located on roller

conveyer adjacent to the belt. These pans, handled on roller conveyer, are carried past inspection and fining stations. After which, also on roller conveyer, they are transported across the scale where they are weigh-counted and delivered either to the pot packing stations or to the hard iron grinders. The hard iron grinders work from a full pan to an empty pan, and the ground castings are delivered on roller conveyer to the pot packing area.

#### Pot Packing and Unpacking

As you will note from the photograph of the Annealing Department (Fig. 15 and 16) all of the heavy, hot, dirty work in the Annealing Department has been eliminated by the use of a pot packing and unpacking unit. The method of operation in this department now is as follows: The trucker delivers the stack of pots to the annealing booth where the bottom board is clamped with an air clamp. Then the truck driver drops the bottom, and backs out with the pedestal. The castings and packing material fall by gravity onto a shakeout, which separates packing material and castings. The castings are delivered in buckets mounted on a monorail for transfer to the grinders for soft iron grinding, when required, or to the shot blast machines for final cleaning. The packing material falls through the openings on the shakeout deck into an endless chain conveyer which elevates it and delivers it to a double-deck screen. There the oversize material and fines are removed, and the cleaned packing material is delivered into a storage bin. This storage bin is equipped with two swivel type outlets with gates to permit delivery of the packing material by gravity into the rings of castings as they are progressively packed. After the castings and packing material have dropped out of the stack of pots, the trucker replaces the pedestal, releases the air clamp, and delivers the stack of pots including the pedestal to one of two platforms. These platforms are hydraulically operated to permit raising and lowering the stack of pots while they are progressively packed with castings and packing material.

After the pots are filled with castings and packing material, the pot is vibrated and after being "muddled" the trucker places the stack in the annealing furnace. It is reported that the use of clean packing material permits faster heat transfer so that the annealing cycle has been reduced approximately 12 hr. It also is reported that heat life on the pot rings has been increased to approximately 50 heats, compared with 30 under the old method. Three men are able to pack, unpack, load and unload the ovens for production of approximately 5,000 tons of castings per year. After the annealed castings have been cleaned, they are delivered to the sorting tables in the Shipping Department for bagging prior to shipment.

#### Ventilation

Wherever dust and smoke are created, the facilities are hooded, housed, and the air is exhausted so that optimum working conditions are maintained. The unit heaters in the foundry provide adequate heating for sufficient intake air so that the atmosphere of the

foundry does not become "starved" and good ventilation job is accomplished. In many foundries the exhaust equipment creates a partial vacuum in the foundry so that smoke and fumes are not removed. In order to have air circulation, it is necessary to furnish adequate air input into the foundry.

#### Production Results

It is reported that the 36 molding stations are producing molds at a rate to permit shipment of approximately 5,000 tons of good castings per year with an average casting weight of 0.409 lb each and that the man hours per ton have been reduced to approximately 70 to 75 man-hours. All of the hard, hot, dirty jobs in the foundry have been either eliminated or provided with labor servicing devices, which permit the work to be done with minimum fatigue under the best working conditions. The total cost of such a modernization and mechanization program is returned to the foundry within 30 months of operation out of the reduced costs of operation. At the same time, the closer control of operating variables permits the production of high quality castings at low cost.

Each of the above distinctly different foundries has been modernized and all responded to modernization permitting cost reduction which return sufficient money to pay for the changes in from 1 to 4 years of operation.

#### Summary

In summing up, the author lists some of the items which should be investigated and properly modernized to secure maximum results.

- Sales methods and prices
- Organization and personnel
- Cost control
- Patterns, rigging, production equipment and methods
- Wage incentives
- Plant layout, material handling, good house-keeping
- Mechanized or motorized facilities.

Modernization and mechanization are not synonymous because many types and sizes of foundries cannot effectively use or economically justify mechanization. No foundry, today, can afford not to modernize and provide good working conditions under which the producers can produce maximum units with adequate return to labor, management and owners. Foundry modernization must be "tailored" to fit the particular plant under consideration. It is a continuing program to meet the increasing demands for higher quality, and lower costs, and to take advantage of technological developments. Every foundry can modernize and reduce its costs and improve its product.

#### DISCUSSION

*Chairman:* JAMES THOMSON, Continental Foundry & Machine Co., East Chicago, Ind.

*Co-Chairman:* E. W. BEACH, Campbell, Wyant & Cannon Foundry Co., Muskegon, Mich.

H. W. JOHNSON (*Written Discussion*):<sup>1</sup> Figure 7 shows the proposed side-dump skip-type charger for use with the No. 4 cupola. I believe it would be well to stress the importance of this method of charging being confined to cupolas smaller than

No. 6 cupolas. Our experience with side-dump skip-type charges on a No. 7 cupola has been most disappointing. Piling up of materials on one side of the cupola creates non-uniform melting conditions requiring excessive coke charges to produce high temperatures. A test made in a malleable foundry showed that at least 50 lb more coke per charge was required in a side-dump charger than from a cone-bottom type bucket.

What type of charging equipment is recommended for various size cupolas? What are advantages and disadvantages of various types of buckets?

W. R. JÄSCHKE:<sup>2</sup> The side-dump charger performs satisfactorily on cupolas up to and including the No. 5 cupola. Beyond the No. 5, it begins to be questionable in performance because in any end-dump device there is a difference in the momentum of material that is cascading from any end-dumping chute. Certain materials will go to one side of the cupolas and other materials with lower momentum will drop to the near side of the cupola. This is likely to cause channeling of the blast up one side and channeling of the limestone down the other side with resulting unequal melting conditions in the cupola. In order to correct the unequal melting conditions, a greater amount of coke is required.

The best way to charge a cupola, in my opinion, is by means of the cone-bottom bucket. This bucket places the material tightly around the periphery of the cupola and leaves the center loose and open, and promotes a more uniform travel of the gases up through the cupola. That is the way most of the old-timers taught us to charge a cupola by hand. Mechanically, it can be done best with a cone-bottom bucket. This, of course, cannot be accomplished with a cone-bottom bucket unless the bucket itself is properly loaded. We like to load cone-bottom buckets from some vertical type hopper, so the charge drops down and distributes itself uniformly. Whereas, if you chute-load a cone bucket, you are apt to have some unbalanced distribution of the charge unless the system permits you to rotate the bucket, say the first load in as the bucket sets and the next load in a position 180 degrees from that.

Lately another type of charger has become popular because of the low cost of material handling, and that is the inclined skip with a quick release bucket. From a mechanical and the material handling viewpoint, it appears satisfactory, but in quickly releasing material into the cupola, the first charges drop quite a height onto the bed resulting in tightly packing the coke bed and defeating the ideas you might have on penetration of blast and uniform gas travel through the cupola. It might be 3 to 3½ hr before the effect of that packing in the initial part of the heat burned itself out. After the cupola is charged to its normal height, of course, the impact of the charge is not so great and it tends to create a more uniform density but it still does not leave the center open and loose like a cone-bottom bucket does.

Some critical operators, who are getting good results go farther, and use a combination system. They use the cone-bottom bucket for the metal and the hinged-bottom bucket or some other type of drop bottom bucket for the coke and stone. In that way, they get a better distribution of the limestone than they can get with the cone-bottom bucket. They get all the good thermal efficiency of the cone-bottom bucket system but none of the exaggerated damage to the linings caused by the limestone from the cone-bottom bucket being deflected out toward the lining.

The inexpensive side-dump skip type charger is performing satisfactorily on small diameter cupolas but for the larger diameter cupolas the cone-bottom bucket systems have proven best. They usually call for some type of crane charger and are flexible enough to use the cone-bottom bucket or any other type of bucket or combination of buckets. Skip-type chargers are generally limited to the use of one style of bucket.

CO-CHAIRMAN BEACH: Would you care to state the minimum diameter for cone bucket use?

MR. JÄSCHKE: About 42 in. We can make them smaller.

CO-CHAIRMAN BEACH: I am asking for a recommendation.

MR. JÄSCHKE: We make that recommendation because the customer will not make the material small enough to get good passage from the bucket on diameters lower than that. If you remember some of our A.F.A. exhibits of years ago, we had a cone-bottom bucket operating. The bucket was 6 in. in diameter and the material was broken down proportionately in size. If

<sup>1</sup> Wells Manufacturing Co., Skokie, Ill.

<sup>2</sup> Whiting Corp., Harvey, Ill.

charges you did the same thing in commercial applications, you could use a cone bottom bucket down to any size cupola.

F. A. JENSEN:<sup>2</sup> Reference was made to a combination iron and brass foundry producing castings from 1/2 lb to about 3 1/2 tons. Is it possible to operate a foundry of this type with one type of sand?

MR. KNIGHT: Yes, if you will use the same base sand for the heavy class of work and a special facing for the light work. There are some claims by the chemically-coated sand people that it is not necessary to use the special facing.

MR. JENSEN: In a foundry of that type a basic sand should be applied to the small and medium castings and a special facing sand should be applied to the heavy castings.

CHAIRMAN THOMSON: Mr. Knight discussed illumination for foundries. One objection to putting in many lights in a foundry is the large number of units you must install for adequate illumination. They require a lot of maintenance.

Two of the large companies recently brought out for high bay mounting 3-kw tube lights with satisfactory results. Spacing those units about 40 ft apart both ways, you can light up a large area with very few units giving 25 to 30 ft-candles of light.

MR. KNIGHT: In some of the larger foundries they are putting in the combination mercury and filament lamp, high-level illumination, spot-type light. They are a sealed-beam light. But I was not thinking of that type of lighting so much as I was of lighting at the work level in the class of foundry that we were discussing. We would not attempt, in most cases, to establish a 20 to 40 ft-candle level of illumination throughout the plant. The expense would be uneconomical. But it should be that high at the work level. Generally speaking, if you clean up the walls, roof or ceiling of your plant and wash the windows and then provide additional illumination at the working stations, you can well afford it by improvement in quality and increase in production and reduction in scrap.

CHAIRMAN THOMSON: If you do not keep those lights clean you will not obtain required illumination.

MR. KNIGHT: The same applies to the wall surface and the windows. They must be kept clean.

CHAIRMAN THOMSON: The more windows there are the less chance of their being kept clean.

Two years ago the A.F.S. Plant and Plant Equipment Committee sponsored a Symposium on ventilation (Foundry Dust Control). It has since been published by A.F.S. and contains many ideas that are worthwhile.

C. C. WILSON:<sup>4</sup> What is the best method for getting the dust off the monorails, trusses and overhead work in a foundry?

MR. KNIGHT: I have seen a few portable vacuum cleaners used. Users of these vacuum cleaners speak favorably of them. That is the best method I have seen so far.

CO-CHAIRMAN BEACH: There are cleaners that take into their hoppers about 1,000 lb of dust at a time. They are not too expensive and work efficiently. The cleaner heads are sufficiently light so that they can be handled on a long pole to reach the lower chord of the trusses in the average low-tonnage foundry.

MEMBER: I would like to know of a means of controlling the slag wool that comes from the cupola. We have quite a problem with it in our shop.

MR. KNIGHT: I do not know whether there is a satisfactory way of completely controlling it. Front slagging virtually eliminates it, but with a short heat you probably would not want to resort to front slagging. In connection with back slagging I have seen a number of catchers rigged, some of which are more or less successful.

MR. JAECHKE: Front slagging can be applied to the cupola, even in very short runs of 1 1/4 hr. Front slagging is the cleanest way, besides giving the advantages of no loss of air through the slag hole and better control of cupola operations.

MR. KNIGHT: Do you recommend front slagging for that short a heat?

MR. JAECHKE: I would, yes. I like to run a front-slagging cupola. I like it better than the others, and think it can be worked on a cupola running 1 1/4 hr.

MR. KNIGHT: How about 1 hr?

MR. JAECHKE: I would not slag it at all. I would not even put any flux in it but would run the cupola without flux and without slagging. I have seen that done in a cupola for longer than

1-hr duration, melting as much as 22 tons a day, and the cupola dropped clean.

MR. KNIGHT: With today's materials?

MR. JAECHKE: That was before the war. There are water-sprayed collectors. Most of them I think are home-designed. Ford Motor used a shield in the back that was water-jacketed with the water discharging into the slag chute.

MR. KNIGHT: Did they still have slag wool?

MR. JAECHKE: Yes, but they did not have these jacketed shields on all their cupolas. This happened to be on one bank of cupolas. They had as many as 30 cupolas in that shop so the wool may have come from somewhere else. From this particular set-up of cupolas, I do not think much wool was produced.

In a little shop in Indiana I saw a funnel arrangement connected to an exhaustor with a water spray at the inlet to that exhaust. That also was a home-made affair. In some of Mr. Schneible's illustrations you will see another idea. Water spraying seems to enter into all of those ideas to minimize slag wool. The surest way is to front slag.

C. B. SCHNEIBLE:<sup>5</sup> A satisfactory arrangement consists of a water-cooled funnel with hollow center; the blast from the slag hole is drawn into that funnel and the dust goes up to a small collector.

J. E. QUEST:<sup>6</sup> Mr. Knight referred to a minimum daily tonnage required for complete mechanization of a small foundry. What would you recommend as the minimum daily tonnage for a charger?

MR. KNIGHT: I doubt that anyone today who is building a foundry would put in an elevator and a charging deck in preference to a charger, regardless of what the tonnage was to be.

MR. QUEST: In so many of the old buildings it is a matter of conversion difficulties and conversion costs.

MR. KNIGHT: The answer to that is, how many men do you have and how many men will be eliminated by its use? How much maintenance cost would be eliminated by elimination of cost of maintaining your present elevator?

MR. QUEST: In other words, there is no simple answer to the amount of tonnage required.

MR. KNIGHT: I do not think the tonnage has anything to do with it. I think it is a question of man-hours.

MEMBER: We have a cupola in the center of the foundry. To run a charger in, involves an overhead carry of about 30 ft. Is that practicable for a charging system?

MR. KNIGHT: Under some conditions, yes, very definitely.

MEMBER: Then in regard to the pouring distance and pouring crews, when you are running a foundry of about 12 tons do you not run into a problem of trying to get your melting rate down so you do not exceed the capacity of the pouring crew?

MR. KNIGHT: That is right. You either have to get the melting rate down or the pouring crew up.

MEMBER: What would be the minimum rate that you could melt and maintain a pouring crew efficiently?

MR. KNIGHT: That would depend on the amount of metal per mold to a large extent. In the example that I gave, we had eight men pouring and eight men shifting 4 hr, plus two bull-ladle men to keep them going, but during the other 4 hr, they hauled castings to the cleaning room, cut sand over and did other laboring work around the foundry, where we otherwise would have additional people. It is a question of setting up a manual of work to be done and using those people for the jobs that have to be done.

MR. QUEST: Does the foundry in this case incorporate mechanical sand-handling and sand-processing equipment?

MR. KNIGHT: There was none in that foundry.

MR. QUEST: Was the shakeout operation eventually taken care of in the daytime by that crew?

MR. KNIGHT: The shifters actually did the dumping.

MR. QUEST: And the shakeout?

MR. KNIGHT: The castings were picked out of the sand and placed at the end of the floor by the shifters.

MR. QUEST: Was there any night work at all?

MR. KNIGHT: The pourers came in about 1/2 hr before they were supposed to start pouring and they worked for 3 1/2 hr after the molders quit so that for all practical purposes by 7 pm the work was done.

<sup>2</sup> National Engineering Co., Detroit

<sup>4</sup> Enterprise Wheel & Car Corp., Bristol, Va.

<sup>5</sup> Claude B. Schneible Co., Detroit

<sup>6</sup> J. F. Quest Foundry Co., Minneapolis



MEMBER: What did you do about your mold gas problem when you were pouring off? When pouring 4 hr while your men are molding do you not have a mold gas problem?

MR. KNIGHT: We never pour in the molding area. The men use portable machines and work back on the heap. The foundry does not pour until after there are 4 hr of mold on the floor. In the 4 hr there are almost as many molds on the floor as there used to be in 6 hr because the molders do not have to pour them.

MR. JAESCHKE: Is it not true that when you stretch out the pouring period and pour at a slower rate, the natural ventilation of the building will take care of the gases much better than when you pour off a large heat in a short period of time?

MR. KNIGHT: That is true. There is no comparison between the conditions you have when you pour off in 1 or 1½ hr as compared with pouring off over a 4 or 8-hr period. There is no interference with the molder, because the pouring is at least 20 to 30 ft away from the molder.

MR. QUEST: Is that a minimum distance?

MR. KNIGHT: No, but that happens to be the case in the shops where this system is practiced.

W. R. TUTTILL:<sup>7</sup> In the malleable shop mentioned by the author where there were 36 molders I think he said they cooled the molds after pouring. They put them on a conveyor to go at right angles to the shakeout. How did they handle the excess sand going to the molder hoppers above? Did that have a chute and was it handled on the same conveyor or disposed of separately?

MR. KNIGHT: A chute delivered the overflow to the return conveyor.

C. B. JUDAY:<sup>8</sup> Referring to Mr. Jaeschke's problem on wool from the slag spout, we solved this problem by putting a hood over the slag spout and running a 27-in. diam exhaust pipe from this hood vertically up the side of the cupola and into the cupola stack on a 45-degree angle, 5 ft above the charge door. We found that this arrangement removed not only the rock wool, but also the gas from the slag hole. We had previously had numerous complaints from gas.

MR. JAESCHKE: Where does it go after leaving the cupola stack?

MEMBER: Did you have a water-cooled hood on that?

MR. JUDAY: The wool seems to disintegrate in the stack because we have never seen any wool on the ground or in the air. We did not have a water-cooled hood.

R. S. FORBES:<sup>9</sup> Is it feasible to put in floor ventilation where you intend to pour and mold at the same time, using a system such as the one discussed, whereby the molders are molding and the pouring is going on at the same time? Is it practicable to successfully aerate at the floor level, so it would be possible to

pour and mold for the whole 8 hr continuously? Is there an inexpensive way of doing that in an old foundry?

MR. KNIGHT: Inexpensive is a relative term. How much would you be willing to spend to improve your working conditions? How far do you want to improve them? The ventilation problem is one that I think is most mistreated. It is one on which a good engineering job for foundries has rarely been done. It is one on which an untold sum of money can be spent.

MR. SCHNEIBLE: On this immediate problem, in this short-distance-of-pouring zone, it should be possible to do an inexpensive job of ventilation at such a station.

MR. KNIGHT: That is not a station. That is poured through out the whole foundry.

MR. SCHNEIBLE: Let us consider it as a whole foundry with roof monitors. The trouble with most installations in the past has been that there has been no way of letting air into the building so that the air could get out of the monitor. In the summer when the windows are open there is a natural thermostatic stack action. In the winter, however, when you close all the windows there is no way of admitting fresh air into the building to allow the hot, smoky air to get out. Your problem will have to be studied from the standpoint of bringing in air from the outside, undoubtedly heating it and admitting it in proper quantity. I know of no other way of doing it. There have been some applications where the warm air from the top of the foundry has been pumped down when it was not too smoky up above, but your winter operation is always bad in the monitor-controlled foundry. The primary part of the study would be in the admission by an air pump or fan, so you would deliver a regulated amount of air into the foundry near the work floor to replace that which will go out through the monitor.

MR. KNIGHT: In one shop that poured about 18 tons a day, we installed a series of unit heaters and brought in approximately 40 per cent of the replacement air necessary and heated that and delivered it at certain control points around the foundry and allowed 60 per cent of the exhaust air to come from infiltration. Power ventilators in the roof moved a given volume of air, 40 per cent of which was replaced by heated air drawn from the outside through ventilators. In the summer time, it is drawn through filters with no heat. In the winter time it is heated. That has permitted 4-hr pouring in a shop where they did have a high percentage of cores and where concentration of core smoke was a problem not for the molders but for the pourers.

MR. SCHNEIBLE: Do you not agree that a proper set-up from an engineering standpoint would be to pump a definite amount of air out of the building with power roof ventilators and to blow in a definite volume of air, heated in winter, not heated in summer, less building leakage? Such a plant could be engineered and would be a feasible way of keeping the shop clean.

<sup>7</sup> American Seating Co., Grand Rapids, Mich.

<sup>8</sup> Perfect Circle Co., New Castle, Ind.

<sup>9</sup> Taylor-Forbes, Ltd., Guelph, Canada

# THE DEVELOPMENT OF A PERMANENT MOLD FOR ALUMINUM TENSILE TEST BARS<sup>1</sup>

By

L. J. Ebert<sup>2</sup>, R. E. Spear<sup>2</sup>, and G. Sachs<sup>2</sup>

## ABSTRACT

A permanent mold has been developed for the casting of aluminum alloy test bars with a short time operating cycle. The two-bar mold cavity was designed in such a manner that sound castings can be obtained with the pouring of a small amount of metal, and with a minimum of trimming to remove the gates and risers.

The factors which influence the mold operation and test bar properties have been examined by means of the casting of five different aluminum alloys. The pouring temperature was found to affect the mold operating cycle only slightly, but the mold temperature had a great effect upon the cycle. The mold temperature was also found to influence the tensile and hardness properties. Heat treated alloys were especially responsive to mold temperature change. The thickness of the mold coat, the time of the casting in the mold, and the rate of pouring the casting were found to affect the tensile properties only under extreme conditions.

## Introduction

THE RISING IMPORTANCE of permanent molds in the aluminum foundry industry has brought about an increasing need for information on the operation and design of permanent molds. Because of the unique nature of each permanent mold, it is difficult to establish definite rules concerning them. For this reason there is little data available on the effects of operational variables. This is especially true for test bar molds.

In the design and operation of a permanent mold, a number of factors and variables must be considered. Of prime importance is the production of a sound casting; all other factors must be subordinate to this. At the same time, it is highly desirable that the trimming of the casting be kept to a minimum and that the amount of metal (scrap) trimmed off be small. It is likewise beneficial to retain the cast surface on the trimmed casting to make use of the higher properties of the chilled "skin".

Concerning the operation of the mold, it is advisable to keep the operating cycle short in order to

increase the productivity of the mold and to utilize the heat transferred from the casting to the mold. In line with this reasoning, any mechanical aids in the mechanics of the operation will help to minimize the operating cycle.

In addition, it is advantageous if the cost of the mold can be kept to a minimum. This means elimination of all die-sinking machining operations and replacing them with simple milling, drilling and reaming operations.

Attempts to obtain a permanent mold which embodies a correct balance among all of these factors are presented in this paper. The development of the mold discussed necessitated the investigation of all pertinent variables.

The data contained herein may be applied specifically only to the mold discussed. However, the general trends of the variables considered and investigated should be the same for a large number of the permanent molds employed for the casting of aluminum alloys.

## Design of the Mold

### Material

The design of the mold used in this investigation is shown in Fig. 1. The material used for the mold block and hinges was cast meehanite. This was in accord with the commercial practice of choosing a mold material of any good grade, small grained cast iron or steel.

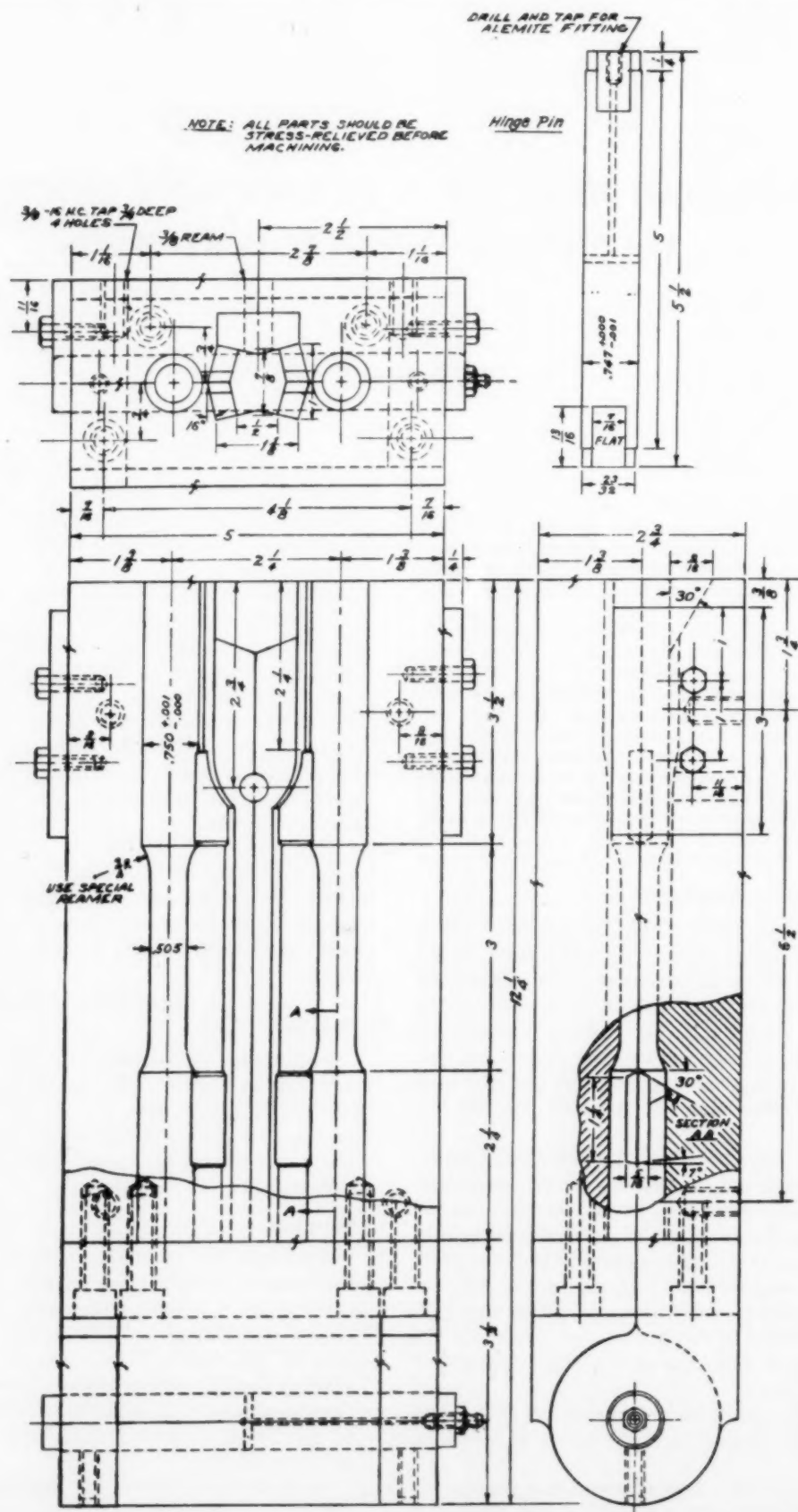
### Hinging

Although the mold shown in Fig. 1 was hinged at the bottom, a slightly less expensive mold, having the same cavity, was made with a side hinge, Fig. 2. In the case of side hinging, the hinge was cast as an integral part of the mold. The main disadvantage of side hinging lied in the possibility of wearing in the hinge, thus producing a misalignment of the mold. Wearing of the hinge of the bottom-hinged mold would have no such effect.

In order to forestall the possibility of wearing in the side hinge, short dowel pins with rounded ends were added to the mold to produce positive alignment. As an added alignment precaution for the bottom hinged mold, close fitting guide plates were

<sup>1</sup> This paper is one of a series of reports in a research program on cast aluminum alloys conducted at Case Institute of Technology for the Aluminum Research Institute, Chicago, Illinois.

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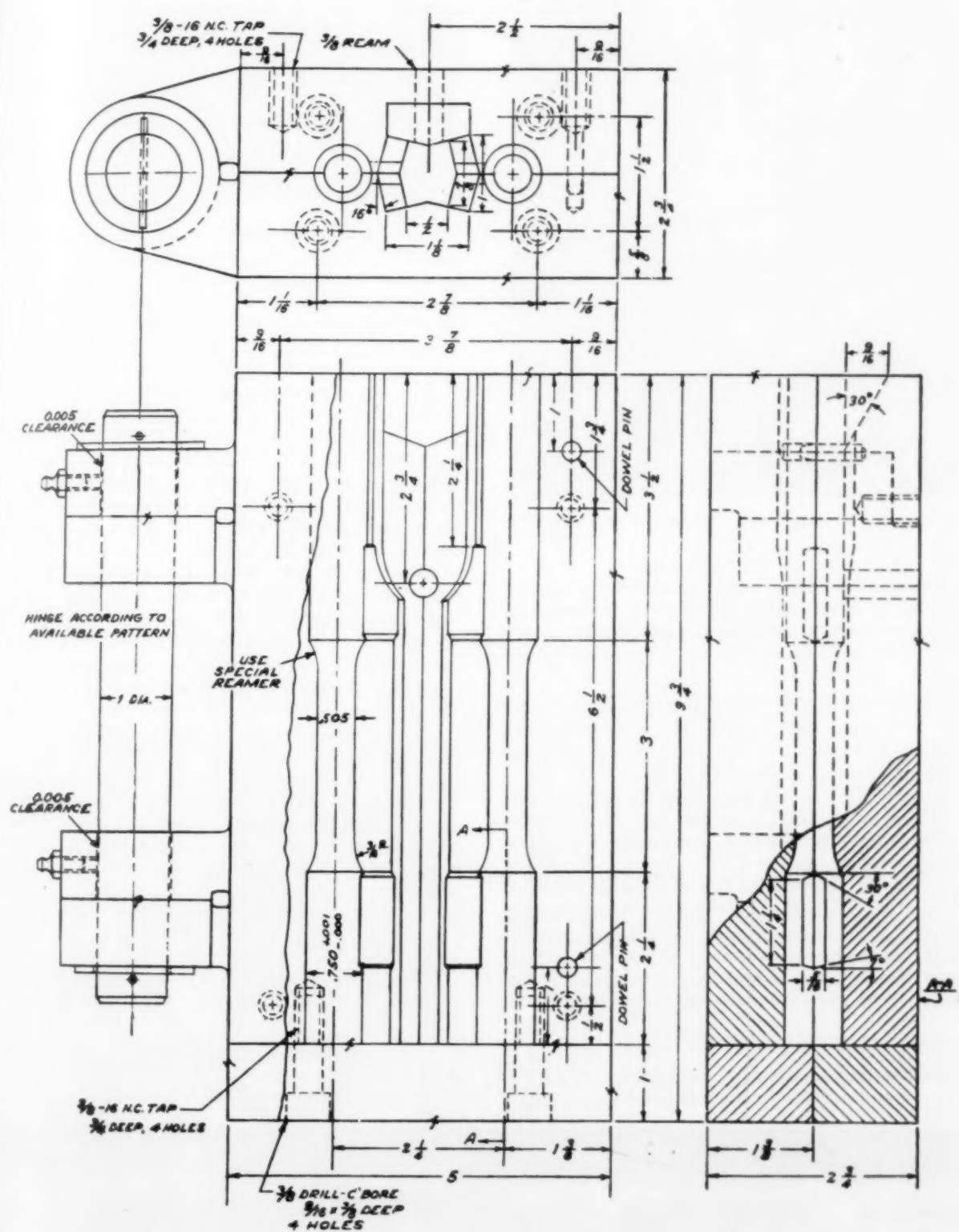


Fig. 2—Side-hinged tensile bar permanent mold.

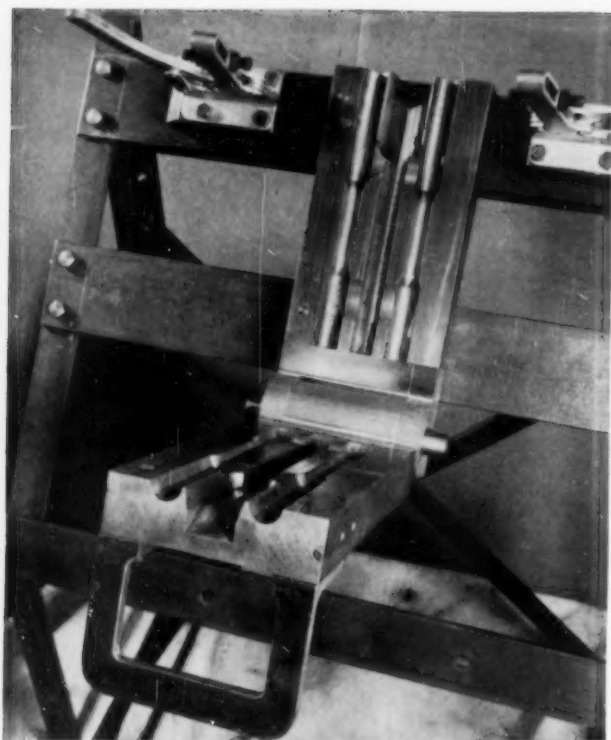


Fig. 3—View of mold cavity without mold coating.

added to the top of the mold. No difference was observed in the performance of the two molds, nor in the properties of the resultant test bar castings.

#### Gating

As shown in Fig. 1, the mold contains a center sprue. Four gates were so located that they spanned the reduced section of the test bars, but in no place intersected the test section of the bars. The upper end of each test bar was made long, in order to provide adequate feeding from the top of the mold. Figure 3 shows a closeup view of the mold cavity prior to coating.

The size and location of the sprue with regard to the test bars represents a compromise among several conflicting factors. In the first place, it was desirable to make the sprue cross section as large as possible to provide sufficient molten metal to feed the test bar cavity. However, if the sprue were too large, the

reservoir of hot metal in the sprue would produce a slow cooling rate in the side of the test bar nearest the sprue, and so yield bars which were not uniform in the hardness of their cross section. Another limitation which was to be considered was that the gates should not be so long that the molten metal could cool down in passing through them. The final design shown is the compromise among these factors, and consists of a fairly small sprue with reasonably short gates. Adequate feeding can be maintained by proper insulation of the sprue cavity, as will be discussed later.

Originally, the mold was built with gates of wedge-shaped cross section. Since the casting tended to shrink about its horizontal centerline, it was thought that gates of this design might utilize this shrinking to loosen the casting from the mold without damage to the mold coating. However, when the mold was put into operation it was found that shrinks appeared in the lower heads of each bar, indicating that the lower end of the casting was improperly fed.

To correct this fault, a heavier mold coating was added to the lower gates and the lower section of the sprue. This eliminated the visible shrink, but radiographs showed that a small amount of shrinkage remained in both the upper and lower heads. Figure 4 shows a casting before the extra coating was added to the wedge gates, and Fig. 5 shows a casting after the extra coating was added.

To eliminate the shrinkage completely, the gates were enlarged by making them rectangular. The outside dimensions of the rectangular gates were the same as those of the previous wedge gate. A draft angle of seven degrees was placed at the edges of the gate to facilitate operation of the mold. However, even when graphite was added to the edges, the mold opened with difficulty and the casting sometimes cracked or chipped the mold coating. This difficulty was eliminated by increasing the angle on the inside gate edges from seven degrees to 30 degrees. Figure 6 shows a casting with the rectangular gates. As seen in Fig. 6, the shrinkage has been visually eliminated by the rectangular gates. Radiographs also showed no internal shrinkage.

#### Test Bar Cavity

The test bar cavity was cut to produce bars according to the ASTM Standards. The design of the mold

Fig. 4—A casting before extra mold coating was added to wedge gates of the mold.

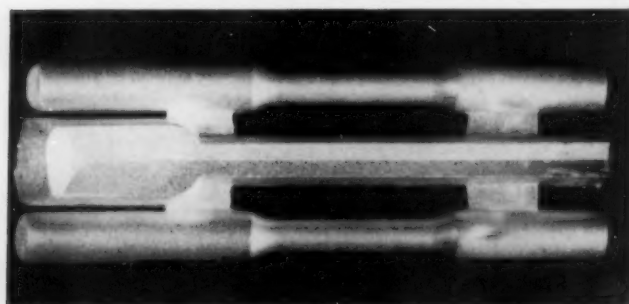
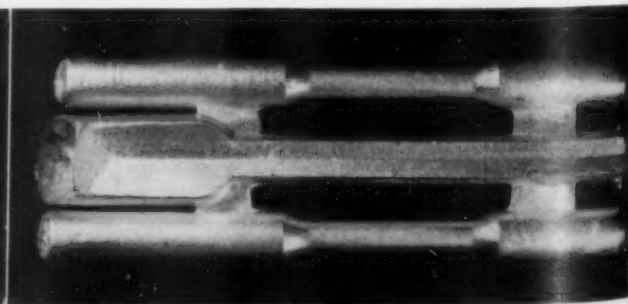


Fig. 5—A casting after extra mold coating was added to wedge gates of the mold.



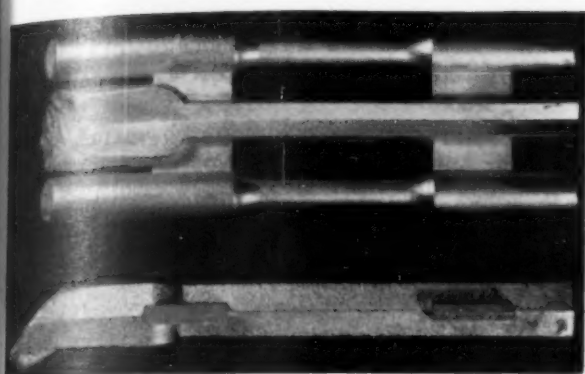


Fig. 6—A casting after mold was changed to rectangular gates.

permitted this cavity, as well as all others, to be cut by simple machining operations.

### Design of Mold Accessories

#### Casting Ejection

In operating the mold it was found advisable to add a simple ejector system to aid in the removal of the casting from the mold. The ejector is shown in Fig. 7 mounted on the mold. The ejector was primarily a simple lever system with a mechanical advantage large enough to remove the casting from the mold with a slight pedal pressure. It will be noted, Fig. 7, that the ejection was accomplished by a single

Fig. 7—Back view of mold showing the ejector mounted.



pin located slightly above the centerline of the upper gate. The location of the ejector pin provided a simple check on the removal of a casting which was not completely solid. The pin had a tendency to bend or break the sprue instead of ejecting a casting of which the sprue was still "mushy".\* Therefore, if the casting could be ejected without injury to the sprue, it was certain that the casting had completely solidified.

#### Mold Frame

The angle of mount of the mold frame (30 degrees) was determined experimentally so that any splashing at the mold bottom would be avoided, and yet so that maximum mold filling was provided.

Two toggle clamps, each capable of exerting 350 lb. force, were used to hold the mold together. The position of the clamps is shown in Fig. 3.

#### Thermal Measurements

The temperature of the mold was determined by the use of a chromel-alumel thermocouple located directly behind the center of the reduced section of the left test bar cavity, and half way between the bottom of the cavity and the back of the mold. All temperature readings were made on an L and N potentiometer.

#### Preheater

Since the mold was operated at elevated temperatures, the need existed for establishing an initial mold temperature of the approximate value which was to be used in the mold operation. This temperature was obtained easily with the aid of a simple preheater

\* It was found experimentally that the top of the sprue was the last part of the casting to solidify.

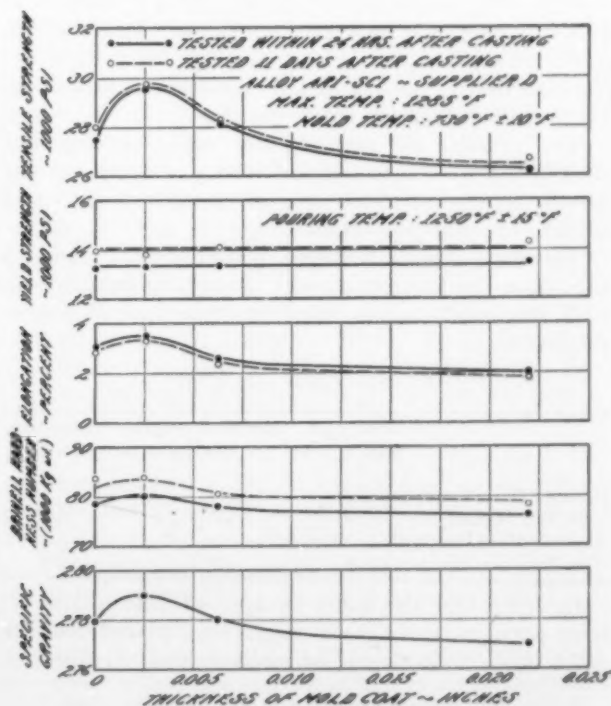


Fig. 8—Effect of thickness of mold coating on properties of alloy ARI-SC1.



which consisted of a perforated cast iron pipe closed at one end and connected to an air and gas supply on the other. The preheater was made of  $\frac{3}{4}$ -in. pipe with three  $\frac{1}{4}$ -in. holes along the side. The holes were located so that when the outer two were placed directly over the two test bar cavities, the center hole would be directly over the sprue.

The preheater was used at the top of the mold. The mold was opened enough to allow the preheater to fit between the two halves, and the flame was directed on the inner faces of the mold. The flame was kept neutral so that no carbon residue was deposited on the mold face, nor was the mold face oxidized.

### Mold Operation

#### Mold Coating Application

To maintain consistency in procedure, a single composition of mold coating was used in this work. The coating chosen was a proprietary aqueous suspension of china clay and iron oxide, which had been found to give good results in commercial applications.

The coating was applied with a spray gun using a pressure of about 20 psi gage pressure. In applying the coating, it was found preferable to lay down a number of thin coats rather than a single thick coat. The coating was sprayed on while the mold was at a temperature between 400 and 600 F.

The mold coating on the sprue and gates was sprayed on to a thickness of about  $\frac{1}{16}$  in. The heads of the test bars were coated to a thickness of about  $\frac{1}{32}$  in., while the reduced sections of the test bars were coated to a thickness of about 0.005 in. This procedure used the insulating property of the coating to help control the initiation and direction of freezing of the metal, so that freezing began at the reduced section and progressed to the sprue. To establish a definite place of initial freezing, the mold coat at the center of the reduced section was tapered so that the thinnest coating (0.003 in.) was present at the center of the reduced section of the test bar.

The practice of tapering the mold coating was established after it was found that a uniform coating in this section resulted in small surface shrinks and slightly lowered properties.

The data from a few tests (listed below) made on bars from the same heat, with a uniform coating and with a tapered coating, show the difference in tensile properties.

#### ARI-SC1 Alloy—As Cast\*

	Yield Strength, 1000 psi	Tensile Strength, 1000 psi	Elongation, Per Cent
Uniform Mold Coating in Test Section	12.0	21.0	1.5
Tapered Mold Coating in Test Section	13.2	24.2	2.5

In order that the spray be applied only to the desired portions of the mold, masks were placed over the mold surfaces on which no mold coating was desired.

\* The alloy designations given in this paper are those established by the ASTM for aluminum permanent mold alloys. The letters 'ARI' (Aluminum Research Institute) are used as a prefix to indicate that the alloys also meet ARI standards.

After the mold had been sprayed, all excess mold coating was scraped off the parting surfaces of the mold to insure proper closing contact, and thermal transmission across the parting surfaces.

To facilitate removal of the casting from the mold, graphite in the form of "aqua-dag" was sprayed over the mold coating on the gate cavities of the mold.

#### Mold Coating Thickness

The mold coat has as its functions the protection of the mold; the control of the surface conditions of the casting; and partially, the chilling effect of the mold. The thickness of the mold coat is primarily concerned with its effect on the chilling capacity of the mold.

In investigating the effect of mold coat thickness, the mold coating was varied only at the reduced section of the test bars. The remainder of the mold was coated as described above, and maintained constant. Test bars were cast into the mold for coatings at the reduced section from zero to 0.022 in. thick. The thickness of the coat was determined by measuring the bar radius and subtracting that value from the radius of a bar cast when no mold coating was used.

Figure 8 shows the effect of mold coat thickness on the properties of alloy ARI-SC1 in two conditions. With the exception of the yield strength, all properties reached a maximum at about 0.0025 in. of coating. The variation in the specific gravity in Fig. 8 explains the change in properties. Since the yield strength is not affected by small changes in density, it is to be

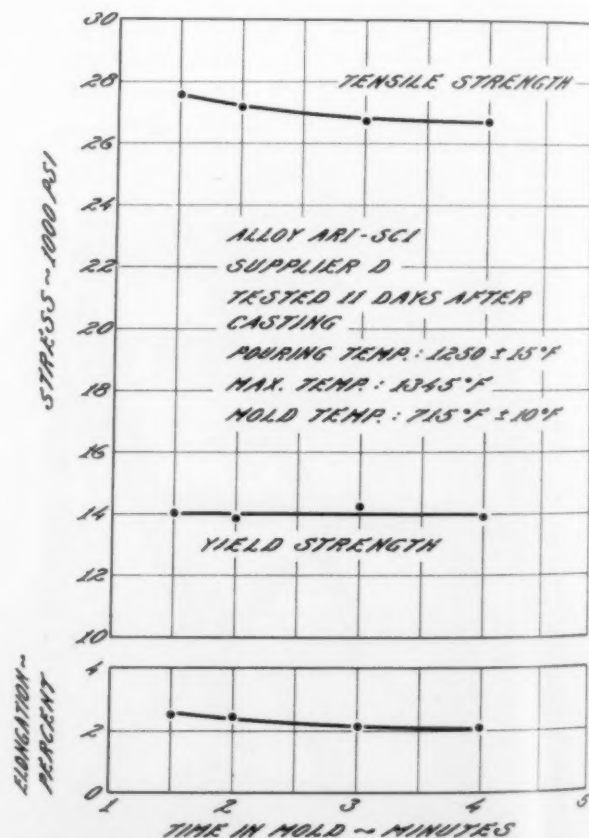


Fig. 9—Effect of time of casting in the mold on tensile properties of alloy ARI-SC1.

expected that no change in yield strength within this range of mold coating thickness would exist.

It should be noted that, although a considerable change in properties for the normal variation in thickness, there is only a slight change in properties for the normal variation in thickness which would be expected in commercial operation. In the operation of this mold, it was found that the maximum variation in thickness was from about 0.001 to 0.005 in. This would mean a variation in test bar diameter from 0.500 to 0.492 in. The variation in tensile strength for this much change in mold coat is only about three per cent, Fig. 8.

#### Time in the Mold

The length of time that a casting is left in a mold influences the physical properties of the casting proportionally to the effect it has upon the cooling rate of the casting. The longer the casting is left in the mold after solidification, the slower is the cooling.

The effect of the time in the mold on the tensile properties of alloy ARI-SC1 was determined by allowing the casting to remain in the mold for several different times. The minimum time in this investigation was limited by the necessity for a solid casting.

Figure 9 shows the effect on the tensile properties of the time of casting in the mold. The tensile

strength increased about three per cent for a decrease in time in the mold from 3 to 1.5 minutes. An increase of time in the mold from three to four minutes had no effect upon the properties. The yield strength was practically unaffected by the variation in time in the mold.

It is to be expected that the properties would increase slightly for shorter times in the mold from a consideration of the cooling rate. When the casting is removed from the mold it is allowed to cool considerably faster than when in the mold. During the time when the casting is at a high temperature, this effect of added chilling rate is measurable as shown, Fig. 9, at times from 1.5 to 3 minutes.

#### Pouring Rate

The rate at which metal is poured into the mold influences the turbulence of the liquid metal as well as the temperature at which it reaches the casting proper. These two factors tend to oppose one another since a fast rate of pour results in turbulence but a more suitable casting temperature, while a slow rate of pour results in viscous flow but a lowering of the metal temperature to a point that it might freeze before filling the mold.

To determine the effect of the rate of pour upon the castings' properties, alloys ARI-SC1 and ARI-CS4 were poured through nozzles with various hold diameters from  $\frac{1}{8}$  to  $\frac{5}{8}$  in. The time required to fill the mold using these various nozzles was measured and the castings were weighed. Then the pouring rates in pounds per second were calculated. The largest pouring rate was chosen well within the maximum possible rate allowed by the narrowest portion of the feeding system. Figure 10 shows the change in tensile properties of ARI-SC1 with a variation in pouring rate for the "as cast" condition of tensile bars tested within 24 hours and after two weeks.

From Fig. 10 it can be seen that the tensile strength reached a maximum at a pouring rate of about 0.20 pound per second. For pouring rates above this value it was noted that the castings lost some of their smooth surface. At high pouring rates (1.5 seconds to fill the mold) an obvious wrinkling of the skin resulted. All bars having this disturbed skin effect had tensile strengths which were slightly lower than smooth castings.

Bars poured at the slowest rate, in Fig. 10, also showed by tensile strengths. This was attributed to the fact that at slow pouring rates the test bars contained cold shuts in the reduced section, indicating that the rate of pour was so slow that solidification began before the test bar cavity was completely filled.

The elongation shown in Fig. 10 follows the tensile strength, but the yield strength was unaffected by the pouring rate. Since it has been well established that small defects have no effect on the yield strength, it would be expected that the yield strength would not be affected by the pouring rate.

Figure 11 shows the effect of the pouring rate upon the tensile properties of alloy ARI-CS4 in the solution heat treated and aged condition. These curves indicate that, for alloy ARI-CS4, a fast pouring rate does not reduce the tensile properties, but a very slow rate

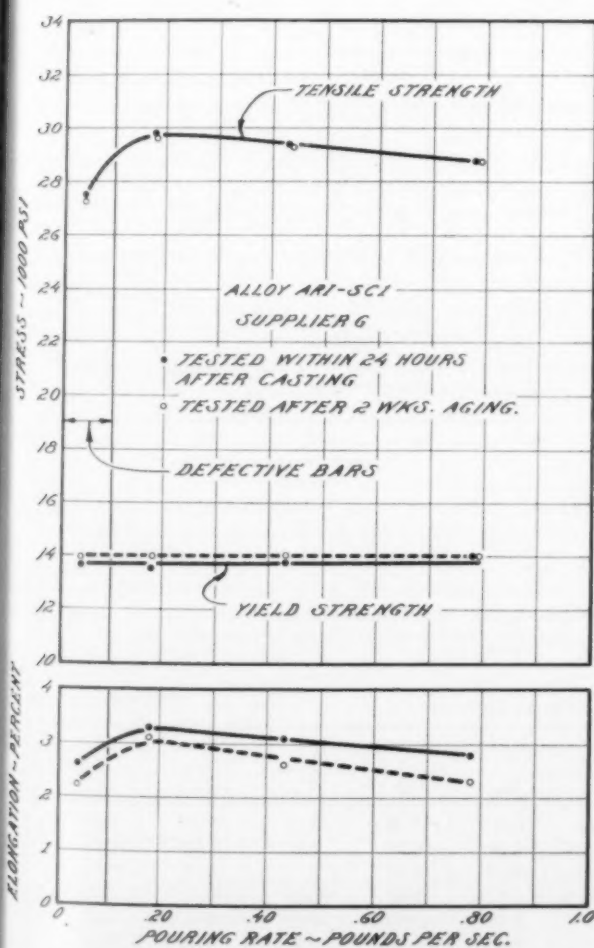


Fig. 10—Effect of pouring rate on tensile properties of alloy ARI-SC1.

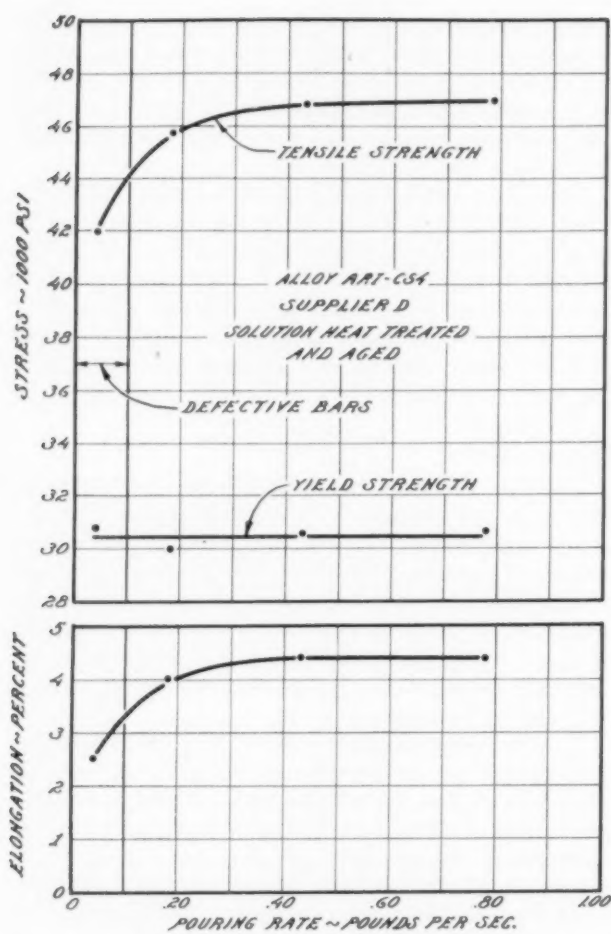


Fig. 11—Effect of pouring rate on tensile properties of alloy ARI-CS4 in the solution heat treated and aged condition.

is quite harmful to the tensile strength and elongation. It might be mentioned that the bars containing shrinks were not rejected unless the fracture took place through the shrunken portion of the bar. Thus, the values given in Fig. 10 and 11 do not show the effect of the pouring rate on the number of defects or the magnitude of the defects.

#### Mold Operating Temperature

In the operation of a permanent mold, a mold operating temperature, or operating cycle\* must be established to maintain consistent results. Extensive study has been made on the relationship between

mold temperature and operating cycle for the permanent mold under discussion.

The alloys, pouring temperature ranges, and the maximum melting temperature used in this study are given below:

ARI Alloy	Pouring Temp.	Max. Temp.
SC1, CS22, S2	1250 +25 F	1350 F
CS4, CG1	1275 +25 F	1350 F

The chemical compositions of the alloys used are given in Table 1.

**Operating Cycle**—The mold operating cycle is composed of three parts: the time for the casting to solidify completely, the time required in handling the casting, and the time for the mold to return to its operating temperature. Since the handling time is determined only by the minimum time required for the mechanics of removing the casting from the mold, the operating cycle varies according to the effect of

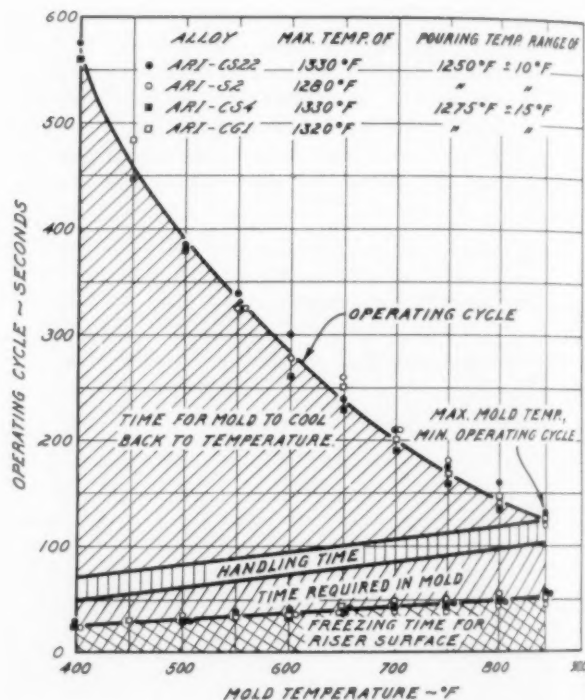


Fig. 12—Effect of mold temperature on operating cycle for different permanent mold alloys.

\* The operating cycle is defined as the time between the beginning of two succeeding pours. This cycle is dependent primarily upon the mold operating temperature.

TABLE 1—CHEMICAL COMPOSITION

Composition—Per Cent by Weight												
Alloy	Supplier	Cu	Si	Fe	Mg	Ni	Mn	Sn	Pb	Zn	Cr	Ti
ARI-SC1	D	4.46	5.71	0.71	0.04	0.19	0.40	0.10	0.15	0.27	0.08	0.06
	I	4.70	5.78	0.78	0.02	0.30	0.32	0.03	0.06	0.30	0.03	0.04
	G	4.67	5.66	0.73	0.04	0.24	0.38	0.03	0.05	0.13	0.04	0.06
ARI-CS22	D	7.03	3.01	1.03	< 0.01	0.33	0.50	0.07	0.31	1.70	0.05	0.11
ARI-S2	G	0.58	5.07	0.65	0.03	0.02	0.42	0.05	0.02	0.03	0.08	0.03
ARI-CS4	D	4.41	2.45	0.89	< 0.01	0.05	0.45	< 0.01	0.01	0.12	0.10	0.11
ARI-CG1	G	4.63	2.45	0.80	0.03	0.10	0.40	none	none	0.20	0.02	0.03
	L	10.36	0.61	1.11	0.27	0.19	0.29	trace	0.02	0.11	0.08	0.04



the mold temperature on the casting solidification time and the time required for the mold to return to the operating temperature.

For this investigation, the time required for a completely solid casting was established empirically as twice the time required for the riser surface to freeze. By allowing the casting to remain in the mold for this time, it was possible to obtain completely solid castings in the minimum time. This resulted in the casting being in the mold for a slightly shorter time for a low mold temperature than for a high mold temperature, Fig. 12, because of the increased chilling effect of the colder mold.

The time required for the mold to return to the operating temperature, of course, varied greatly with the magnitude of the operating temperature, Fig. 12. This time was governed by the radiation losses from the mold surfaces.

The alloy composition had no effect upon the mold operating cycle as seen in Fig. 12. This was to be expected since all alloys were poured within narrow limits of temperature. The mold operating cycle may be influenced by the alloy composition in three ways:

1. The change in pouring temperature necessary to produce sound castings;
2. The differences in the heats of fusion and the heat capacities;
3. The differences in the thermal conductivities.

The consistency of the mold operating cycle reported in Fig. 12 shows that the latter two factors are of no importance for the alloys under consideration in this report. The implication of this fact is that the heats of fusion, heat capacities, and thermal conductivities of the various alloys do not differ enough to cause a change in the operating cycle.

The maximum mold operating temperature was approximately 850 F, Fig. 12, because the time necessary to allow the casting to solidify completely in addition to the handling time was sufficiently long that the mold temperature never rose above this value.

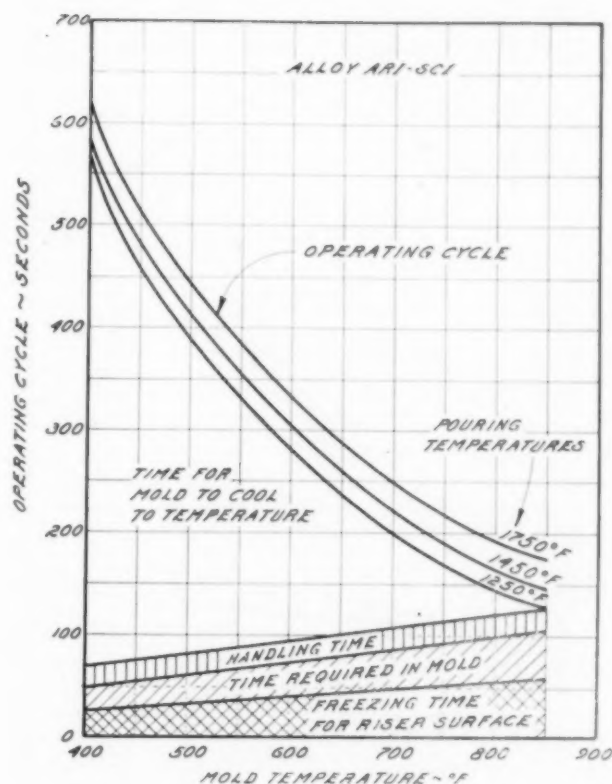


Fig. 14—Effect of pouring temperature and mold temperature on operating cycle.

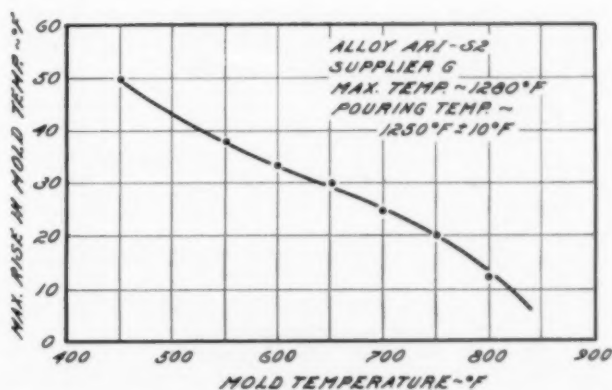


Fig. 15—Rise in mold temperature at various mold operating temperatures.

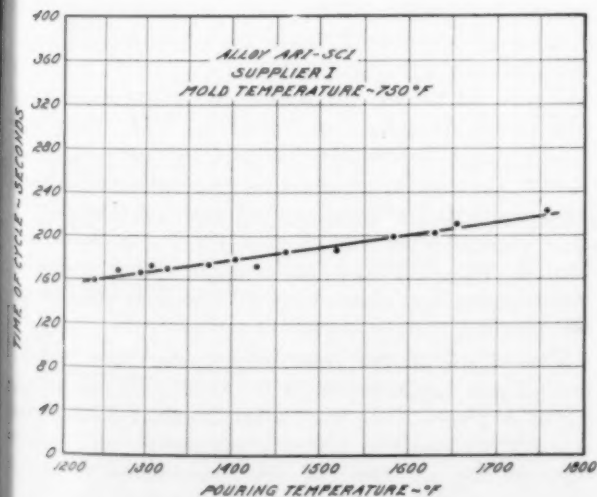


Fig. 13—Effect of pouring temperature of the metal on length of cycle if a constant mold temperature is maintained.

Since no external heat was applied while in operation, the maximum operating temperature was the temperature at which the rise in mold temperature was just sufficient to allow for the radiation losses which took place when the casting was removed from the mold. This was also the minimum operating cycle.

The effect of the pouring temperature upon the operating cycle was not measurable for a change in pouring temperature within the specified ranges (1250 to 1300 F). However, for a wide change in pouring temperature a considerable variation was noted, Fig. 13. To check the effect of the pouring temperature on the operating cycle, the mold was maintained at a constant temperature, while the pouring temperature

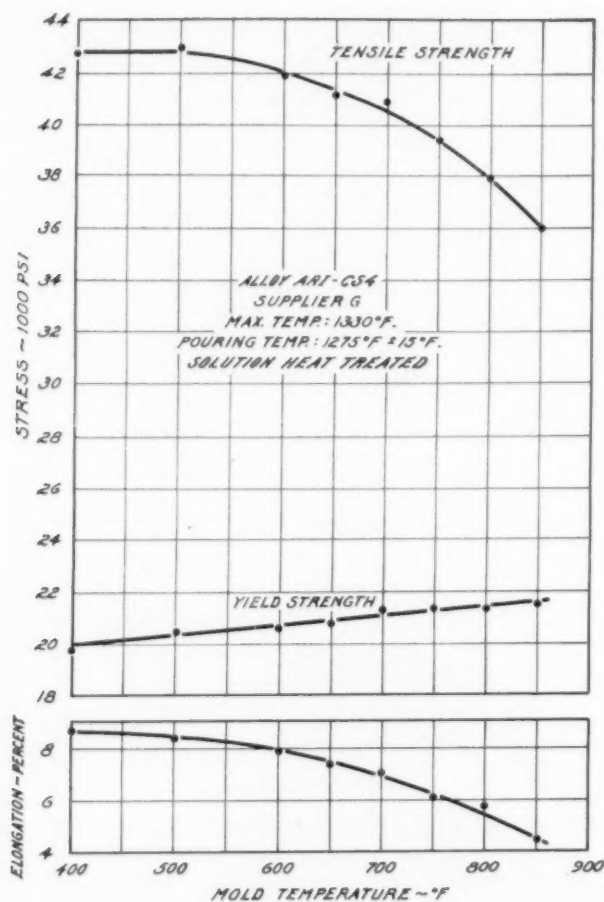


Fig. 16—Effect of mold temperature on tensile properties of ARI-CS4 permanent mold alloy in the solution heat treated condition.

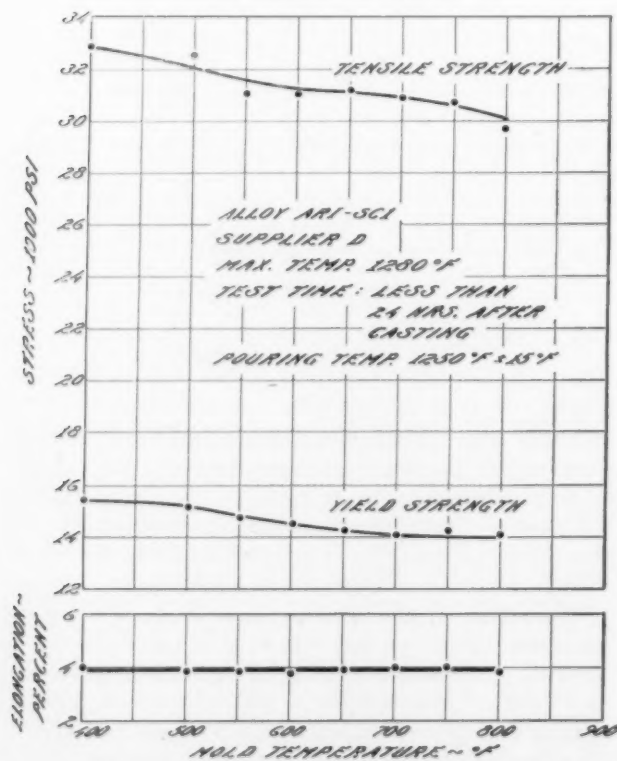


Fig. 18—Effect of mold temperature on tensile properties of ARI-SCI permanent mold alloy.

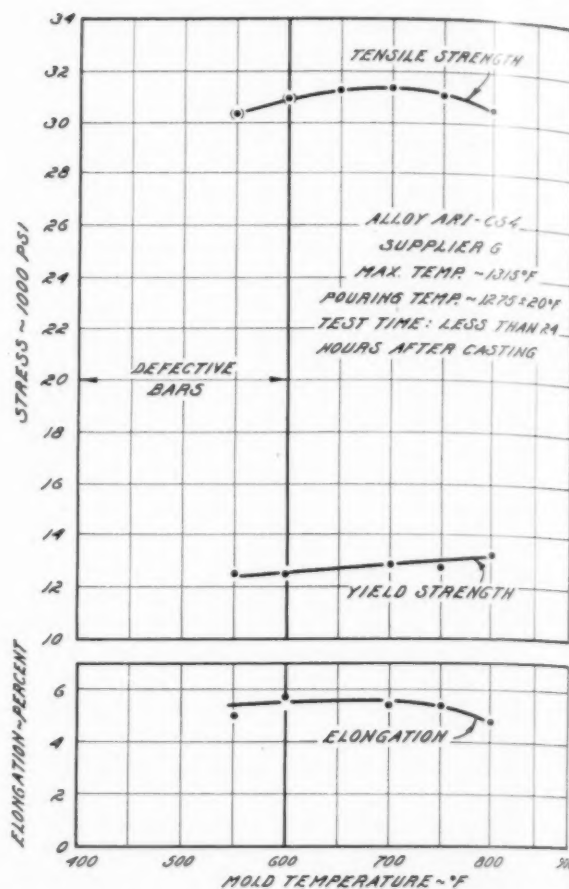


Fig. 17—Effect of mold temperature on tensile properties of ARI-CS4 alloy in the as-cast condition.

for a single alloy was varied from 1240 to 1760 F. It will be noted from Fig. 13 that the pouring temperature has a linear effect upon the operating cycle. The slope of the line is 0.114, which means that for an increase in pouring temperature of one degree an increase in the operating cycle of 0.114 second is necessary to maintain the same operating cycle. Therefore, it is to be expected that within the limits of approved foundry practice only slight changes in the operating cycle would result from the normal variations in pouring temperature.

Figure 14 shows the variation in operating cycle with three pouring temperatures, 1250, 1450, and 1750 F. Even at the extreme pouring temperature of 1750 F the operating cycle is not displaced greatly from those of the more normal pouring temperature. It is evident from Fig. 14 that the major factor controlling the operating cycle is the mold operating temperature and that it is only slightly affected by the pouring temperature.

The shape of the operating cycle curve is determined to a large extent by the time required for the mold to dissipate the heat added by the molten metal. A mold operated at a low temperature will transfer heat to the air more slowly, receive heat from the metal more quickly, and will absorb more heat from the metal than the same mold operated at a higher temperature. All of these factors tend to increase the time required for the mold to return to the operating

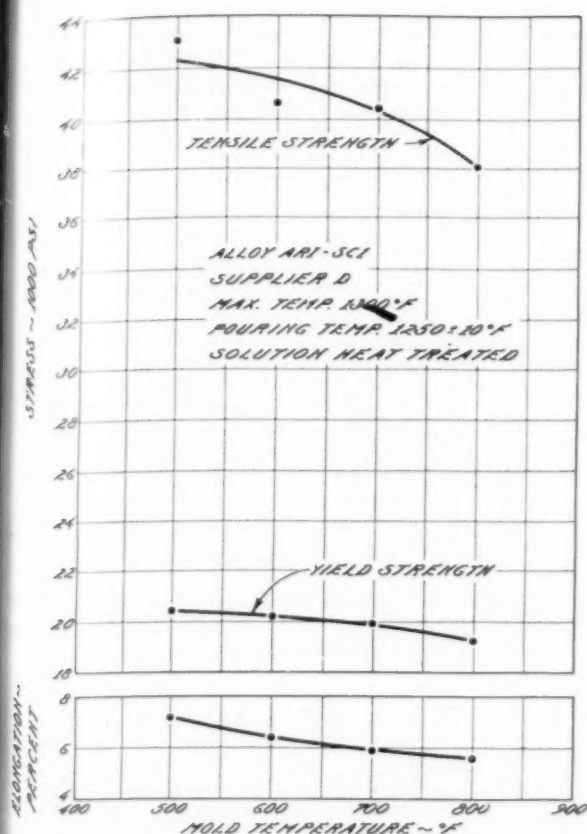


Fig. 19—Effect of mold temperature on tensile properties of ARI-SC1 permanent mold alloy in the solution heat treated condition.

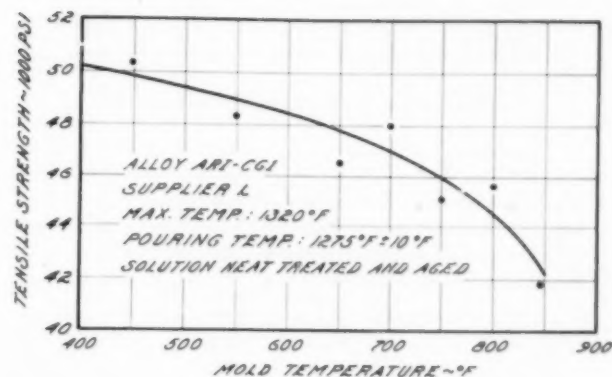


Fig. 20—Effect of mold temperature on tensile strength of ARI-CGI permanent mold alloy in the solution heat treated and aged condition.

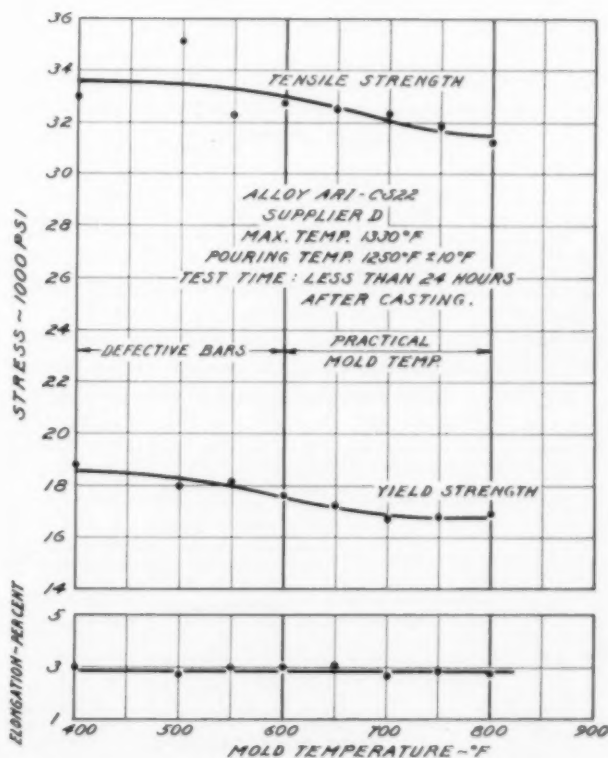


Fig. 22—Effect of mold temperature on tensile properties of ARI-CS22 permanent mold alloy.

temperature as the operating temperature is lowered.

When molten metal is poured into a permanent mold, the mold temperature will change, the change being dependent upon the relationship between the radiation of heat from the mold to the air and the absorption of heat from the metal to the mold. Figure 15 shows the maximum rise in temperature of the mold from the operating temperature for various initial mold temperatures and for a pouring temperature of 1250 F. It will be noted that the curve in Fig. 15 would cross the abscissa (zero rise in mold temperature) for a mold temperature slightly higher than 850 F. This is in agreement with the fact that the maximum operating temperature was found experimentally to be 850 F for the pouring temperature 1250

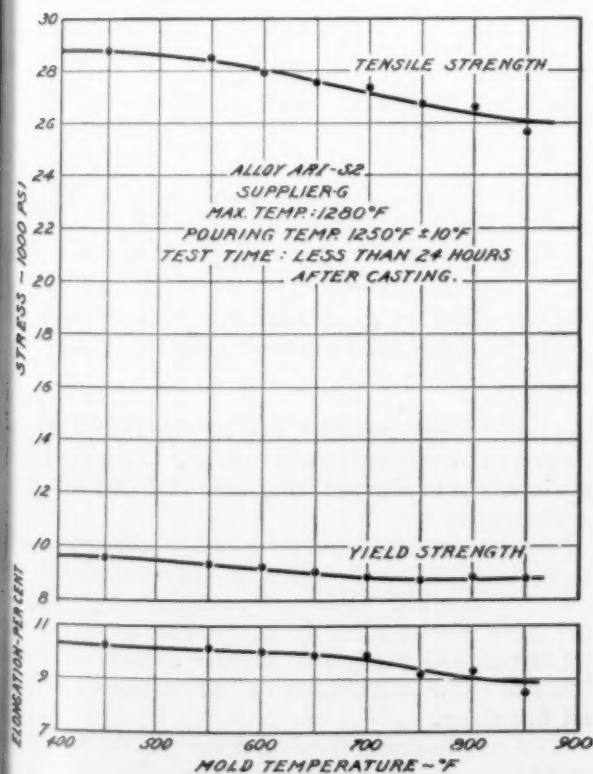


Fig. 21—Effect of mold temperature on tensile properties of ARI-S2 permanent mold alloy.



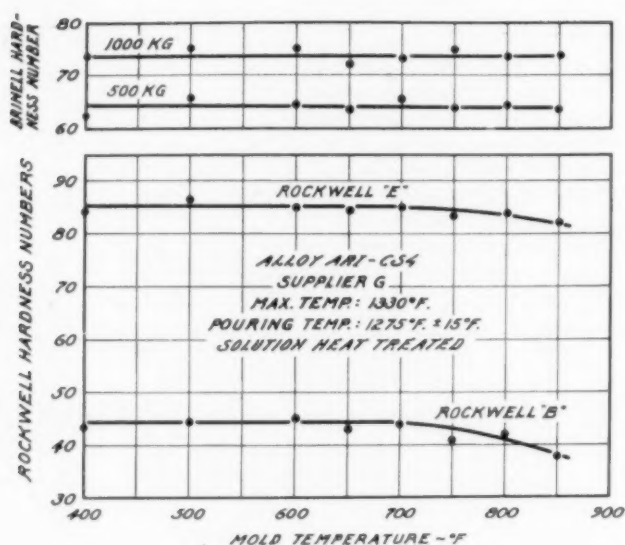


Fig. 23—Effect of mold temperature on hardness of ARI-CS4 permanent mold alloys in the solution heat treated condition.

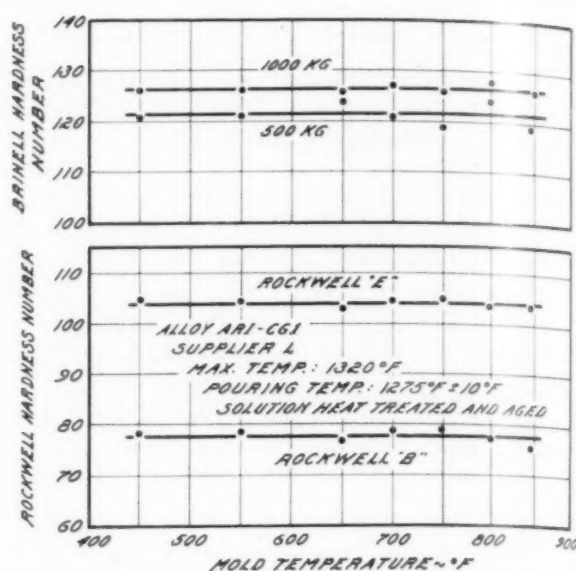


Fig. 24—Effect of mold temperature on the hardness of ARI-CG1 permanent mold alloy in the solution heat treated and aged condition.

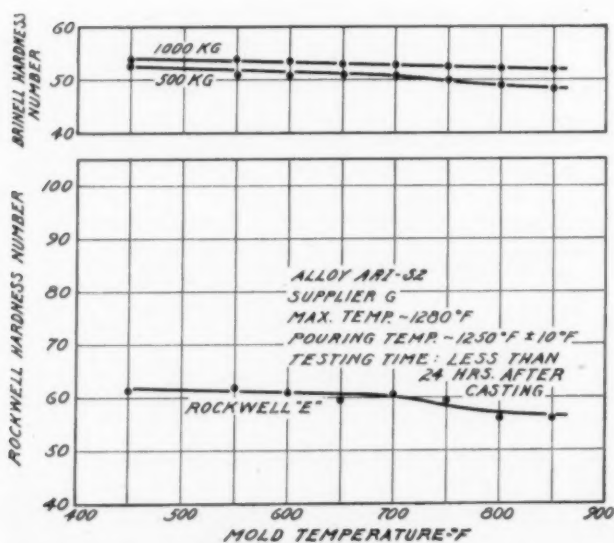


Fig. 25—Effect of mold temperature on the hardness of ARI-S2 permanent mold alloy.

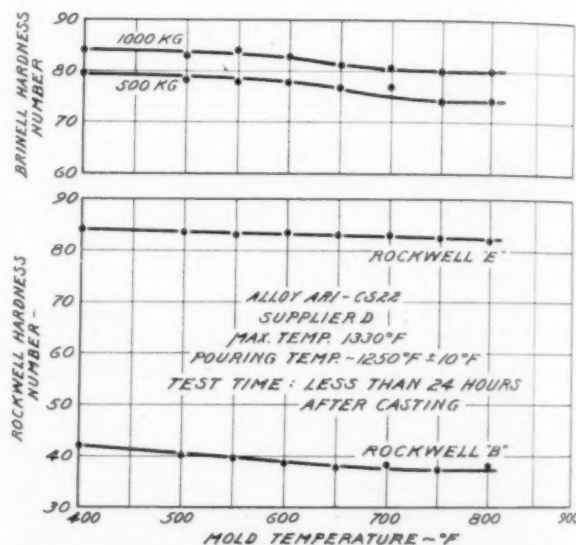


Fig. 26—Effect of mold temperature on hardness of ARI-CS22 permanent mold alloy.

+25 F. A mold temperature above 850 F may be obtained only by the addition of heat from an outside source. Below 850 F the mold may be operated without the addition of heat other than that supplied by the metal. All mold temperature measurements represent equilibrium thermal conditions. This necessitated preheating the mold to the temperature under consideration and then pouring castings until no fluctuation in mold temperature occurred for a given cycle. If the mold was preheated carefully, two castings were sufficient to establish an equilibrium temperature.

**Tensile Properties**—The effect on the tensile properties of the mold temperature was investigated by holding the melting and pouring temperatures within

the limits specified earlier, and varying the mold temperature through mold cycle changes. Generally, three castings (six tensile test bars) were cast for each mold temperature.

The properties of test bars cast at different mold temperatures generally varied in the same manner for all alloys studied. Figures 16 to 22 show the effect of the mold temperature on the tensile properties for all five alloys investigated. Figures 23 to 27 show the effect of mold temperature on the hardness values of all five alloys.

The tensile strength of all alloys decreased as the mold temperature increased. Alloys which were heat treated showed a much more marked reaction to mold temperature than the non-heat treated alloys. Alloys

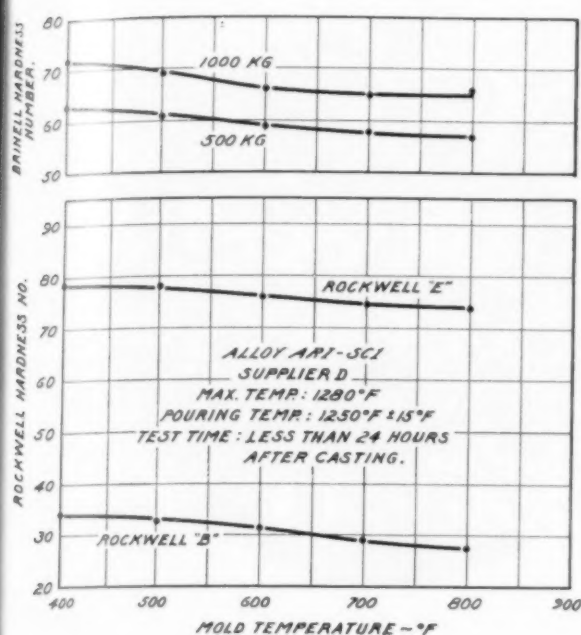


Fig. 28—Effect of mold temperature on cooling curves of alloy ARI-CS4.

ARI-CS4 and ARI-SC1 were investigated in both the solution heat treated condition and the "as cast" condition to show the greater response of heat treated specimens to mold temperature.

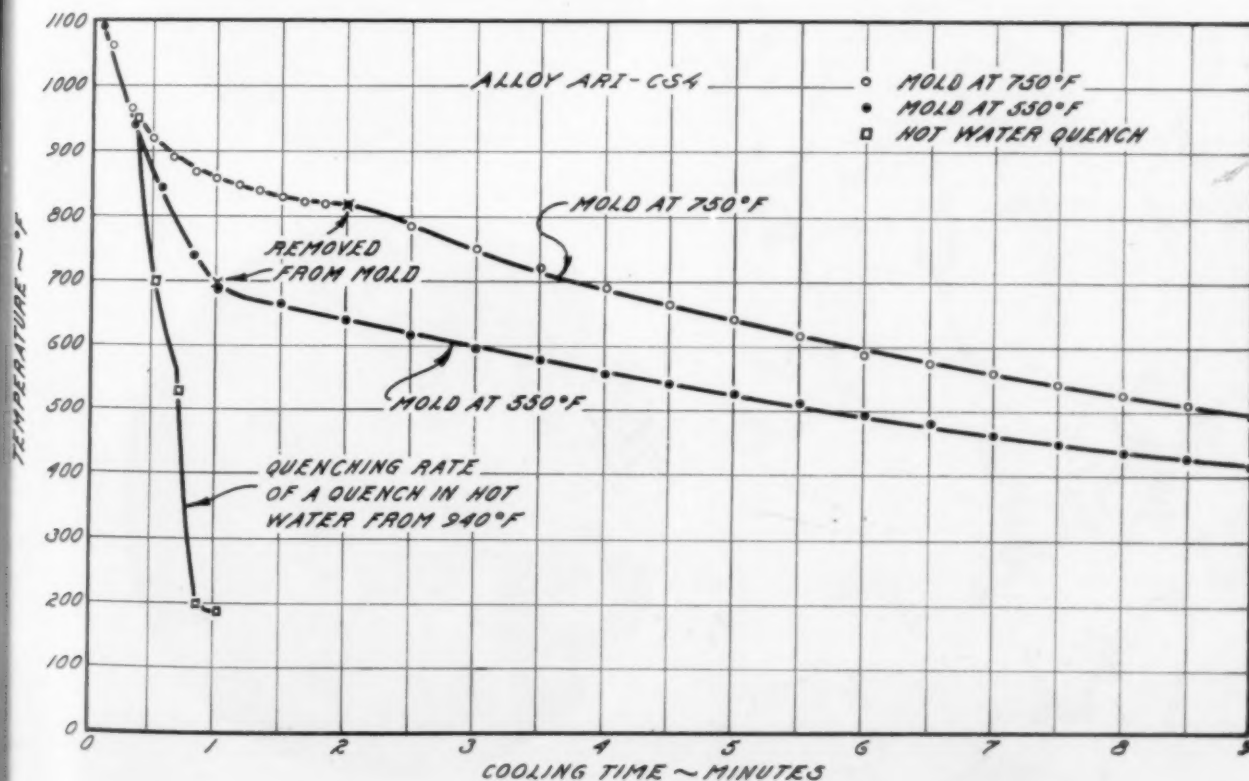
A comparison of Fig. 16 and 17 shows the effect of mold temperature upon heat treated as against "as cast" alloy ARI-CS4. For an increase in mold temperature from 600 to 800 F, the heat treated specimens,

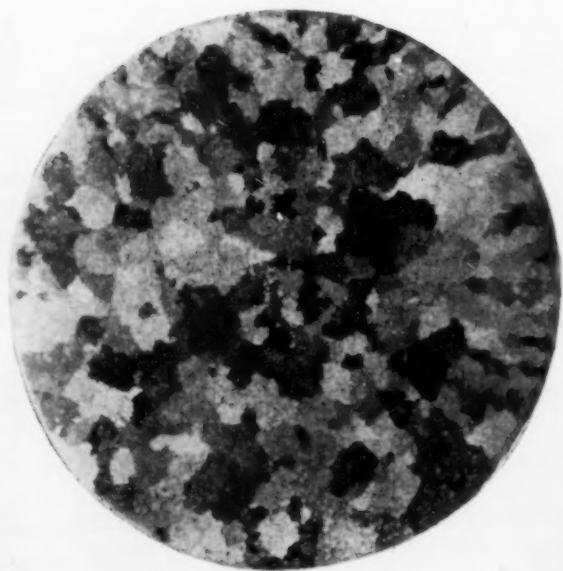
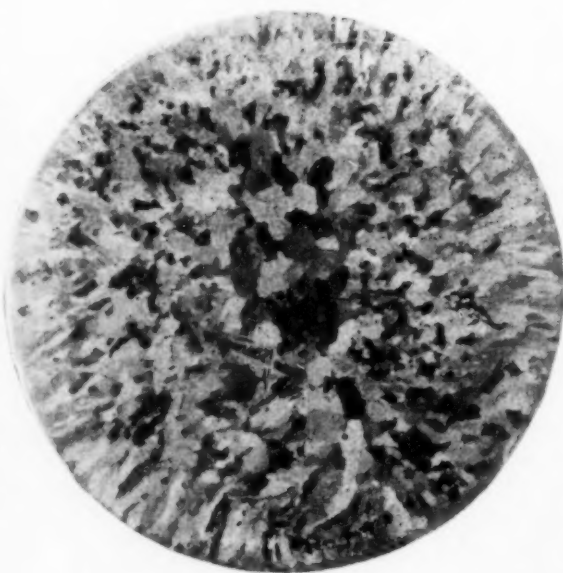
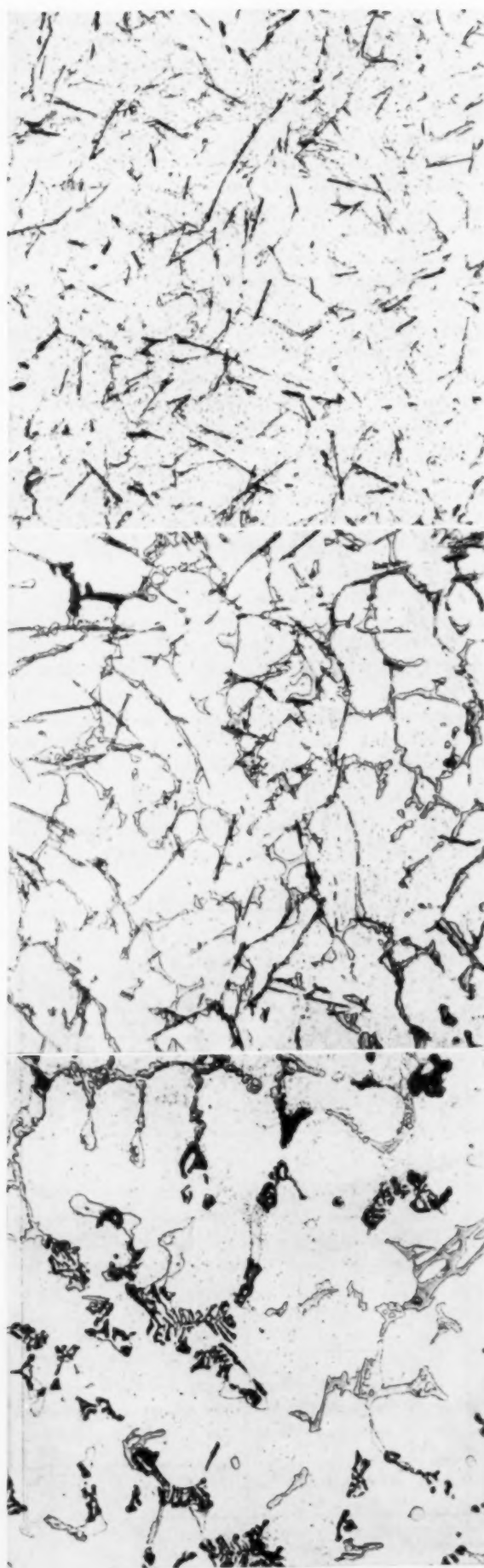
Fig. 16, showed a reduction in tensile strength of about seven per cent. The "as cast" specimens of the same alloy, Fig. 17, revealed a reduction in strength of only 1.5 per cent for the same increase in mold temperature. A comparison of Fig. 18 and 19 shows this same increase in effect of mold temperature with heat treatment, for alloy ARI-SC1. The "as cast" tensile strength of ARI-SC1, Fig. 18, drops about three per cent for an increase of mold temperature from 600 to 800 F, while the solution heat treated tensile strength of the same alloy, Fig. 19, drops about nine per cent for the same increase in mold temperature.

The greater effect of the mold temperature on the tensile properties of the heat treated alloys than the non-heat treated alloys can be explained by the fine grain structure of the more highly chilled castings, and the fact that more of the soluble constituents remain in solution in the "as cast" structure of the more highly chilled specimens. Thus, specimens cast into a cool mold not only contain more solid solution before heat treatment, but they also respond more readily to the solution heat treatment because of the finer grains which provide larger areas of micro-constituents for reaction.

For all of the alloys, a considerable drop in tensile strength took place for an increase in mold temperature from 600 to 800 F. However, for mold temperature from 600 to 400 F the change in tensile strength was present but in a smaller degree. However, it must be kept in mind that the probability of casting defects caused by the lower mold temperatures may counteract the benefits of the more rapid chill.

Fig. 27—Effect of mold temperature on hardness values for ARI-SC1 permanent mold alloy.





A—800°F  
Fig. 29 (above)—Effect of mold temperatures on micro-  
etch, Fig. 30 (below)—Effect of mold temperature on  
macrostructure of same alloy ARI-CS22. Mag. 6X.

B—600°F

C—400°F

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The effect of the reduced castability of alloy ARI-CS1 at low mold temperatures is shown in Fig. 17. The specimens shown in Fig. 17 for mold temperatures below 600 F were visually defective because of improper filling of the mold. With special care this fault was eliminated, as is shown by the fact that the bars, whose properties are represented in Fig. 18, were sound down to a mold temperature of 400 F. Of the alloys investigated, those requiring special care in casting into low temperature molds were ARI-CS22, ARI-CS1, and ARI-CG1. The alloys ARI-SC1 and ARI-S2 resulted in no casting defects for mold temperatures as low as 400 F, which was the end point for this investigation.

The yield strength of all alloys except ARI-CS4 showed variations with the mold temperature similar to those of the tensile strength. Alloy ARI-CS4 exhibited a slight increase in yield strength with increasing mold temperature in both the solution heat treated and "as cast" conditions, Fig. 16 and 17. The reason for this behavior has not yet been determined. It should be noted that this behavior is unique with alloy ARI-CS4, since alloy ARI-SC1, which is chemically similar to ARI-CS4, decreases in yield strength with increasing mold temperature in the heat treated and "as cast" conditions, Fig. 18 and 19.

The elongation for the ductile alloys exhibited the same type of curve as the tensile strength. The elongation for the less ductile alloys showed no variation with mold temperature, probably because the changes were of the same order of magnitude as the errors in measurement of the elongation. This supposition is substantiated by comparing Fig. 18 and 19. Alloy ARI-SC1 was used in both cases.

The elongation of test bars in the "as cast" condition, Fig. 18, did not respond to mold temperature change, but those in the solution heat treated condition, Fig. 19, showed a considerable change in elongation. Although this increased effect may be explained partially by the increased response to heat treatment, some of the change must be accredited to the greater sensitivity of the higher elongations.

The effect of the mold temperature on the hardness of the five alloys as shown in Fig. 23 to 27 is only slight. The hardness values tend to follow the same trend as the tensile strength for variations in mold temperature. However, the insensitivity of the hardness tests for cast alloys makes it difficult to examine changes of the order shown in the tensile tests.

**Chilling Rate**—The change in properties with a change in mold temperature seems to be dependent almost entirely upon the different chilling rate of the mold at various temperatures. Consequently, to observe the difference in the rate of chill of molds at different temperatures, bars were cast with a thermocouple at the reduced section. The thermocouple was coated, except at the tip, with a thin layer of mold coating for insulation. Before the casting was poured the thermocouple was heated to the temperature of the mold to reduce its chilling effect.

Figure 28 shows the cooling curves for bars cast into molds at 750 and 550 F. The bars were removed after two minutes and one minute, respectively, to be

consistent with the practice (established at the start of this investigation) of keeping the casting in the mold for a period of time twice the time required for the riser to freeze.

In Fig. 28 the cooling rate of a casting quenched from 940 F into water at 210 F was superimposed upon the other two curves, to give a comparison with the quenching rate achieved in the usual heat treatment. The results of this investigation are especially interesting in the region from 1000 to 700 F, which range contains the usual solution heat treating temperature. The more rapid cooling in the colder mold through this range agrees with the previously noted fact that the heat treated alloys cast in a relatively cold mold show greater response to the heat treatment because of a larger amount of retained solid solution.

To verify the explanation of the increased properties the response to heat treatment of the more drastically chilled specimens because of small grain size, micro- and macro-examinations were made of castings of alloy ARI-CS22 which had been cast at various mold temperatures.

Figure 29a shows the microstructure of a bar poured into a mold at 800 F. Figures 29b and 29c show the finer grained constituents resulting from casting into molds of 600 and 400 F. The effect of a change in mold temperature from 800 to 600 F is much more appreciable than a change from 600 to 400 F. This agrees with the fact shown previously that the tensile properties are affected more drastically in the mold temperature range 800 to 600 F than the range 600 to 400 F.

The macrostructure shown in Fig. 30a, 30b, and 30c exhibit the same grain structure differences as were found in the microstructure. Figure 30a shows equiaxed grains throughout the cross section, which are about the same size as the few grains at the center of the structures shown in Figs. 30b and 30c. This suggests that the cooling rate of the center of the bars cast into molds at 600 and 400 F is about the same as the whole bar cast into a mold at 800 F.

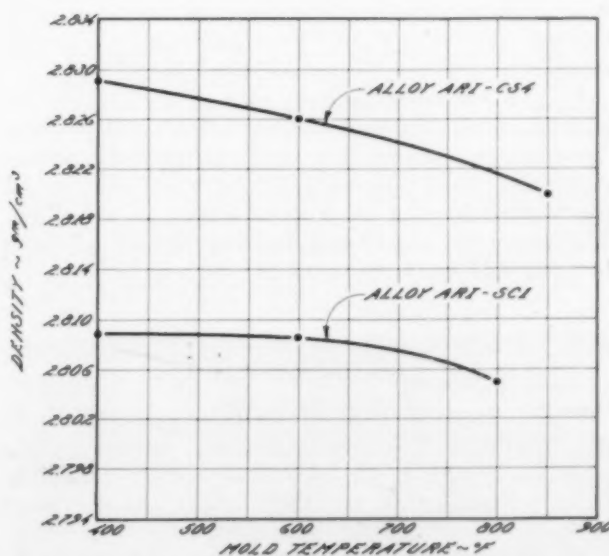


Fig. 31—Effect of mold temperature on density of alloys ARI-CS4 and ARI-SC1.

**Density**—In view of the reduction of grain size with a decrease in mold temperature, it would be expected that an increase in density would also be accomplished by the use of low mold temperatures. To establish this, a set of density measurements was made on alloys ARI-SC1 and ARI-CS4. In measuring the density, small cylinders were machined from the reduced section of bars cast at different mold temperatures, measured dimensionally very accurately, and weighed carefully. The densities as computed from these data are plotted in Fig. 31. The resulting curves show a slight increase in density with a decrease in mold temperature.

### Conclusions

From this investigation the following conclusions may be drawn.

1. A practical operating temperature for the mold described may be chosen over a range of temperatures from 600 to 800 F. The selection of the operating temperature is dependent upon the demands imposed by the time required for the operating cycle, the maintenance of a continuous mold coating, and the physical properties of the castings. As the mold temperature decreases, the time required for operation increases and the mold coating is more susceptible to damage, but the physical properties are higher. Therefore, a cycle must be chosen which permits good physical properties and at the same time produces a sufficient number of castings without damage to the mold coating.

2. The thickness of the mold coating has a slight effect upon the properties over a wide range of thicknesses. In the range of usual operation conditions, the mold coat thickness has an effect upon the properties within the limits of ordinary scatter. The yield strengths and the aging characteristics remained unchanged for all thicknesses investigated.

3. The time for which the casting is left in the mold affects the tensile strength about three per cent for times from 1.5 to 4 minutes in the mold. The shortest time in the mold results in the highest tensile strength and elongation. The yield strength is not measurably affected by various times in the mold.

4. The optimum pouring rate varies slightly for various alloys. For some alloys, a too rapid pouring rate forms a wrinkled skin which decreases the tensile properties. Slow pouring rates, less than 0.20 pound per second, have a detrimental effect on all alloys. The range of proper pouring rate for all alloys is between 0.60 and 0.20 pound per second.

### Acknowledgments

The authors wish to express their appreciation to the Aluminum Research Institute for its sponsorship of this work and to the individual member companies of this organization for their advice on the mold design.

An expression of gratitude is due to Messrs. M. L. Fried, A. R. Toole, and M. H. Jones for their aid in the experimentation.

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### DISCUSSION

Chairman: W. J. KLAVER, Aluminum Industries Inc., Cincinnati.

Co-Chairman: M. E. GANTZ, American Magnesium Co., Cleveland.

L. F. SWOBODA (Written Discussion)<sup>1</sup>: The authors of this paper are to be commended for their thorough investigation of the effect of mold operation variables on properties of permanent mold castings.

As a user of permanent mold castings in highway transportation vehicles, I am pleased to know that the mechanical properties of castings are not too adversely affected by the many modifications or compromises frequently found necessary to get satisfactory appearing castings from a mold. The design characteristics of many castings are such that wall thicknesses, gate size and location, are fairly well determined in advance and cannot always be altered to suit best foundry practice. Therefore, it is often necessary to resort to changing thickness of mold wash, mold temperature, pouring temperature, pouring rates or modifying gates in order to completely fill mold cavity and produce castings without objectionable shrinks. Likewise, a poor guess in the mold design details may also frequently be responsible for the necessity of resorting to operating practices not consistent with obtaining the maximum mechanical properties. But again, when castings are urgently needed there is little chance of any major mold change being made when one or more of the simple "Compromises" previously mentioned will permit production of usable, but not necessarily the best castings.

There are, however, many applications where only the best quality castings can be used and it is for their production that the information presented in this paper should prove invaluable.

We would also like to comment on the fact that the experimental work was made on a casting of simple shape and fairly uniform cross section, and that it would probably not be possible to realize the same gain in mechanical properties in each part of an odd-shaped casting having considerable variation in section thickness.

It would have been interesting to repeat tests on effect of mold wash thickness (Fig. 8) with other types of mold washes to determine whether or not 0.0025-in. wash thickness was optimum for all types.

Other information of interest would be the determination of optimum mold wash thickness for various casting thicknesses. However, this would have required another permanent mold for producing a stepped test specimen.

The goal of designing a mold having a short operating cycle, that can produce good castings with minimum amount of metal, having gates located where they can be removed with least effort, without defacing casting is one not always possible to attain. However, it would be well for the mold designer to consider and apply wherever possible the principles given in this paper.

WALTER BONSACK (Written Discussion)<sup>2</sup>: The authors should be

<sup>1</sup> Reynolds Metals Co., Louisville, Ky.

<sup>2</sup> Apex Smelting Co., Cleveland.

congratulated for a fine presentation of a subject as important as this one. Not only do they give a fine detailed description of the mold design construction and operation, but they also bring excellent data on the effects of the process variables.

There has always been a need for a permanent mold for aluminum alloy test bars. Since this mold is relatively inexpensive to make and since it proved itself a reliable reproducer of properties on aluminum alloy test bars it recommends itself for universal use in the laboratory and in permanent mold foundries where test bars have to be cast for acceptance of castings. Several copies of this mold are in existence and have been used to check the work done by the authors. The agreement in results was very good. The A.F.S. Aluminum and Magnesium Division's Committee on Test Bars have this mold design under consideration and needless to say that the adoption of this mold design by the American Foundrymen's Society as a standard would be a step in the right direction.

A few words of warning should be given to those not familiar with permanent molding. The various factors and effects discussed in this paper are characteristics of this mold only. All permanent molds will follow this mold in a general way but each mold must be studied in a similar manner to find its best work cycles. Again, generally speaking, the heavier a mold, the greater the chill, the lower the mold temperature, the shorter the working cycle, the more castings can be cast per time unit. On the other hand the lighter the mold, less chill is obtained, the longer the work cycle, but the thinner a casting can be cast.

It is very hard to determine theoretically the right proportioning of all these factors on permanent molds. A mold may be so heavy that an operator is not able to keep the mold at operating temperature without additional heat or a mold may be too light so that not enough castings can be made from the mold without artificial cooling. Both facts indicate an improper proportioning of the mold mass, the casting mass, and thicknesses of wall sides of both. Designing a mold properly is mostly guided by experience. Therefore in order to utilize a "standard" test bar mold all dimensions and factors must comply with the standards because any deviation may cause different characteristics.

I believe that the authors and their sponsors have rendered a good service to the foundry industry by this work and this presentation.

#### MR. SUGAR'S DISCUSSION

ALFRED SUGAR (*Written Discussion*)<sup>3</sup>: Until quite recently very little data on permanent mold casting has been presented for general consideration. The authors of this paper, "The Development of a Permanent Mold for Aluminum Test Bars," should be commended for the thought and effort that they devoted to the accumulation of information contained in their study.

If considered strictly on the basis of title, it is quite possible that my comments may be out of order, but when reading the paper it seemed quite natural for me to think of the engineering application of these various points brought out and my remarks are based on this broad basis.

Under the sub-heading of "Hinging," the wear advantages of the bottom hinge over the side hinge are indicated. In my experience, the incidence of wear is affected more by the design of the mold than by the location of the hinge. This study does not seem to offer any concrete evidence of superiority in either direction. Dowel pins do not reduce possible wear as stated by the authors but do minimize the effect of wear by making more positive the correct alignment of mold segments. Some guide for positive mold alignment is good design practice in all cases.

The gating system shown in the paper is quite obviously a compromise. It is good foundry practice to keep the sprue full of molten metal for the duration of the pour. The gating system should, at least partially, control feed and rate of pour. The system as recommended places these controls in the hands of the pourer, thus introducing another variable. As noted further in the paper, under "Pouring Rate" a hard pour will result in a defective casting and a very slow pour will also give poor results. A well designed gating system should tend to give controlled rapid feed with little or no turbulence. A decided improvement could have been realized by placing feeders between the gate and casting cavity,—leading metal from gate to feeder to cavity.

The exact point at which the temperature is measured is quite important. It would be well to cite the location of the thermocouple in relation to the face of the mold cavity. I would be most interested in mold temperatures at several spots on the face of the cavity. Temperatures here will vary throughout the cycle as molten metal is poured into the cavity, as it solidifies, and also with the cross section of the casting. The further from the cavity surface the thermocouple is located, the less the likelihood of showing measurable thermal fluctuation. I am sure that had a number of thermocouples been used in different areas at different distances from the face of the cavity and behind different casting cross-sections, a great deal of interesting and valuable data would have been obtained.

The paper over-simplifies the role of mold coating. One of the major functions of mold coating is to compensate for deviations in mold design. On page 320 and in Figs. 4 and 5, the authors point out the elimination of shrinkage through a change in coating thickness—true enough the change was made in the gate area but nevertheless the results were obtained through a change in coating thickness. The authors also state that the mold coating in the casting cavity must be tapered, a change in thickness, to eliminate small surface shrinks. Since "the production of sound castings is of prime importance" due recognition should be given to the part played in the production of sound castings by the correct trimming of mold casting. Will the authors please clarify the first sentence in the first paragraph on page 321 on "normal variations in thickness."

The paper indicates a very pronounced effect on soundness of casting as determined by the pouring rate. Did the authors stop at a pouring rate of 0.80 lb per sec because castings poured above that rate were defective? It would be interesting to know the percentage of defective castings produced at different pouring rates for although, as indicated, physical properties may not be affected, an engineering casting with shrunken sections is generally not acceptable.

There should be a noticeable effect on the mold operating cycle for variations in alloy composition. A marked difference in operating cycle should be noted between alloy ARI-CS-22 and ARI-SC-1 or ARI-S2 and a eutectic silicon-aluminum alloy.

The importance of pouring temperature is, in my opinion, under-rated in this paper. It is one of the prime factors in the control of mold operating temperature which, in turn, is among the most important factors in the successful operation of a permanent mold. In practice, it is not feasible to keep a mold idle until it returns to a predetermined temperature through part of each cycle. External heating or cooling, adjustments in pouring temperature and in continuous operating cycle are means used to obtain the desired ends,—the control of the mold operating temperature. It is comparatively easy to pin down a desirable low pouring temperature for a test bar mold but in practice the pouring temperature chosen may be considerably higher than that deemed to be "good practice"—due to insufficient feeding by the gating system or to extremely thin sections in the casting but, of course, these points are extraneous when considered in the light of application to the development of a test bar mold.

I have the impression that very few castings were poured at each set of conditions. Was a sufficiently long run made in every case to stabilize the mold?

R. A. QUADT (*Written Discussion*)<sup>4</sup>: The authors are to be congratulated for presenting an interesting study of the variables affecting the successful operation of a permanent mold. Figure 32 shows a test bar casting produced from a permanent mold

<sup>4</sup> American Smelting & Refining Co., Barber, N.J.

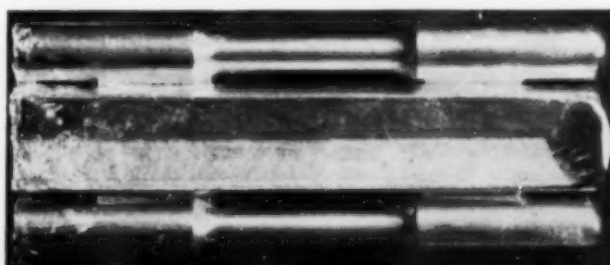


Fig. 32

<sup>3</sup> The American Metal Co., Ltd., New York.



developed by Federated Metals Division of the American Smelting and Refining Company at their Central Research Laboratory. When the Case School program on permanent mold properties was initiated we were requested to supply the details for the production of this mold since it represented an extremely cheap mold to build, one that was simple to operate, one that produced optimum mechanical properties, and a mold of very little bulk since the dimensions of the entire mold in operating position is only 2 in. by 4 in. by 10 in.

As will be noted from the photographs presented in the paper the authors reduced the sprue dimensions on the theory that the proximity of the relatively heavy sprue to the reduced tensile section would produce a hardness gradient across the tensile area. In the thousands of test bars produced from the Federated mold no indication of this condition has ever been experienced. Further, the authors have modified the design of the mold as if it were planned to produce a test bar mold that would be operated on a continuous production basis. From the foundry point of view the ideal mold is one that can be removed from a preheating furnace and into which metal can be poured with the expectation that acceptable bars will be produced from the first six or seven castings poured. After these few castings have been produced, the mold is returned to its storage place to be used at a later period. Under these conditions, for a constant mold temperature, no great effect would be expected in view of the relatively large mass of the mold.

Much of the preliminary work required and described by the authors with regard to shrinkage at the gates would never have been experienced had the sprue dimensions been utilized as specified in the original design. Our developmental work on this mold also included a smaller sprue than that shown in the photograph (Fig. 32) but we were forced to utilize the present dimensions in order to eliminate the necessity for painstaking control of the mold wash at the gates and on the shoulders of the test bars.

It is believed that the sprue design used in the Federated mold is superior since it permits the production of tensile specimens of equivalent or better properties without the need for such sensitive control of mold wash thickness. It also produces better properties, due to the shorter gates and better feeding, from alloys with poorer casting characteristics.

MEMBER: Do the properties given in the paper hold true in the general run of castings?

MR. BONSAK: In general, test bars have their properties and castings have theirs. You cannot conclude from test bars that the castings will have the same properties. Nevertheless in producing test bars along with castings you have a check on your foundry technique, melting, and heat treating. That is the only reason test bars should be made.

This mold, which has been developed by Case Institute of Technology in cooperation with the Aluminum Research Institute, has been tried by several members of the Aluminum Research Institute. The results of the check work are in very close agreement with those found by the Case Institute of Technology. It is gratifying to have a permanent mold which can be used as a standard mold. We can improve on it, although all the checks that have been made are very close and indicate that this is a good mold to have around to study alloys or to check your casting process. I would recommend that the A.F.S. Test Bar Committee seriously consider this design and adopt it as a standard mold.

SAM TOUR<sup>2</sup>: We have long needed a good test bar mold in connection with permanent molding of aluminum alloys. I am a little disturbed, however, by the tendency to judge the goodness or badness of the test bar mold by the yardstick of how high a tensile strength result you can get on a tensile test. Is it the purpose in developing a standard test bar to find out how high a tensile strength can be obtained from a given alloy, or is it the purpose to develop a mold which in itself will give uniformity from plant to plant, whether the results are up at this high level or a little lower? What I am trying to point out is that the mold presented in this paper has been described as being very sensitive to the amount of coating and the distribution of that coating in the die, the criterion being to distribute the coating to get maximum tensile strength.

It seems to me the best approach would be to develop a type

of coating which would make that mold the least subject to variables in tensile strength whether that tensile strength happens to be the ultimate or not that can be obtained from the alloy. If we have to teach men in different plants how to apply the mold coating, to make it 0.005 in. thick at the center of the gage section and to make it 0.008 in. thick on the shank and 0.012 in. thick on the riser, we have certainly put a lot of stumbling blocks in the way of making this mold useful in many plants. I should like to see a mold developed that is more foolproof. You might not get such high tensiles but the shop man could use it and find out whether his alloy was any good.

MR. EBERT: Perhaps I made that mold coating technique sound too complicated. Actually it is not difficult. The coating of any permanent mold is a matter that requires some experience. I think you will all agree with that. The mold coating procedure was outlined briefly to the employees of several companies which have the same mold. After a brief instruction, they were able to use a mold coating technique which was perhaps not exactly the same, but a mold coating technique that gave the tensile properties to which Mr. Bonsack referred. The agreement in properties was very good. I think that there were seven or eight different companies that tried it.

As to the matter of application of the coating, although I mentioned figures in a thousandth of an inch, we did not know actually what they were until we measured the resultant castings. We measured a large number of castings and found that our own technique did produce very reproducible mold coating thicknesses (as determined by the dimensions of the test bar) although we made no particular effort to control it.

I might describe the mold coating technique briefly. Test bars were put into the test bar cavity prior to the mold coating and the sprue and ingates were then sprayed heavily. The test bars were removed and the top of the test bar, which acts as a riser, was sprayed rather heavily. The coating thickness on these portions of the cavity was perhaps  $\frac{1}{16} \pm \frac{1}{4}$  or perhaps even  $\frac{1}{2}$  in. After the sprue and end-gates were coated and the test bars removed from the cavity, the test bar reduced section was sprayed lightly. Then the center part of the reduced section was rubbed down with steel wool until it reached a characteristic color. In other words, most of the mold coating is rubbed off at the center of the reduced section. The taper probably was not uniform from the center to the end of the reduced section. However, it did yield good reproducible results.

MR. BONSAK: Mr. Tour brought up the subject of doping. There are many permanent mold foundrymen who know the art of doping permanent molds. That is inherent in the process. You cannot get away from it.

We have material below the coating which will gradually oxidize and from its own coating. That is the iron of the mold. It will oxidize and form its own insulation layer and will throw results off considerably if you do not watch that coating too. That means, occasionally the test bar mold must be sandblasted to get the oxide coating off. Then you start over again. Unless we can produce a mold out of a self-insulating, non-oxidizing material you must resort to measures used in all permanent mold processes.

MR. SUGAR: I doubt if the authors have stressed the importance of coating enough. They did tell about tapering and its importance. In the paper they do bring out how they eliminated shrinkage merely by changing the thicknesses of mold coating. As Mr. Bonsack stated mold coating is one of the requirements in production of permanent mold castings.

There were several other items in the paper that I would like to bring up. One of them is the gating system. As Mr. Ebert said, and as is quite obvious, the gating set-up is a compromise. I doubt if it controls the feed into the mold cavity enough. I do think if a parallel feeder had been placed between the sprue and the casting cavity itself possibly some of the need for accurate control on the coating would be minimized. That is, you could be less careful of your mold coating and still be able to get a uniform feed, etc.

Pouring rate was mentioned in the paper. This also comes back to the gating system. You stop your curve at a pouring rate of 0.80 lb per sec. What is the reason for that? Above that pouring rate did you get nothing but scrap castings?

In your paper you also mentioned that you did get defects in castings poured both slowly and at a high rate of speed; yet the

<sup>2</sup> Sam Tour & Co., Inc., New York.

physical properties of the casting were not affected by these surface shrinkage. Unfortunately, no engineering castings with shrinkage are acceptable, and I do not think a test bar should be acceptable if it shows a shrink, even though its physical properties may be satisfactory. I think it would be interesting if you could give the percentage of scrap or defective castings at different pouring rates.

The author touched on mold temperature rather lightly. As I understand it, he had only one thermocouple placed behind the narrow section of the test bar. I do not know how far from the cavity that thermocouple was, but if he were to place the thermocouple, or a series of thermocouples at varying distances from the cavity surface and behind varying section thicknesses of the casting, I think he would find several different temperatures from which he could draw interesting conclusions.

MR. EBERT: We did not intend to overlook the points which Mr. Sugar mentioned. We tried to make the mold as simple as possible so that die-sinking operations were not needed in the manufacture of the mold. It is quite possible that a feeder would help to make the mold coating less sensitive. There is another disadvantage to the feeder, and that is that feeding takes place in the reduced section of the bar. This means that the machining difficulties and the trimming difficulties are increased. In a casting of the type produced in the mold described in the paper no machining operations are needed other than just sawing off the sprue.

I do not have figures on pouring rates and the number of defective castings. The reason that the properties look good in spite of the fact that there are defects is that all bars with defects were earmarked before testing. Any bars that broke in the defect were discarded. It was only when the defect was sufficiently small so that the test bar broke some place else that the value was used. The defective bars were added merely to show the range in which defective bars are produced. At the excessively high pouring rates, there were some good bars, but at the very low pouring rates none of them was good.

As to the matter of thermocouple location of the temperature, we did not make a temperature exploration of the mold. The thermocouples were placed behind the reduced section of the test bars and half way through the thickness of the mold wall at that point. The temperatures were recorded merely as a control measure. We had to know relatively what the temperature of the mold was. I am quite sure that it would probably be different in different parts of the mold. I do not know how much different. The temperature measurements were for control.

R. F. HAUSER: Did the author say that the pouring rate was 0.8 lb per sec? How is the pouring rate determined?

MR. EBERT: Yes. The pouring ladle used was one with a closed spout. By varying the size of the nozzle in the spout (by adding inserts) we could vary the dimensions of the stream of the liquid metal. By measuring the filling time and also weighing the final casting, it was possible to determine how many pounds of metal were poured per second.

MR. HAUSER: Actually you only measured the pouring rate of the metal leaving the pouring lip then.

MR. EBERT: That is right.

MR. HAUSER: If you were to use this data for a casting with a higher head on the sprue it would not hold then, would it; say a deeper casting and a longer sprue?

MR. EBERT: Yes, that is probably right.

MR. HAUSER: The data is only good for this one particular mold?

MR. EBERT: Yes.

MR. SUGAR: We did not touch on temperature control either of the mold or of the metal. Normally in practice you do control cycle at least partially through mold temperature. You have cycle effects on mold temperature, etc., due to variations in metal pouring temperature. How many pours did you make in order to get a measurable effect? My impression from the paper was very few, and it does take quite a while for any mold to stabilize. When you pour, say, 1½ lb of metal into a mold that weighs 10, 15 or 20 lb the effect is rather slow in coming. Did you run over a long period of time before taking test samples?

MR. EBERT: Yes, that was not pointed out very clearly. The mold was preheated artificially to the normal operating temperatures as determined by the thermocouple. After that from 10 to 12 castings were poured and scrapped immediately. They were probably alright, but they were poured just to have conditions the same for the first good bar as they were for the last one. Each point on these figures represents a minimum of 10 tests. On some of the curves in which the mold temperature was exceedingly low it was not possible to get that many because the casting damaged the casting mold. We had to recut the mold and start over again or at least patch the mold.

MR. EBERT (*Authors' Closure*): The authors wish to express their appreciation for the interest exhibited by those who have discussed this paper.

In regard to Mr. Sugar's comments the authors are in general agreement concerning the hinging of a permanent mold. The gating of a test bar mold must be such that all of the alloys which are cast in the foundry should be castable in the mold. This means that the mold must lose some of its efficiency for any one alloy in order that it may be acceptable for all of the alloys. The location of the thermocouple is defined on page 319. The determination of the thermal gradients in the mold would be quite interesting but was somewhat out of the scope of this paper.

The effect of the mold coat must be broken into two parts. The first consideration is that of determining the proper coating to compensate for small mold errors. This might necessitate considerable effort when the mold is first used. However, once the coating technique has been established little effort should be required. The second consideration regarding mold coating is the effect of the human element when the coating has to be renewed. For this recoating, the effect upon the tensile properties is shown in Fig. 8 while the normal variations are given in the first paragraph on page 321.

The curves showing the effect of the pouring rate were terminated at 0.80 lb per sec because that was as fast as the authors could fill the mold (1.5 sec to complete a pour). Unfortunately the per cent of defective castings is meaningless unless a large number of castings is made. The effect of alloy composition and pouring temperature on the operating cycle are understandably small when the heat balance established is considered. Small variations in the heat content of 1 lb of aluminum can have only a very slight effect on the heat content of 40 lb of steel mold.

Mr. Quadt presents one of the two methods of making test bars. The one approach is to make the test pieces under production conditions while the other approach is to make test castings under laboratory conditions. The mold discussed in this paper has been an attempt to make castings somewhat similar to production castings. It would undoubtedly be of considerable interest to the permanent mold industry to have more information on the operating characteristics of the mold discussed by Mr. Quadt.

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# ENGINEERING EDUCATION FOR THE FOUNDRY INDUSTRIES

By

George K. Dreher\*

THERE are over 5000 castings companies in the U.S.A. Of these there are over 1200 employing more than 100 men and 944 more employing more than 50 men. These foundries employ a minimum of 475,000 men.<sup>1</sup> If we, empirically, assume one member of management for every 25 employees, then there are 19,000 such positions at the present time. If only one-half of these are to be engineering graduates (approximately 10,000) then the industry is confronted with generations of a continuing effort to supply candidates for such education. The industry must also, by one means or another, support their education and then provide the requisite training to indoctrinate the graduate engineer into the industry.

In 1824 the first English speaking collegiate school of engineering was established at Rensselaer Polytechnic Institute. Prior to that time all engineering knowledge was transferred by the laborious apprenticeship, craftsman, master cycle. Many industries were quick to adopt engineering education. This was true of the construction and, later on, the electrical, hydraulic, mechanical and chemical industries.

The foundry industry retained the old methods for over a century before serious changes were contemplated. More recently, economic conditions plus regulating local, state and federal measures have served to break up the old cycle. Thus the industry is presently in need of engineering knowledge to bridge the above gap in its existence and to keep it in step with the tempo of the industries it supplies with basic products.

## Purpose of This Paper

This paper, therefore, has a threefold purpose:

1. To recrystallize the objectives of engineering education as applied to the foundry industry.
2. To provide an interpretation of engineering education for the present executive, managerial and supervisory members of the industry.
3. To provide an avenue for constructive criticism, opinions, suggestions and objectives on the subject

which in turn will serve as a basis for a more thoroughly refined analysis of industry requirements blended with actual experience.

## I. The Foundry Industry

The growth of any type of education is dependent on its value to and acceptance by society. The growth of engineering education in the U. S. has been phenomenal when we realize that the first English speaking engineering school opened at Rensselaer Polytechnic Institute in 1824.

The report recently issued by the Engineers Joint Council<sup>2</sup> gives further evidence to this fact of growth. Since 1910 the profession has grown from approximately 85,000 to about 325,000 members. The numerous industries that have availed themselves of graduates include the electrical, machinery, public works, and steel, as well as those generally looked upon as competing industries such as welding and plastics.

The foundry industry has made extensive use of engineering in the design and development of productive equipment. Only recently has this "engineering method" found application in the products of the foundry. Need for a scientific approach to foundry problems and methods has been a subject of considerable discussion and has been pointed out in various talks and articles. Among the latter is the summary "College Graduates in the Castings Industry."<sup>3</sup>

The industry has worked closely with many educational institutions through the American Foundrymen's Association and the various trade societies on many technical subjects. This is especially true of the Educational Division of the American Foundrymen's Association under the direction of Fred G. Sefing\* as chairman, and H. F. Scobie. This latter work has brought forth a variety of magazine articles and papers which have become a background for educational activity. These references include "College Foundry Courses,"<sup>4</sup> "A University Course in Foundry Control Methods,"<sup>5</sup> "Foundry Practice for Engineering Students,"<sup>6</sup> "College Graduates in the Castings Industry."<sup>3</sup>

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The foundry industry has also accepted and is implementing a training program. The paper "Foundry Training Course for College Graduates"<sup>7</sup> by the American Foundrymen's Association subcommittee in training graduate engineers, and the article on "Co-operative Education"<sup>8</sup> have presented a well conceived plan for practical use that is applicable to large and small companies alike.

Foundry management seeks certain qualifications in these graduate engineers which are the result of prior experience on the former's part. In the past, graduates have been too unsettled on their jobs because of their desire to progress too fast without going through a program of sound development. This results in a type of engineer who eventually gets "promoted out" of the firm. The graduate must also possess the ability to have the staff like him at the outset and to further this regard through desirable action as time passes. Tolerance, friendliness and sincere interest in the members of the worker group are essential, but the graduate must also appreciate that he and his crew or department must meet or exceed standards of quality and quantity in order for him, or his crew, to become an economic asset to the firm. Diplomatic honesty in reports and interdepartmental cooperation, acknowledgment of ideas, suggestions and good work, plus a willing attitude, are essential to the individual's as well as the company's success.

The Foundry Educational Foundation has been successfully conceived and implemented to bring about the desires of the industry in regard to the preparation and attraction of engineering graduates. The support of the founding members (American Foundrymen's Association, Gray Iron Founders' Society, Malleable Founders' Society, Foundry Equipment Manufacturers Association) and the voluntary efforts of some of the leading foundry executives of these societies<sup>9</sup> has provided the impetus and means of accomplishing the desires of the industry. No better proof can be found for the interest of foundry management than in their providing funds for the realization of these needs. Educated and trained men are actively being sought by the leading foundry firms of the nation.

## II. The Universities

Engineering education in the U.S.A. is undergoing a major revision at present. As engineering education progressed through the nineteenth and presently through the twentieth century, new courses were constantly added. These, in turn, would displace humanistic, political, economic and business courses till the pre-World War II era when engineers, educators and businessmen alike suddenly realized that the graduate engineer was lacking in some qualities.<sup>10</sup> Various societies became interested in this problem. In time, this took the form of magazine articles, papers and propaganda. John S. Crout,<sup>11</sup> in proposing a plan for post-graduate training of engineers, is quoted:

"His work has brought him out of his natural habitat of the laboratory, drafting room, foundry, mine and factory to a position of prestige in a world governed by social, economic and political forces not taught him in the usual engineering curriculum."

His plan goes on to show how these omissions can be

eliminated by a post-graduate study and training plan. James C. Zeder<sup>12</sup> approaches the same problem by analyzing the common complaints of engineers in industry and recommending self-correcting procedures for the individual. Specific points of engineering success are identified as (a) ability to get along with people, (b) ability to look beyond his own department to see where he can be of help to the rest of the organization, (c) ability to plan his own progress by eliminating shortcomings from his personal mechanism. Edmund T. Price<sup>13</sup> tells what a young engineer should know. Quoting from his article:

"Bear in mind that success for you in your outfit comes from having a more than speaking acquaintance with sales, manufacturing and accounting."

Referring again to W. G. Van Note<sup>10</sup> the author finds him pointing the trend in reform along lines mentioned by the above engineers. T. L. Joseph<sup>14</sup> expresses the reaction of educators in a very simple way by stating:

"Ways and means are needed for weaving some of the threads of liberal education into the fabric of engineering education."

Research into industrial psychology is a comparatively recent science. Cleveland College of Western Reserve University has a Personnel Research Institute for the purpose of developing new techniques, tests, and measuring human reactions in relation to industrial environment. The advent of biomechanics or technobiology promises some revolutionary concepts of human work. Dr. Leonard C. Mead<sup>15</sup> calls it Human Engineering in the following quotation:

"Human engineering is engineering design with reference to man's anatomical, physiological and psychological capabilities and limitations."

Prior to World War II such things as X-ray, spectrograph, spectrophotometer, magnaflex, supermicrometers and electron microscopes were more objects of curiosity than of widespread use. Now they are standard industrial tools of a fundamental nature.

University function is to educate rather than train student engineers. Again the educational world uniformly strives to become the means by which fundamentals are taught. These vary with industrial usage and technical development, but always are subjects which are applicable to a variety of industrial processes and not restricted to one narrow field. W. G. Van Note<sup>10</sup> points out this trend in this quotation:

"—the philosophy underlying engineering education is today being based more and more on a broadening of the base of engineering education in which strong emphasis is placed on fundamentals, rigorous instruction in the use of English, and the inclusion of a well planned social humanistic stem."

Virtually all bulletins issued by colleges of engineering set forth the same principles in their prefaces. Professional experience and training is left for industry and the engineering graduate to accomplish. Craft, trade and technical skills are more and more being assigned to post-high school level training in technical schools.<sup>16</sup>

Selection of students has become possible to an ever increasing degree of refinement through overflowing numbers of applications for entrance. This has en-

abled each engineering college to restrict entrants to those whose qualifications predict successful undergraduate work and professional life upon graduation. The type of student who formerly entered college with no observable ability to meet the ultimate requirements of the university is becoming a rarity. For these latter men, the post-secondary technical school is being developed.

Fundamental studies during the first year of the engineering college curriculum are arranged to permit a student the option of electing or changing his major course after the first year. Some universities actively promote the idea of preliminary groundwork, at a liberal arts college followed by two or three years of study at the engineering college.

Elective courses tend to be controlled through the establishing of sequences or stems.<sup>10, 14</sup> This is contrary to the freedom of choice which has characterized elective subjects for several decades. The sequence may be only a few or consist of many courses in such fields as the social studies, business administration or in the purely cultural fields.

### III. The Students

Observations of student desires are difficult to substantiate in the literature. The following, therefore, are subject to error and should be so weighed. Most of the points are the result of conversation with a number of undergraduate students.

Most of the present upperclassmen are ex-service men with a more mature outlook on the future than can normally be expected. Their influence on the younger men is good, and for that reason we can expect the present goals to be sustained for many years to come.

The opportunity to learn through a well conceived training program in the company of their choice is anticipated. They accept, as all engineers must, the necessary unpleasant aspects of metallurgical work, such as heat, fumes, dust and dirt, but want the opportunity and backing to effect improvements on the human as well as the technical sides of their jobs. They will be, as many of us are, discouraged by managerial resistance to the elimination of unnecessary hazards and unpleasant conditions.

They want to work in an organization that is well planned and has properly defined duties under a progressive and thorough management.

They are divided in their opinion as to the size and kind of foundry they would choose. There is no fixed trend towards large or small firms.

They seek work that is humanly satisfying outside of their industrial progress in the form of friends, family and living conditions.

They look forward to making a science out of the many foundry operations that are still cut and try. They seek the fascination in their work that changes it into a game.

They believe that the foundry industry is an engineering frontier in that it is the largest business in the nation, which is in need of and seek an engineering type of management and production.

### Conclusions

#### I. General:

Industry generally divides its employees into classifications along the following lines:

- a. Executive, management
- b. Staff, technical, sales
- c. Supervision
- d. Skilled mechanics
- e. Operators and semi-skilled
- f. Laborers and helpers

Preparation for the lower three brackets does not require college level education. For the upper three, a college education is a desirable shortcut to greater efficiency and broader knowledge of an industrial enterprise and better preparation for managerial duties.

Regular university level education does not fully prepare a graduate for industrial or business management or even for technical proficiency. The graduate is grounded in fundamentals and an understanding of science, which is a pathway to greater knowledge. The top achievement of this graduate is to have thoroughly developed his powers of ingenuity, backed by basic knowledge, to the end that his intelligence can be efficiently put to constructive use.

Preparation for managerial and staff responsibilities must necessarily go beyond education. A student has, at graduation, invested heavily in that phase of his preparation. The training in present techniques and procedures is by precedent the responsibility of industry. Customarily, a scheduled training program is the implement by which this is achieved. The student bears a secondary responsibility in that he continues to study those subjects which his formal training did not provide and he accepts, during his training, a salary usually not more than that paid an operator or semi-skilled workman.

University officials have evidenced both interest and cooperation in regard to formulation of a foundry educational program. The prior activities of the various societies in the industry and the implementing of their wishes through the Foundry Educational Foundation have served to implant the aura of a well organized business among students and faculty alike in regard to the foundry companies. It is this unified approach which the universities recognize in contrast to individual wishes of separate companies.

#### II. The Foundry Industry

The varied operations which characterize the foundry industry involve several different kinds of engineering. No four or five-year collegiate curriculum can accomplish the necessary education which a so-called "foundry engineer" would require. There is a field of "foundry engineering," but it will be served by all kinds of engineers. More common among them will be "metallurgical, chemical, mechanical and industrial engineering plus considerable opportunity for engineering administration" graduates, as well as need for some electrical engineers.

The desires of the foundry industry as expressed through the concept of the Foundry Educational Foundation, the personal views of individual members and expressions in the literature which reflect the experience of the leadership up to this time, can be

satisfied with very little in the way of supplementary courses. This is largely due to the development of desirable curriculums which have been effected in most universities by alert educators profiting through the successes and failures of the past.

The result is a desirable alteration of engineering courses to provide fundamental studies in other fields to create a more adequately balanced personality in graduate engineers. In turn, their chances for business and managerial success are thereby enhanced.

These fields have been loosely identified in the literature. For the purposes of this paper we suggest the following definitions:

*Social Humanics studies*—develop an appreciation and understanding of the accomplishments of others, and provide the path for pleasant living outside of, as well as within, a chosen profession.

*Engineering*—is the application of science to the design, development, and construction of useful structures and articles. It is essentially a study of concrete facts established through formula or experience.

*Industrial Administration*—is the development of the understanding necessary to create a conscientious response to the importance of all phases of management in the success of a company.

*Human Engineering*—is engineering design with reference to man's anatomical, physiological and psychological capabilities and limitations. Science is here applied to the task of making human living as perfect as possible by maintaining his mental and physiological health through elimination of adverse conditions as far as is practical.

Table I demonstrates these educational stems by offering some, not all, of the subjects in each field which are of interest to foundry management. Most of the courses shown are available in engineering colleges or in other schools of the same university. Some of the subject material may be included in other courses or regrouped under a different title. Choice of these subjects as electives, if they are not part of the regular course, will tend to satisfy the needs of the industry as previously expressed.

TABLE I—EDUCATIONAL STEMS OF INTEREST TO  
FOUNDRY MANAGEMENT  
SOCIAL HUMANICS STEM

- 1001—Composition and Literature  
Fundamentals of grammar, expository writing, reading, value and appreciation of verse and prose. Expression through proper choice of words and construction.
- 1002—General Psychology  
Individual differences, intelligence, and emotions. Study of personality. Memory and learning. Motivating media. Application to industrial management.
- 1003—History  
World history of the twentieth century and up to present time. Background for economic, political and social structure of the modern world.
- 1004—History of Science  
Development of present day scientific knowledge from early history to current achievements. Includes all phases of science.
- 1005—U. S. Government  
A study of the political background and legislative methods of present day governments. Includes local, state and federal administrations.

#### 1006—Public Speaking

Presentation for committees and management groups. Technical and educational lecturing. Platform presence, audience interest and acceptance.

#### ENGINEERING STEM

- 2001—Fundamentals of the Casting Process (2)  
A course designed for study by all engineering students giving fundamental operations and striving for appreciation of the foundry as a metal forming process.
- 2002-3—Foundry Technology I-II  
A study of the application of science to foundry operations. The prior engineering fundamentals are focused upon the problems and operations of the foundry. Includes control of materials and process, maintenance of tolerances, metallurgical inspection and repair.
- 2004—Casting Design  
A study of patterns, molds, dies and casting design in their relationship to each other in the engineering of a foundry product. Emphasis on design of casting for metallurgical perfection.
- 2005—Metallurgical Calculation  
Calculations to determine requirements for metallurgical reactions in manufacturing operations.
- 2006—Survey of Process Metallurgy  
Appreciation course. Raw material production from mine to plant. Stresses condition and relative purity as same affect subsequent use.
- 2007—Ferrous Alloys  
Advanced study of alloy steels and cast iron (metallurgy, metallography, heat treatment, testing).
- 2009—Non-Ferrous Alloys  
Advanced study of non-ferrous alloys.
- 2010—Engineering Economy  
Selection of process or production design for greatest economy, quality, or both, as applied to a given product.

#### INDUSTRIAL ADMINISTRATION STEM

- 3001—Fundamentals of Economics  
Theory of trade, credit, business cycles and motivating influences. Present world economics.
- 3002-3—Fundamentals of Accounting I-II  
Use of journals, ledgers, simple financial statements. Elements of cost and cost control. Use and need for various company records. Advanced courses cover overhead administration, corporate financial structure and dynamic use of cost data.
- 3004—Business Law. Survey.  
Legal responsibilities and authority of management. Rights of ownership. Elements of contracts and kinds. Relationship to agencies and carriers. Effect of local, state and federal regulations.
- 3005—Fundamentals of Industrial Organization  
Organization of present day corporate and business management. Industrial trends. Required controls, problems of selling and marketing, purchasing and stores, manufacturing and research, distribution and costs, financing.
- 3006—Corporation Finance  
Financing of corporate or partnership enterprise. Regulations due to local, state and federal laws. Administration of income, financing of going business.
- 3007—Marketing  
Market product research, distribution channels, pricing and branding. The market structure and regulations affecting sales and distribution. Relation of sales to other departments.

#### HUMAN ENGINEERING STEM

- 4001—General Biology  
Fundamentals of plant and animal life, emphasizing man's place in this world.
- 4002—Advanced Psychology  
Human reaction to favorable and adverse stimuli. Study of resulting physiological and mental effects. Group psychology.
- 4003—Industrial Relations  
Theory of employee relations and policies. Selection, training, promotion, health and discharge technique. Collective



bargaining, labor laws, safety codes, morale factors, job evaluation and wage systems are studied in relation to each other.

#### 4004—Technobiology

Principles of anatomy, physiology, psychology and health applied to job specification, machine design and facilities of the plant.

#### 4005—Motion and Time Study

Fundamentals of motion economy and time factors for human and mechanical motion. Technique of observation, charts, graphs and formulas are practiced. Relationships to other factors of industrial management is stressed.

#### 4006—Safety Engineering

Industrial safe practices. Engineering for prevention. Psychological factors of acceptance and maintenance of a safety conscious work force. Human and economic factors of safe management.

#### 4007—Techniques of Executive Control

Nature of supervision. Responsibilities of the line of authority. Stimulation of subordinates, contact with superiors and maintenance of support by associates.

### III. The Universities

University level engineering education has persisted in its efforts to restrict the major portion of the curriculums to studies of fundamental science. Subjects which have a narrow field of application are deleted and, conversely, those which have wide uses are considered and developed.

The present tendency is to expand the humanistic studies at the expense of specialized engineering courses in the expectation of a more desirable personality among engineering graduates.

University officials and staff are emphatic in their efforts to provide education which will offer the broadest opportunities to their graduates by not restricting the employment potential to a narrow field. They repeat that training for a specific industry or company is the task of that industry or company. Education is a part of the preparation needed to qualify a graduate for the responsibilities of management.

Universities offering a full program of engineering education (including metallurgical) can meet the needs of the foundry industry with very little modification of present courses. Where sufficient elective hours are available, any engineering course can be adapted through the guided selection of proper subjects. Usually, refinement of the general foundry course for all engineers is desirable. The addition of a few courses to bring about a focus of the student's fundamental knowledge towards foundry technology will complete the picture. The presence in the area of an active and progressive foundry industry will supplement this education through direct observation by students during inspection trips.

Courses in the same field of engineering will have slight variations from one university to another. Each is under the influence of the history of the particular school. Thus, a metallurgical engineer at one may lean towards the chemical phase, while in another it might be more towards mechanical engineering. A study of the course outlines which the individual student has taken will serve as an excellent guide to the fullest utilization of his knowledge. Furthermore, a variety of metallurgical engineers from various universities in a company will spread the available knowledge over a wider field than a staff made up of the graduates of just one college.

### Addenda

This paper presents several problems which may well be answered by papers upon specific subjects. In Table 1 some expression as to context of advanced foundry courses (2002-3) is needed. Course No. 2004 can be broken up into several units, each of which requires original development of subject matter. Casting design and the related production tools is a "must" if the industry is to lead itself into new fields of application.

A paper or magazine article outlining the specific positions within the industry and the opportunities offered in each, might well be followed by other papers outlining specific engineering courses and electives which would adequately prepare a prospective member of the industry for that particular niche.

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### DISCUSSION

Chairman: W. H. RUTEN, Polytechnic Institute of Brooklyn, Brooklyn

Co-Chairman: L. G. PROBST, National Engineering Co., Chicago

M. O. WITHEY (*Written Discussion*):<sup>1</sup> It is a source of gratification to observe the increasing interest of foundry industrialists in obtaining well-trained technical personnel to improve the quality of their products. The impact of the Foundry Educational Foundation should be sufficiently pronounced to insure that the objectives sought will be in the near future obtained. The past decade has witnessed visible evidence of a markedly increased interest in providing foundries with better facilities for the workmen and improved methods of housekeeping in

<sup>1</sup> Dean, College of Engineering, University of Wisconsin, Madison, Wis.

general. The attention now being given to the provision of better technical personnel is in logical sequence with such movements. It is also demanded by the competition which foundry products meet in the market. In Wisconsin we are most anxious to see that adequately trained personnel is provided to meet the demands of our numerous and expanding foundry industries. Therefore, at the U. of W. we shall do all that we can to forward this educational program.

As evidence of our interest in the foundry program, it should be noted that an increase in the space and facilities have been provided by remodelling the structure at the rear of our Mining and Metallurgy Laboratory. This structure will make available for the foundry a ground floor area of approximately 6400 sq ft. The main portion of this space, 126 by 41 ft, and its appended ells, is well arranged to accommodate the modern equipment which has been provided. During the past year about \$7500 has been spent on new foundry equipment. Approximately \$5000 of this sum was provided by the Foundry Educational Foundation. Among the new items of equipment are two automatic molding machines, an air weigher for cupola control, and a cleaning machine of the shot peening type. There has been installed a sand testing laboratory as a part of the foundry equipment. The building walls and floors have been well painted to facilitate cleanliness and good housekeeping in operating this laboratory.

The Foundry Educational Foundation is to be commended in its method of stimulating interest in its educational program through the provision of scholarships, the listing of open positions, and the listing of student personnel available for positions. These procedures will do much to enlist young men for careers in the foundry. It seems to the writer that one of the most satisfactory methods of securing qualified and interested personnel will be through opportunities afforded by summer employment for students. By such employment an initial period of acquaintance is provided both to the student and his prospective employer which will give valuable information without committing either and which, under the circumstances, will permit the termination of employment without embarrassment to either party.

It is our intent to keep our metallurgical curriculum in a sufficiently fluid state to provide for changes in electives and required courses which are most desirable from the viewpoint of the foundry program. At present the curriculum contains 22 elective credits; 6 in the sophomore year; 10 in the junior year; and 6 in the senior year. In considering the choice of non-technical electives, the student's attention is brought to the available subjects in economics, philosophy, psychology, speech and commerce. Thus we try to impress upon the students the desirability of obtaining a broad education. We also urge students to seriously consider the possibility of spending more than four years in college in order that they may receive a broad education.

G. J. BARKER (*Written Discussion*):<sup>3</sup> Mr. Dreher is to be complimented upon his excellent presentation of the relationship between engineering education and industry. His work with the Foundry Educational Foundation has given him an excellent opportunity to analyze the educational problem from both the educational and the industrial viewpoint. His analysis as presented in this paper shows how completely he has grasped the conditions as existing. During the past year there have been made significant advances in bringing closer together industry and education. We expect and hope greater progress will be made in the next two years.

The author has stated very well the position the university officials have taken regarding training the young engineer. We do insist on a thorough training in basic fundamentals on a broad basis and from this firm foundation the foundry industry can secure the type of engineers who they must further train for executive positions. I repeat again the premises that the foundry industry must not expect the universities to completely train the young men for positions in a particular phase of industry. The training must be completed in the individual plant and, if this is to be successful and the company desires to obtain men of executive ability, then it must inaugurate a thorough training program. Some few companies do have this program but, in my opinion, the lack of a proper training program in so many of the large companies is the reason more engineers are not attracted to your industry.

<sup>3</sup> Chairman, Dept. of Metallurgy, University of Wisconsin, Madison, Wis.

Industry as a whole is fortunate in having the large number of engineering schools from which to draw its technical personnel. I can assure you the engineering educational profession is not and will not become dormant. The American Society for Engineering Education is a live, aggressive organization. Its members are constantly sifting and, when advisable, installing new methods of education. We must be, and are, constantly on guard against installing in our educational system some radical idea of new course content. Unless we are constantly evaluating our courses and their content we could easily graduate so-called engineers who would prove a failure in our industry.

Recently a survey has been completed which now provides us with definite information regarding the number of technical graduates who will enter your industry. There were 175 colleges and universities contacted and out of the 105 replying only 40 are giving some type of foundry instruction to only 7000 students. Of these 83 per cent are mechanical, 4.3 per cent electrical, 4.0 per cent chemical, 4.0 per cent metallurgical, and the balance civil or industrial engineers. A recent survey indicates there are 244,390 students enrolled in our engineering schools. About 52,000 will graduate this year. Of this large number of graduates our survey indicates only 120 men will enter the foundry industry. This indicates that less than 3 per cent of the casting companies will secure technical graduates this year.

There is a brighter side of the picture. Our survey shows that many schools are interested in the foundry industry and will develop more foundry courses if industry indicates a real desire for technically trained men. The splendid work Mr. Dreher is now doing is a step in the right direction, and all of us should support his efforts wholeheartedly if we are interested in the foundry industry.

F. B. SKEATES:<sup>2</sup> I operate a foundry and feel that somewhere in this educational plan a provision must be made whereby a foundry can train a high grade apprentice under one of these scholarships. I am in favor of the high-trained student. But the foundry industry has been accused too much of lack of interest in training foremen. We do the best we can with the type we have. I reiterate that there must be a place where the foundry can train a high grade apprentice. After completing his training he should be entitled to a scholarship to one of these colleges. Then when he comes back you will have one of the finest foremen in your foundry, and the entire supervision in that shop will be elevated to a higher level by the influence of this one foreman.

Of course we need engineers, salesmen, managers, etc., but first we need good foremen. We are doing the best we can in the foundry in developing these men, but they lack about four years of higher education.

What is a foreman called on today to do? A few years ago the best foreman in your foundry was the man who produced the greatest tonnage. What does he have to do today in the shop? He has to know every paragraph of every labor law that was ever written. He has to know costs, sanitation, human relations, etc., and operate his department on a limited budget.

I think I have a solution to this problem. The foundries are generally divided into districts. Suppose you have a dozen foundries in a certain district. Take the top apprentice in each foundry in that district and put him through a rigid test, qualify the best boy in that group, and send him to school for four years.

A. W. GREGG:<sup>4</sup> I have always been interested in getting technical men into the foundry industry and have had a good deal of experience with it. In my opinion the foundries are entirely at fault for a poor job of selling the foundry industry to the college graduate.

Mr. Bruce Simpson of National Engineering Co., just a short time ago, submitted a questionnaire to the engineering graduates of a well known midwest university. He asked them many leading questions such as, "How much do you think you are going to be worth when you graduate to any employer?" "Would you rather work for a company that has 10,000 men, 1,000 men or 500 men?" His last question was the punch question. He asked, "Regardless of how you answered the preceding questions, would you consider the foundry as a life job, as a career?" The answer of 94 per cent of the boys was "Absolutely not" and gave as their reasons the dirty and unsanitary conditions in the average foundry.

<sup>2</sup> Link Belt Co., Chicago

<sup>4</sup> Whiting Corp., Harvey, Ill.

Now, what does that mean? That means that those boys did not know conditions in today's foundries. They have visited foundries that were built in their grandfather's days and which have not progressed at all. I think the average man on the street has the same idea of the foundry industry. The foundry certainly needs technical men; it is an engineering job today.

I wish to comment on Mr. Skeate's idea. Mr. Skeates is suggesting that you take a boy who has gone through an apprenticeship and send him to college for an engineering education. I think Prof. Barker of the U. of Wis. can say something about that. They are doing something like that in Milwaukee now. They are picking out apprentices and sending them to the U. of Wis. Prof. Barker is very fortunate in getting cooperation from the Milwaukee Chapter of A.F.S. I think he is getting better cooperation today than any university in the country. The Milwaukee Chapter sends speakers to U. of Wis. in Madison to talk to the boys. They have a fine group of boys at the university. Prof. Barker told me just before I spoke before his group that they were too shy to ask questions. But I had more questions out of those boys than from the average foundry meeting that I go to.

PROF. BARKER: I will be very glad to explain our plan. I shall make it brief because I would like to answer Mr. Skeates. I believe he has something in his idea. These two things are tied together and perhaps I should talk about Mr. Skeates' idea first.

You have two points in this discussion. The university is trying to do something and the plant management is trying to do something. We are both working towards the same end. In the university we graduate a man and we send him to you. This man does not expect to go to top management immediately. We have not trained him to go into top management at once. He does not know the details of a foundry. We expect you, in industry, to take this man and put him through your foundry training course, and give him a year or two years plant training.

You have a man there who has built the foundation. You can send him through your program faster than you can send through any of your apprentices because he has a foundation. In two years he will make you an excellent foreman. But are you going to lead him up to something?

On the other hand, let us go back to the foundry. You have trained a man as an apprentice, you believe he should go into the university. But frequently you cannot bring the man into the university unless he is a high school graduate and has the entrance requirements which are set up in the universities. The registrar sets some qualifications and the man has to meet all of these before he can enter the university. After he is admitted we will give him a good training. And I agree with Mr. Skeates thoroughly in that when he comes back to you he will make you a good foreman. But you will not keep him as a foreman. He will make you a good foreman, but unless you have something higher than a foremanship for him, you will not keep him longer than one or two years. You must have something ahead for that man, because with his educational training, he will never be satisfied to stay as a foreman all his life. I believe we might just as well recognize that fact. These men are technically trained. There are thousands of men being sought after each year, and if you do not take him in the foundry industry, he has training and can go somewhere else. We are not training that man in the university for the foundry alone. General Electric, Westinghouse—any of these will take our graduates, the men we are trying to give you in the foundry industry.

I would like to give you just an idea of what we are trying to do at Wisconsin on this cooperation plan Mr. Gregg mentioned. The U. of Wis. has established some fellowships. We are paying a young man \$75 or \$80 a month. That is not much, but it is a pure gift to him. We say to the local Chapter, "You select from your apprentices in the plants the man you would like to propose." They turn the selection over to the Educational Committee of the Chapter. They select the apprentice they would like to have us accept, and if he can meet the entrance requirements he enters the university. We give him a scholarship for four years, and when he graduates he is going back to the foundry industry. He came from the foundry, and he is going back to the foundry, I can assure you. The university provides this scholarship, and it should not be confused with the F.E.F. scholarships.

MR. GREGG: Does the F.E.F. also give aid to that student? How is that working out?

PROF. BARKER: No, F.E.F. does not help that same man. If the university is giving him a scholarship, that man cannot get an F.E.F. scholarship also.

MR. GREGG: Have you had any boys from industry that went through your course at the U. of Wis.? How are they working out? I am interested in the results.

PROF. BARKER: We have not been running long enough to get them all the way through. We had one man who graduated and he went in the Navy. He is still in the Navy. We will have more men graduating soon.

G. G. GOTT: When I asked fellow students at MIT what they thought of foundry work they replied emphatically in the negative. One reason for this attitude is that, at least a year or so ago, they were getting more money elsewhere than the foundry industry was willing to pay. Others did not know about conditions in present-day foundries. So we tried to do a selling job. This summer we will have about 20 students working in the foundry and a number of our graduates are going into foundry work.

There is also a cooperative program at Northeastern University. There are a number of students working under their cooperative program and the school cannot get more students for the foundry program for the reasons just given. Under this program the boys go to school a number of months, then they go into the foundry industry for a number of months.

I worked as a substitute foreman through the vacation period in Boston last summer and I became acquainted with a number of those boys working under this cooperative plan. I started at the low rate of \$.92 per hour and after a month the rate was increased to \$.97 per hour. The boys who were working under the cooperative course started at the rate of only \$.65 per hour. You cannot get the boys to work for such a low rate of pay. They cannot live on that, let alone save money to go back to school on. That was the trouble with the cooperative course at Northeastern. The foundries around there were not cooperating enough.

T. M. RUTTER: Who is eligible for these scholarships sponsored by the F.E.F.?

MR. DREHER: Briefly, the scholarships are granted entirely at the discretion of the university in which they are set up. For example, at the U. of Wis., the university sets the policy and makes the selection. The F.E.F. provides the financial aid. The boys eligible at the U. of Wis., for example, include most everybody in the engineering school who has evidenced an interest in the foundry and who is willing to take, or who will commit himself to take the electives which U. of Wis. has set up in their foundry program.

Essentially the same is true at all the schools where F.E.F. scholarships are given. In most of these universities the selection of scholarship candidates is not left to the professor alone. They have a committee made up of men in the department in which F.E.F. scholarships are operating. At U. of Wis. they are operating in the Metallurgical Department; the same is true at the U. of Cincinnati. At Northwestern it is the Industrial Engineering Department while at Cornell it is the Metallurgical Department. These departments steer the program in the college, but students from other departments can come under this program.

MR. RUTTER: Are you doing anything along the lines suggested by Mr. Skeates?

MR. DREHER: Nothing as formal as Mr. Skeates has suggested. I have personally talked to several apprentices who are going into the program. I have a young man, 33 years of age, who was earning \$2.40 per hour as a pattern maker for years. He is going into Case Institute next Fall under this program. I know that at Ohio State U. there is one graduate this year who was a graduate apprentice molder before he went to the university.

At the cooperative schools if a student continues his foundry education he can pretty well carry himself along if his employer will aid him financially in one form or another during the first year. In the meantime they are eligible for F.E.F. scholarships at those particular schools.

<sup>5</sup> Massachusetts Institute of Technology, Cambridge, Mass.

<sup>6</sup> Richmond Radiator Co., Uniontown, Pa.



Selection of F.E.F. scholarship candidates is a big job. At one of the schools this past year there were twice as many applicants as available scholarships. The school considered the grades of these candidates and classified them in three groups, first, second and third. Then they considered eight or ten character points, such as leadership, personal appearance, speaking ability, etc. Each member of the faculty in that department rated these boys in the first, second or third class. They did not rate them by points numerically or by letter. In this manner they determined who was to get the scholarships.

MR. RUTTER: Do you have any plans for the future to follow along the lines suggested by Mr. Skeates?

MR. DREHER: The future of the Foundry Educational Foundation is in the formative stage. It is criticism such as this that will help formulate its future. There are no specific plans for the future of this Foundation. We are in a period of development. It is almost an experiment if you want to call it that. We are in it to develop and perhaps at the end of this coming year we should have a program ready to launch so that when this first three years is over we will be well underway.

W. W. MOORE:<sup>7</sup> The statement was made that these college-trained men would not be content to remain as foremen. I do not believe that is correct, from the monetary standpoint. I feel that the right man is of as much value to the foundry industry in his particular capacity as to any industry he might go.

PROF. BARKER: I believe you misunderstood me. I did not mean to say that everyone of these graduates is going to be dissatisfied with being foremen. I say that you should not go to all the trouble of training these men and expect them to remain as foremen the rest of their lives. They will naturally want to work up to something higher than foremen. I believe this program has been built to train for executive positions. We will give you foremen. It will be up to you to keep them as foremen. You will have to make the conditions so attractive they will stay there as foremen.

MR. MOORE: That is true, but perhaps we are putting a limit as to what a foreman is and what a foreman can do. A foreman in my estimation can develop his particular department to as high a level as his capacity will permit.

<sup>7</sup> Burnside Steel Foundry Co., Chicago

DAVID MACINTOSH:<sup>8</sup> I have been in the foundry industry for about 58 years now. I would like to ask what subjects the colleges and universities would teach a young man if he wanted to be a foundryman. What would be his main subject and his curriculum? There are many things to be considered. We are in a business as old as time, and I think we are losing the art for all the boys in the shop. What subjects would we teach that a boy can use to advantage in a foundry?

MR. DREHER: Each of the universities has a curriculum that can be reviewed and this paper outlines a few of the things that are suggested, some of which are now being taught. The universities will not teach the art of molding or the art of the foundry craft.

MR. MACINTOSH: I think that is a mistake.

MR. DREHER: The colleges will teach the fundamentals. A boy at the Brooklyn Navy Yard with only two years of college training in three months in the foundry at the navy yard was molding 20,000-lb propeller castings. That is supposed to be one of the most advanced fields of the molding art. This man picked it up so fast that it amazed everybody in that shop. The idea of the university education is to get the man well grounded in the fundamentals so that he understands.

We cannot answer the question completely because you have brought up another matter in this educational program that still has to be worked on. But you are right; something has to be provided to maintain the art because unless the art is maintained we lose some of the field of foundry products. We have lost innumerable chamber jobs, autoclaves and covers and bodies of that type that used to be made by the old time molder who needed nothing but a blueprint to start with.

MR. MACINTOSH: I think some of us in the foundry office today belittle the craft, skill and experience of the men in the shop. I have come through the core room to every branch of the trade. I have been in the foundry since I was 14 years of age. And whether a boy or young man remains a foreman will depend on his ambitions in life. If he has no ambition, he will be stationary. If he has ambition, he will go upwards. But when he gets to the top, he must not belittle the skill and experience of the craft because there is skill required in the foundry. I have made castings from  $\frac{1}{4}$  lb to 40 tons, and it requires skill and experience in the shop to do that.

<sup>8</sup> Sachs-Barlow Foundries, Inc., Newark

# REFRACTORIES USED IN STEEL AND IRON FOUNDRIES

By  
W. H. Owen \*

USE OF REFRACTORIES in the foundry presents an overgrowing list of applications requiring an increasing number of diversified refractory materials. In this discussion, we will review briefly the well-known uses of refractories in the foundry and consider certain special refractories which have more recently come into general use.

It is interesting to observe how the use of refractory material in the foundry has been extended by following the possible applications of refractories in the various steps which the molten metal takes. Some of these applications have long been standard practice; others are so new that they are just beginning to prove their value. First, refractories are requisite in the melting and refining equipment employed in the production of metal. Again, they are needed for the transport equipment used in moving molten metal within the foundry. Just within the past two years refractory materials have been used within the mold which forms the casting. This application is growing rapidly.

## Refractory Applications

In the melting and refining units, whether they be the open hearth furnaces, the electric furnace, or the converter, the refractories employed must serve in the processing of more than one charge. In the electric furnace, for example, it is impossible to generalize on the most suitable refractory, but the overall properties of silica brick, together with its relatively low cost, make its use general in roofs and, to a lesser degree, in sidewall construction. For intermittent operation, super-duty fireclay brick or brick of the high-alumina classes may be used to advantage in roofs, but for hard continuous operation, silica brick generally serve best. Super-duty silica brick are now used advantageously in roofs. Side walls are for the most part constructed of silica brick or basic brick, and the choice, in the base of basic steel practice, will largely depend upon the overall service of each. Considerable progress has been made in the development of bottom materials for furnaces. In the case of the acid electric furnace, careful attention to proportioning of clay and ganister, as well as sizing of the ganister, permits increased bottom

life. In the case of basic bottoms, the introduction of the rammed magnesite materials has afforded a means of rapid installation of the furnace bottoms and has brought about improved bottom life. The length of time during which the strength of the refractories must be maintained at high temperatures in the melting furnaces calls for much thicker sections than are found to be necessary in the single-use type of refractory shapes used later in the casting cycle.

In the iron foundry, the cupola is, of course, the furnace used for melting and refining the metal. The materials most commonly used for cupola linings are fireclay refractories manufactured especially for the purpose. Patching of linings is accomplished by using mixtures of fireclay and grog, in suitable proportions, properly tempered with water, and rammed against fireclay brick backing. Other refractories which are being used are super-duty fireclay brick and to a lesser extent silica brick. The most important of the properties desired for cupola blocks used in the melting zone are refractoriness, namely ability to withstand high temperature, and density, which affects their resistance to slag penetration, to spalling, and to abrasion. The life of the lining depends to a considerable degree upon the nature of the charge, the temperature of melting, the duration of the heat, and especially upon the composition and quantity of the slag.

## Ladle Linings

Transportation of molten metal in ladles from the furnace to the mold requires the use of refractory materials. The linings of the larger ladles are constructed of fireclay brick. Continued work is being done to adapt standard brick sizes to the larger ladles, particularly those of the receiving and holding type, so as to eliminate the use of the more costly special shapes. Smaller ladles more frequently are lined of rammed material forming a monolithic lining. In acid practice, where ganister-clay mixes are rammed to form monolithic linings, the same care in preparation and sizing should hold for the ladle as for the furnace linings and bottoms. Deterioration of ladle linings results from the fluxing action of the oxides of the metal with the refractory. In this application the length of time during which the strength of the refractories must be maintained calls for thick sections, but not to the

\* Harbison-Walker Refractories Co., Pittsburgh.

extent of that demanded in furnace use.

The operational sequence in the foundry has now brought us to the point for consideration of the refractories used in the mold. We have considered the refractories from which the longest possible life is desired and now we turn our attention to those special refractories which are not re-usable. These are called special refractories, not only because of their compositions, but also because of their physical shape, and particularly because their sectional thickness, makes them so to the manufacturer. Gate and runner brick, breaker core shapes for necked-down risers, and cup and gate strainer core shapes comprise some of these refractories which should be so classified.

Foundrymen have long known that a successfully made casting must be properly gated. The primary purpose of the gate system is to furnish a means of allowing the metal to enter the mold. However, there are other functions that a gate can and must perform. Briggs in his *Metallurgy of Steel Castings* ably summarizes the functions of the gate as follows: (1) to fill the mold cavity completely, (2) to avoid erosion of mold and gate by liquid metal, and (3) to establish proper temperature differences in the mold so that the gate will not cause the occurrence of shrinkage cavities in the casting. He adds, further, that the gate may function also as a venting system for the removal of gases from the mold and as a riser to furnish liquid metal to the casting.

#### Use of Special Refractories

To accomplish the first two purposes mentioned above, gate and runner brick are used in many foundries producing heavy and medium weight castings. Gate brick are known in the refractories industry as the shape forming the sprue, and runner brick as the shape forming the in-gate to the casting. Standard gate and runner brick are available in bores from 1½ in. to 4 in. in lengths of from 2 in. to 15 in. Their use will keep mold erosion to the minimum.

The foundry industry has for several years employed in the production of castings an operation-saving method of forming a short neck (a reduction in cross-sectional area) between the riser and the casting. This practice made riser-head removal easier and at the same time left only a small stub of metal on the casting surface to be ground off. The reduction in the number of square inches of metal to be cut and ground reflected directly in decreased cleaning costs.

After a thorough study, refractory breaker core shapes of a standardized design have been developed. These flat, stock, refractory cores are square in shapes, permitting the use of either square or round riser bodies. The neck opening carries a reverse taper to provide a suitable notching effect for ease of riser removal. The amount of neck stub remaining after breaking or cutting off the riser is controlled by the distance of the notch from the casting surface. Although the thickness of the standard core increases with the riser diameter, a neck stub no greater than a quarter inch is left to be ground on cores 5/8 in. to 1½ in. in thickness.

Reference has been made above to the necessity of maintaining strength at high temperature of the refractories used in the furnace, in the ladle, and in gate brick. In the application of refractories to necked-

down riser cores the degree of sensitiveness with which temperature rise affected the stability of the shape was quickly revealed. This sensitiveness is, no doubt, accentuated by the fact that heat energy is transferred from the metal to the core, through its top and bottom surfaces simultaneously. It has been calculated that, with equal thicknesses, it would take only one-fourth the time to heat the core to a selected mean temperature through two opposite faces, as compared to heating it through one face only.

On the other hand, the thickness of the cores must maintain a definite relationship to the diameter of the neck, otherwise, complete freezing of the neck zone will take place before the casting itself has completely solidified.

#### Use of Refractory Cores

Refractory breaker core shapes are power pressed from a clay mixture which is of a high duty class. The material is particularly prepared with the objective of permitting the core to withstand high thermal shock. Unusual tests have been made where molten metal was poured directly on the refractory core shapes without any opportunity of preheating the core. The results were very satisfactory both with open hearth and electric furnace steels. However, experience indicates one may provide conditions which will crack refractory cores quite early in their heating cycle, even before the metal freezes, and cause veining on the surface of the casting. It was found that when both the riser diameter and the core area in contact with the casting were small in relation to the thickness of the necked-down riser core used, temperature differences within the core were so great that thermal stresses generated within the core section soon cracked it. By reducing the core thickness the early cracking of the core disappeared.

Refractory cores come ready for use. This means that the preparation of core boxes, of sand, and the use of core binders and molders' time will be saved in the production of the necked-down riser cores, which now are made in the foundry. It means the elimination of baking and release of oven equipment for the curing of other cores of special contour and shape which may be required. It means reduction in handling, inspecting, and repairing of sand cores which now are used.

Because refractory cores are less porous than sand cores, they are not likely to absorb as much moisture within a given period of time. However, it is not recommended that wet refractory cores be used, because of the risk of blows. We recommend that the same practice as to the dryness of core be used as that employed in the use of stopper rod sleeves which, too, are submerged in molten metal.

Refractory cores are burned at an adequately high temperature to provide a strong ceramic bond. In the burning process all volatile matter or gas-producing elements are removed. Those who have had an opportunity of using refractory cores report that they evolve little or no gas.

Recently, much interest has developed in the use of another type of refractory core, that is, cup and gate strainer core shapes. These cores act as a choke in the gate system, so as to float undesirable materials lighter than the metal, in a reservoir accumulating above the core, thus permitting clean metal to enter



the mold. The interest and consideration of the A.F.A. Sand Division Core Test Committee has been solicited concerning a proposal toward standardization of these shapes. A range of core sizes predicated on a range in the rate of delivery of metal through the core to the mold was submitted for their study and suggestions.

Should it be found that enough similarity in design and in a range of sizes is common among the users of strainer core shapes, we believe that a certain degree of standardization may be accomplished which will provide a basis for manufacturing these shapes in refractory material at the lowest possible cost.

### DISCUSSION

*Chairman:* R. H. STONE, Vesuvius Crucible Co., Swissvale, Pittsburgh

*Co-Chairman:* A. S. KLOPF, Western Foundry Co., Chicago

E. C. ZIRZOW:<sup>1</sup> I was somewhat disappointed because the author did not discuss strainer cores more fully. Using refractory materials in strainer cores is not new. About 15 years ago we tried a series of strainer cores made of refractory material but unfortunately the cost was too excessive. However, the author showed me a strainer core which cost only one cent to make. When you bring strainer core production cost that low, that is close to what it costs the ordinary foundry to produce a strainer core, because a strainer core sand mixture is a rather expensive mix. It must have high hot strength. To have that property it is necessary to increase considerably the amount of dry binders in the mixture due to the additives necessary to create the hot strength.

In standardization of strainer cores we must standardize on the shapes. From what little investigating we have done so far we found as many different shapes and sizes in use as there are foundries. Each foundry has its own pet strainer core. That probably is the result of each foundry's own experience. There is really no scientific explanation why they are used. They obtained good results with them before and they will continue to get results with them. That is the general attitude. The changing of prints might sound simple, but when you have a thousand patterns, especially in a large production shop, and when every one of those prints must be changed, it is quite an expensive item. However, you can begin by putting in your new patterns with the new type of strainer core prints. It will mean for a while that you are going to have more than one type of strainer core, but that is another phase of the economics of this strainer core business.

I think the refractory industry has a selling job on their hands. I do not think it will take hold too easily. We know what we need in a strainer core. We need something to hold that hot molten metal back, and that strainer core must hold up. That is the function of the strainer core. I can think of nothing more refractory than a good ceramic material. I understand that some of these refractories can withstand temperatures as high as Cone 32, approximately 3092 F. That is sufficiently high when speaking in terms of malleable and gray iron because strainer cores are not used much in steel casting practice.

H. M. KRANER:<sup>2</sup> Mr. Zirzow mentioned that one of these strainer cores can now be produced for one cent apiece. How big a core would that be?

MR. ZIRZOW: That strainer core is 1 3/4 in. in diameter. It is a fired product. Do you think this type of strainer core has possibilities?

J. A. BOWERS:<sup>3</sup> Foundrymen ought to appreciate what the refractory manufacturers have done and are doing for the foundry industry. There have been several different shapes that have been mentioned but they make many more. For instance, one of the biggest losses that we used to have in the centrifugal shop was from a monolithic ladle lip. When we would cast the pipe, part of the clay would spall off, fall into the mold and, of course, we would lose a pipe casting. The refractory industry now manufactures for us a ladle lip for the different size ladles in the shop. Now we practically never lose a pipe from that cause. In addition to that they make male and female joints for trough tile and also male and female joints for the lip of teapot ladles. All we have to do is to put the mortar in, drop the brick together and the spout is built and it holds without any runouts.

A. H. THOMSON:<sup>4</sup> A refractory strainer core would cause less inclusions in the casting. A strainer made of sand would be likely to disintegrate and get into the casting, whereas a refractory material would withstand intense heat without going to pieces.

CHAIRMAN STONE: Do you know of any extension of this practice in Canadian foundries?

MR. THOMSON: No, I do not.

CO-CHAIRMAN KLOPF: The economics involved in going to this type of strainer core material came up. There are millions of strainer cores and knock-off cores used in foundries every year. Anything put together from sand and binders and baked under ordinary foundry conditions has weaknesses. A properly made refractory strainer core, or knock-off core, or any special shape of that type will have many physical advantages which should result in tremendous economic advantages, in that we will produce clearer castings and have less cleaning costs in the foundry. The future of this type of material looks promising.

D. W. WUERTHELE:<sup>5</sup> Some 15 years ago there was some experimentation done at National Malleable & Steel Castings Co. in Cleveland in making these strainer cores out of core sand, baking them and then dipping them in a high temperature bonding mortar. If my memory serves me correctly, that was quite successful for a while. There may have been some economic reason why that practice was discontinued. I will be interested to hear if there has been any further experimentation along that line and what the results were.

MR. ZIRZOW: Our attempt to dip the sand in a refractory material was a failure and we finally went to this entirely ceramic material. We were successful in using the ceramic material, but the cost was prohibitive at that time.

We make millions of strainer cores for our use. We could standardize on our shapes and sizes and the cost of the strainer core will then become insignificant. I understand from the refractory industry that the biggest cost in the production of this strainer core is in the cost of the die. It is only by standardization that we can make that die cost insignificant.

It might be a case of going to perhaps 12 to 20 different types of cores which should take care of practically all the conditions which we might encounter in pouring molds. We could get almost any metal screened through and any choking that we would want to use. I think it has many possibilities.

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<sup>3</sup> American Cast Iron Pipe Co., Birmingham

<sup>4</sup> Canadian Refractories, Ltd., Montreal, Que., Canada

<sup>5</sup> A. P. Green Fire Brick Co., New York

<sup>1</sup> National Malleable & Steel Castings Co., Cleveland

# EIGHTH ANNUAL REPORT ON THE INVESTIGATION OF PROPERTIES OF STEEL SANDS AT ELEVATED TEMPERATURES

By

J. P. Fraser \* and P. E. Kyle \*\*

## Introduction

THIS REPORT COVERS the work on the Sand Research Project sponsored by the Sand Division of the American Foundrymen's Society which has been completed during the past year at Cornell University. Part I summarizes the data on hot compressive strength vs. test temperature for six sand mixtures. Some of these data were included in the Seventh Annual Report.† Part II includes the results to date on studies of hot compressive strength vs. exposure time for the 4 per cent western bentonite mixture and the 10 per cent fire clay mixture each containing 5 per cent moisture and a base sand of N. J. No. 60. These data are being analyzed in order to establish a recommended procedure for conducting hot compressive strength tests at elevated temperatures. Part III contains a summary of and conclusions from the results reported.

Those associated with the project at Cornell wish to acknowledge the helpful suggestions by the Sand Division Committee on Physical Properties of Steel Foundry Sands at Elevated Temperatures, J. A. Rassenfoss,†† Chairman, and the advice given by Dr. H. Ries, Chairman of the Sand Division.

## PART I

### Hot Compressive Strength vs. Test Temperature For Steel Sand Mixtures

The data partially reported in the Seventh Annual Report have been completed for the six mixtures which were studied during the earlier stages of this project. These tests were run over a wide range of temperatures, which in most cases was from 300 F to 2500 F, and the specimens were all tested after an exposure time of 20 minutes. Figures 1 to 6 inclusive summarize the data.

Figure 1 indicates that the 4 per cent western bentonite mixture containing 5 per cent moisture has a peak strength about 160 psi greater than that containing 3 per cent moisture. The temperature at which peak strength is reached appears to be approximately 1850 F for both. The addition of 1 per cent corn flour to the mixture containing 5 per cent moisture tends to

reduce the peak strength by about 175 psi and appears to lower the temperature at which the peak strength is attained by about 50 F.

Figure 2 shows similar results for the 10 per cent fire clay mixtures. In this case the peak strength of the mixture containing 5 per cent moisture is about 130 psi greater than for that containing 3 per cent moisture. There appears to be only a slight difference in the temperature at which these peak strengths are attained. The addition of 1 per cent corn flour to the mixture containing 5 per cent moisture increases the peak strength by about 200 psi but does not alter the temperature at which this peak strength is attained.

Figures 3, 4 and 5 show the results obtained on mixtures which have only limited use in the steel foundry and which show very low hot compressive strength at all test temperatures. The 4 per cent southern bentonite mixture shows higher strength at 3 per cent moisture than at 5 per cent moisture, while the 4 per cent fire clay and 4 per cent halloysite show the reverse relationship.

Figure 6 indicates that the 4 per cent illite mixture is appreciably stronger at 5 per cent than at 3 per cent moisture and that corn flour additions of 1 per cent do not affect the strength at the 5 per cent moisture level.

## PART II

### Hot Compressive Strength vs. Exposure Time

In order to collect more data having a bearing on the subject of test procedures at elevated temperatures, an exhaustive study is being made of the effect of exposure time on hot compressive strength at several temperatures.

**Mixtures**—In order to develop data which could be put to use as soon as possible by the foundrymen, only two mixtures were studied. These were as follows:

Mixture No. 1	Mixture No. 2
N. J. No. 60 Sand	N. J. No. 60 Sand
4 per cent Western Bentonite	10 per cent Clay (No. 7)
5 per cent Moisture	5 per cent Moisture

The room temperature properties of these mixtures are shown in Figs. 7 and 8. Figure 7 shows two sieve analyses of the sand taken at two different locations in the storage bins and at two different times during the course of the investigation. These two analyses check

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† TRANSACTIONS, A.F.A., vol. 55, p. 607 (1947).

†† Res. Met., Research Laboratories, American Steel Foundries, East Chicago, Ind.

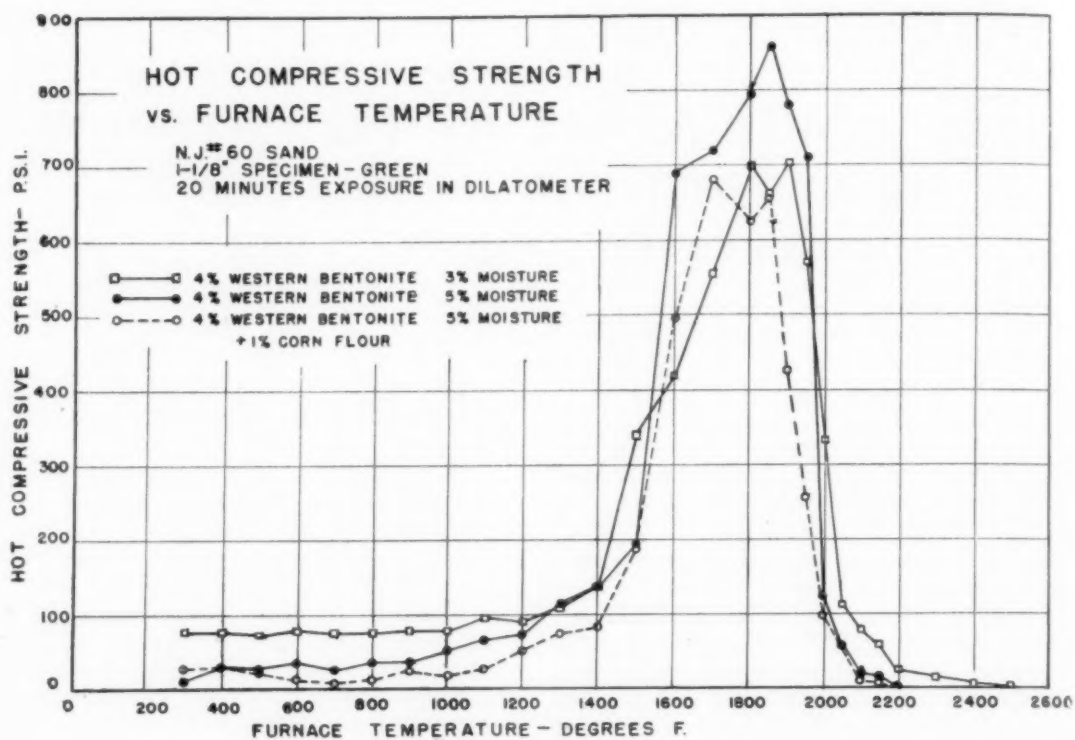


Fig. 1—Hot compressive strength vs. temp. (4% western bentonite mixture).

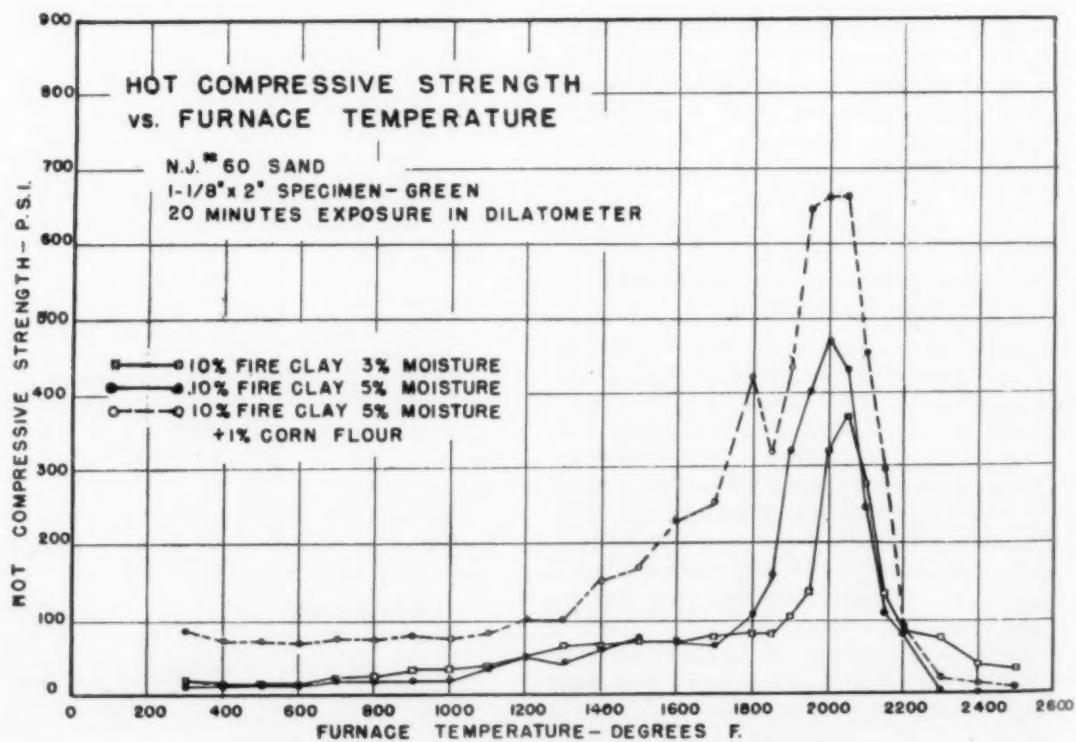


Fig. 2—Hot compressive strength vs. temp. (10% fire clay mixture).



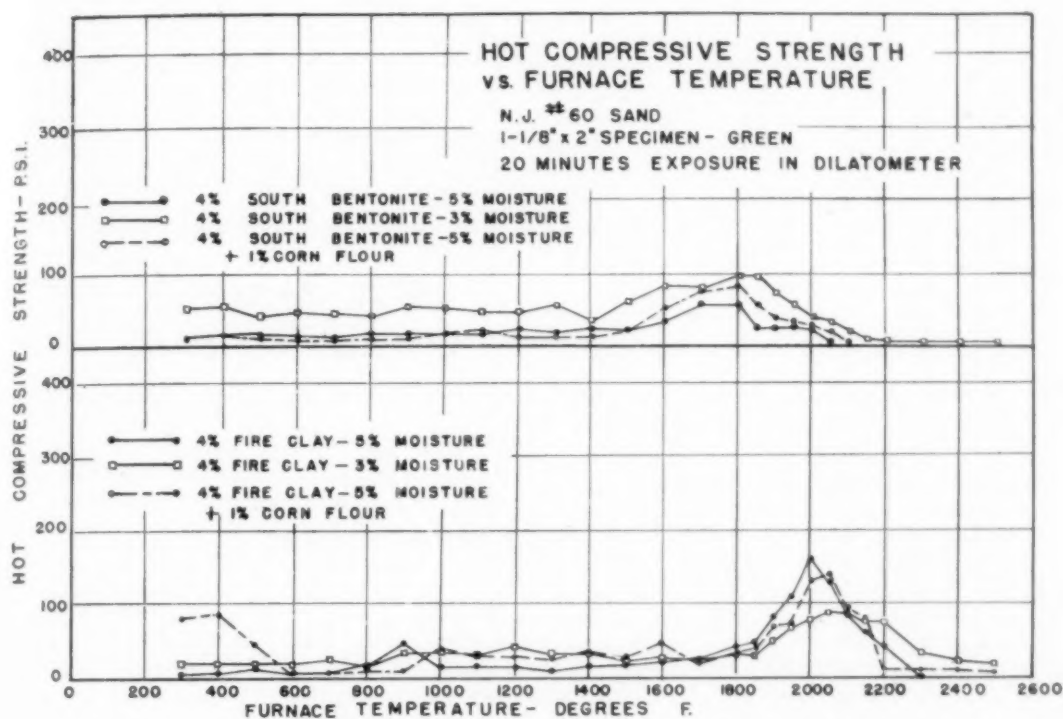


Fig. 3 (Top)—Hot compressive strength vs. temp. (4% southern bentonite mixture). Fig. 4 (Bottom)—Hot compressive strength vs. temp. (4% fire clay mixture).

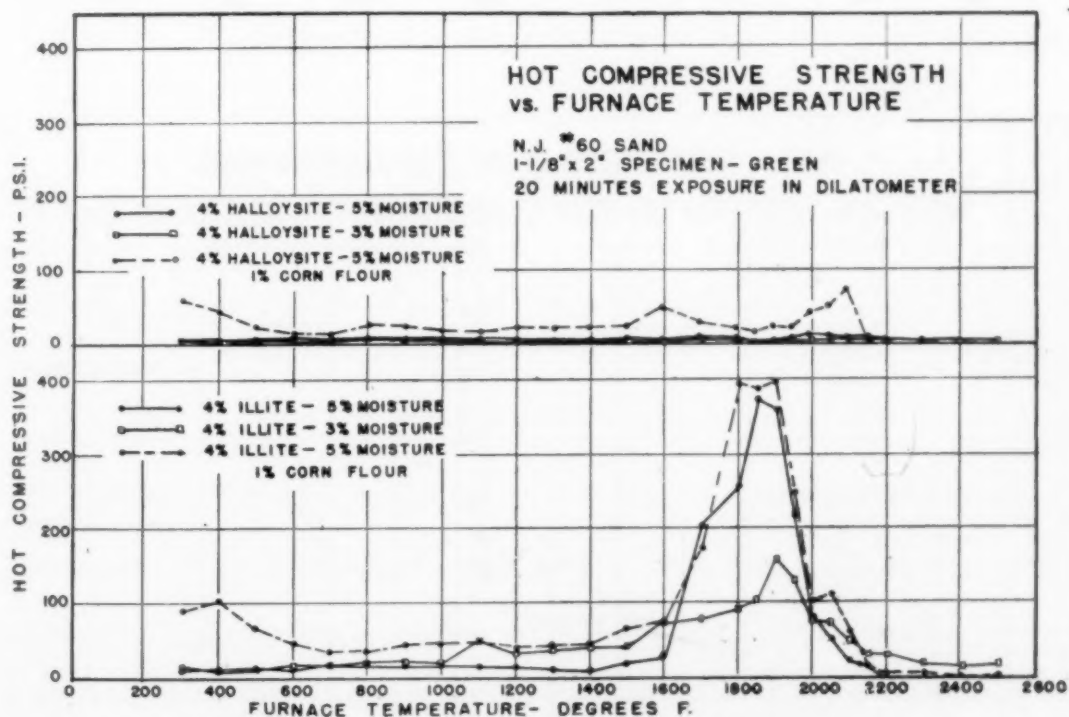


Fig. 5 (Top)—Hot compressive strength vs. temp. (4% halloysite mixture). Fig. 6 (Bottom) Hot compressive strength vs. temp. (4% illite mixture).

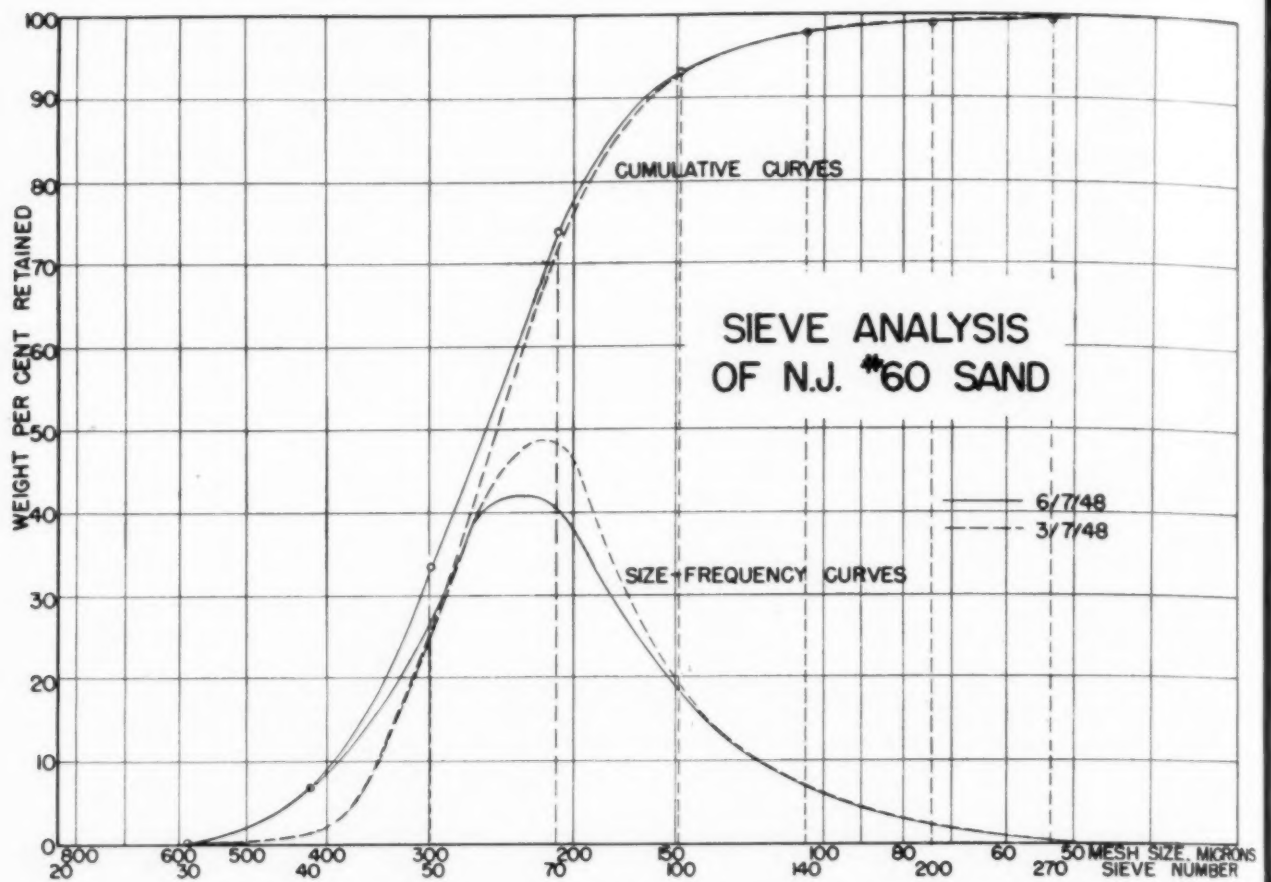


Fig. 7—Sieve analysis of N. J. No. 60 sand.

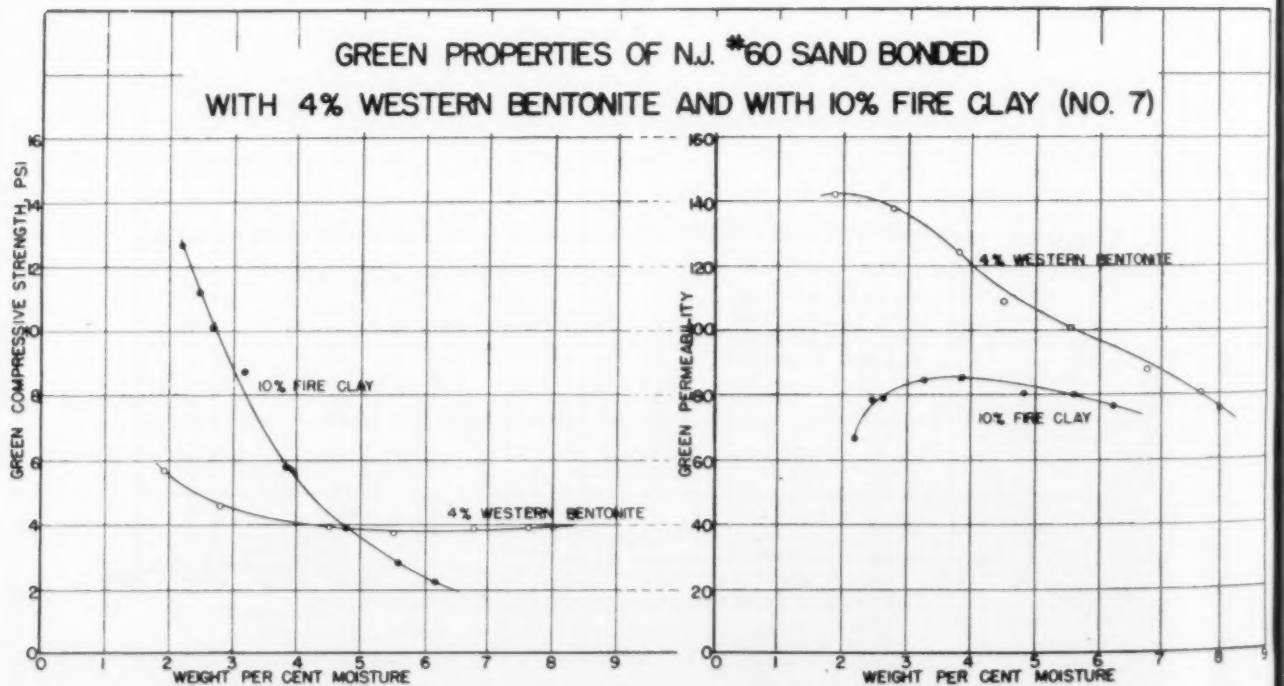


Fig. 8—Green properties of N. J. No. 60 sand bonded with 4% western bentonite and with 10% fire clay.

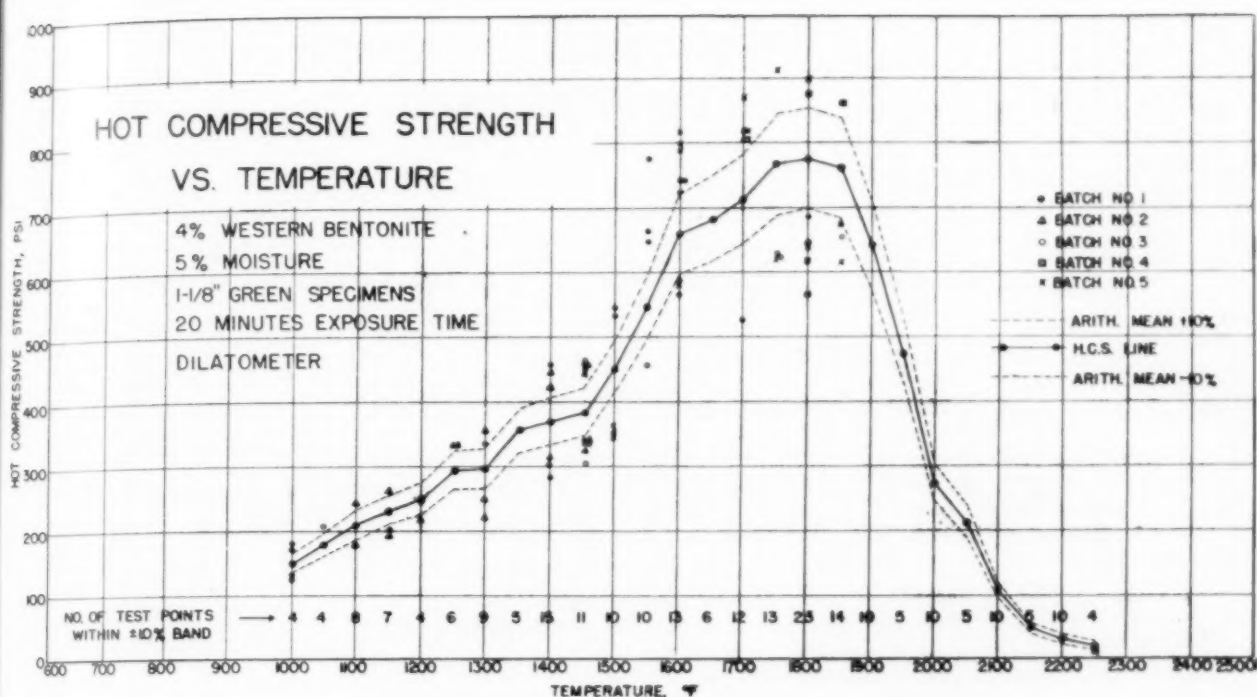


Fig. 9—Hot compressive strength vs. temperature for 4% western bentonite mixture at 20 min exposure time.

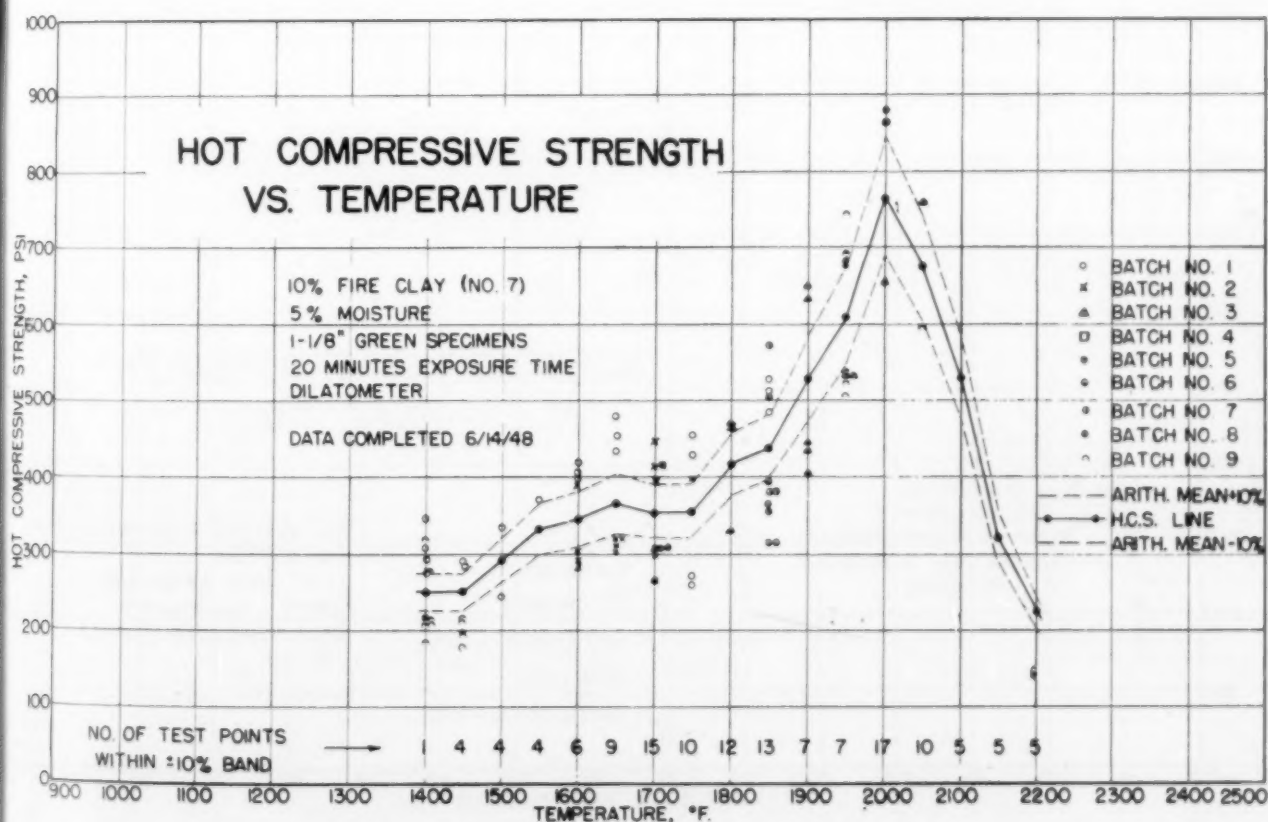


Fig. 10—Hot compressive strength vs. temperature for 10% fire clay mixture at 20 min exposure time.



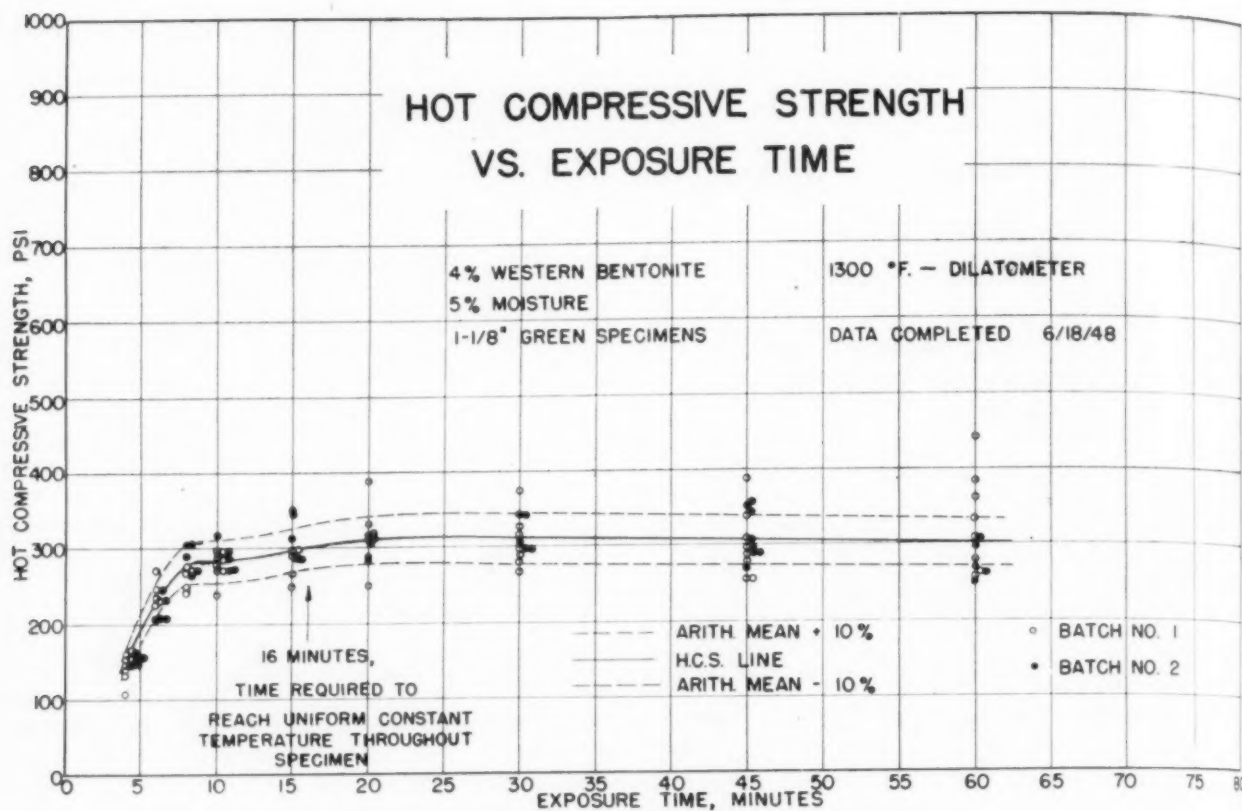


Fig. 11—Hot compressive strength vs. exposure time  
(4% western bentonite 1300 F).

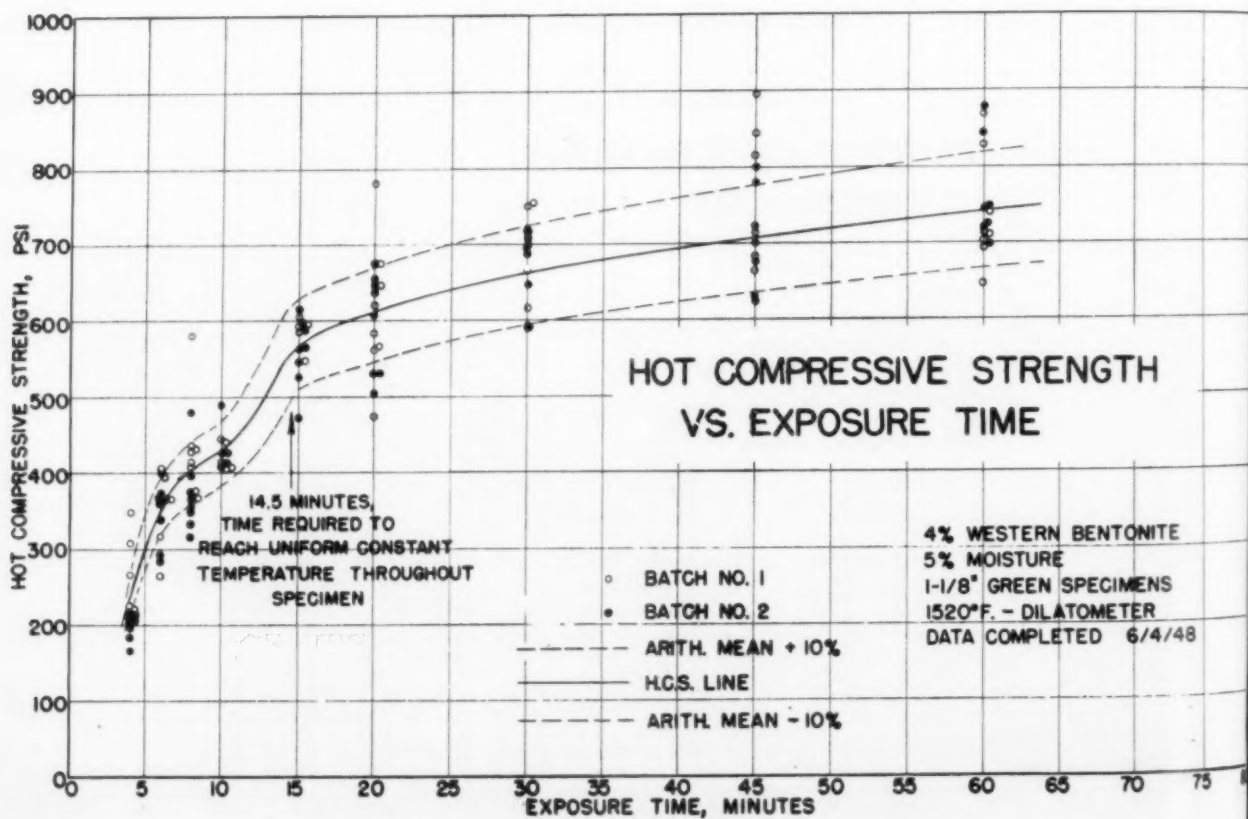


Fig. 12—Hot compressive strength vs. exposure time  
(4% western bentonite 1520 F).

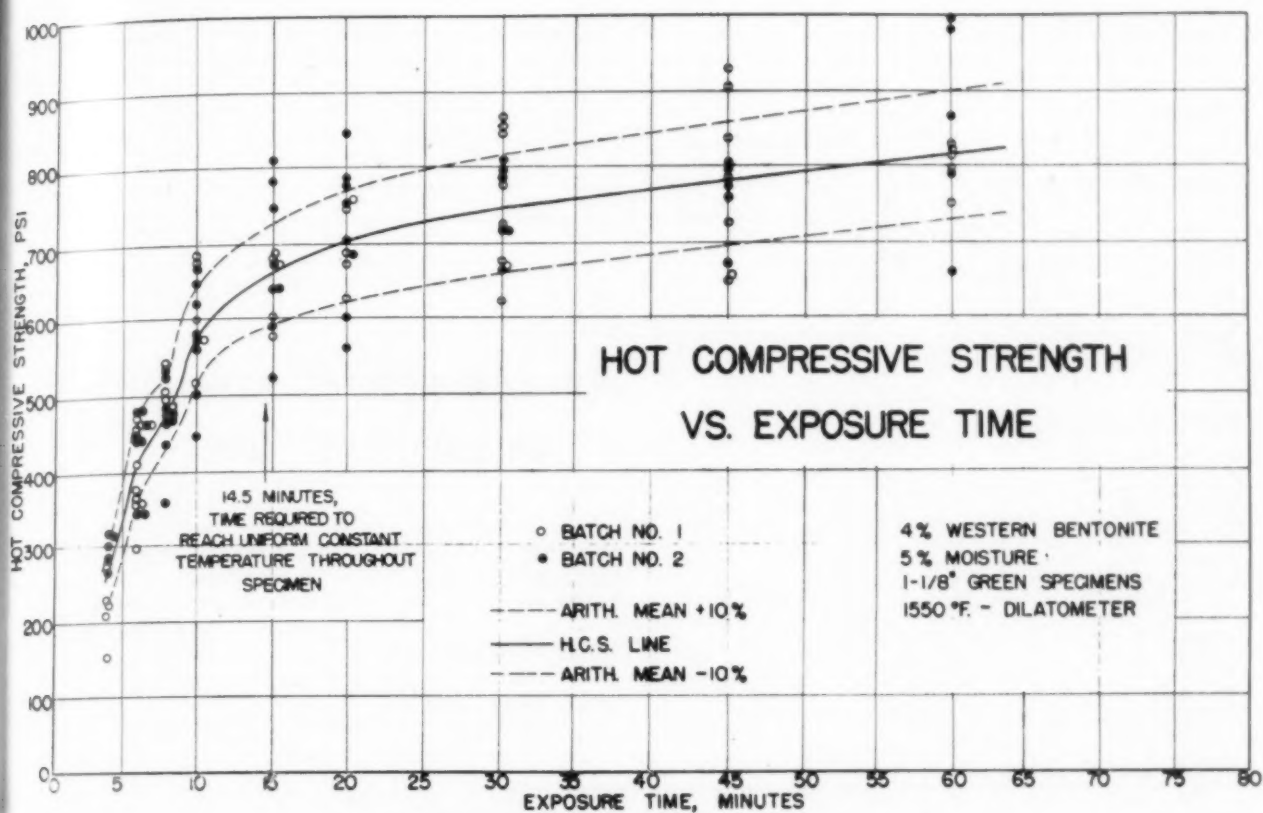


Fig. 13—Hot compressive strength vs. exposure time  
(4% western bentonite 1550 F).

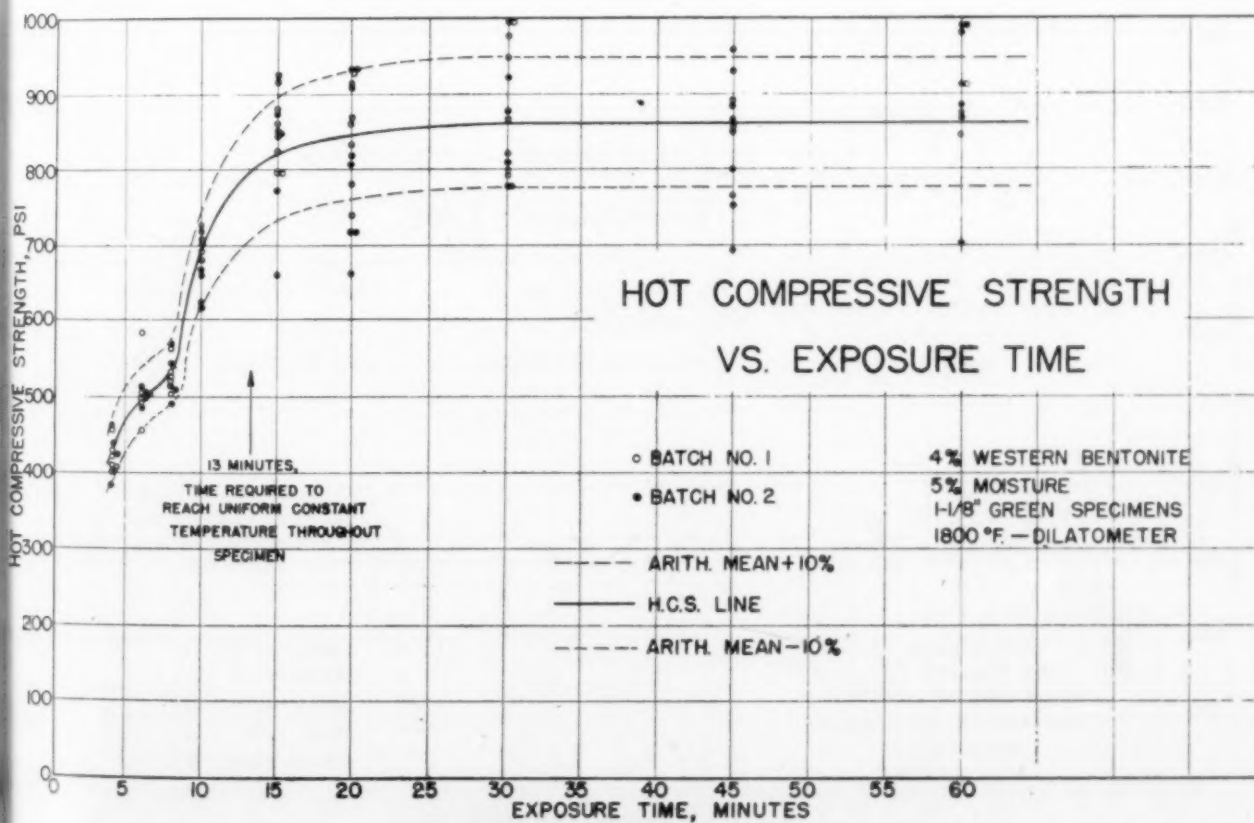


Fig. 14—Hot compressive strength vs. exposure time  
(4% western bentonite 1800 F).

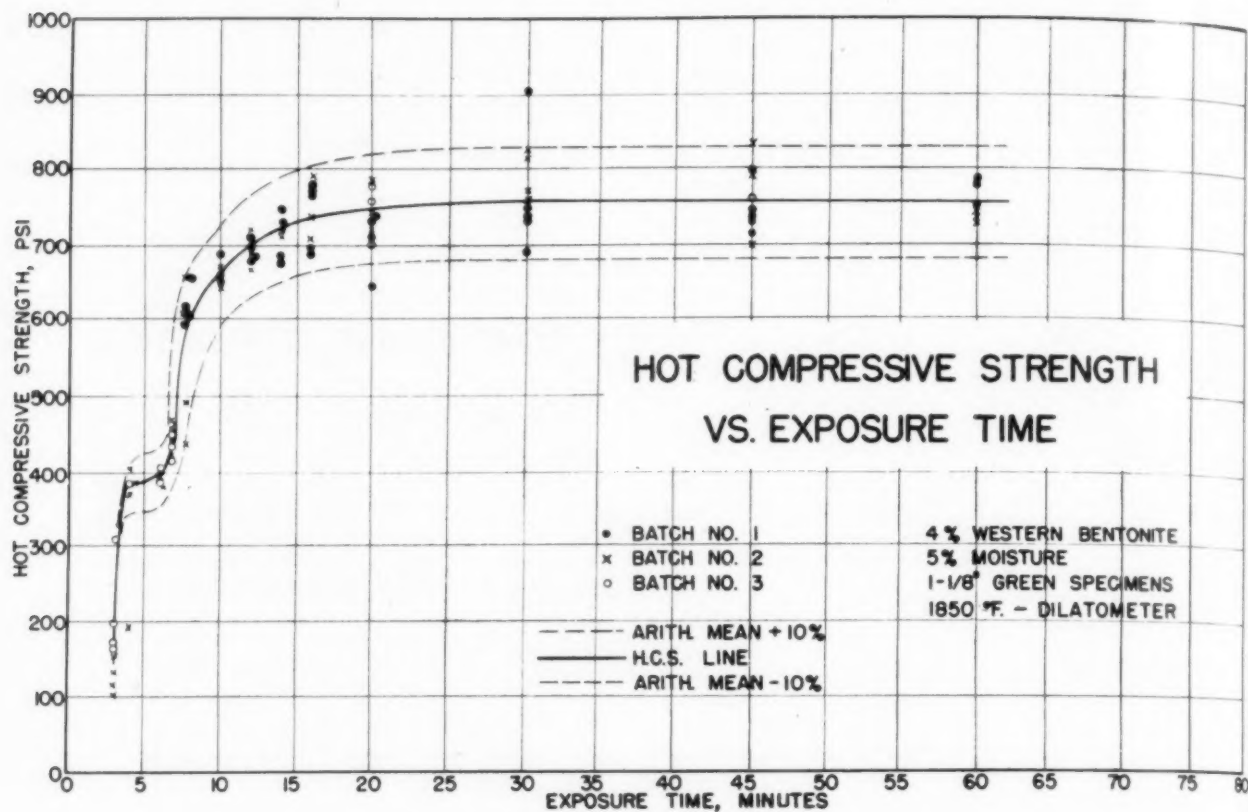


Fig. 15—Hot compressive strength vs. exposure time  
(4% western bentonite 1850 F).

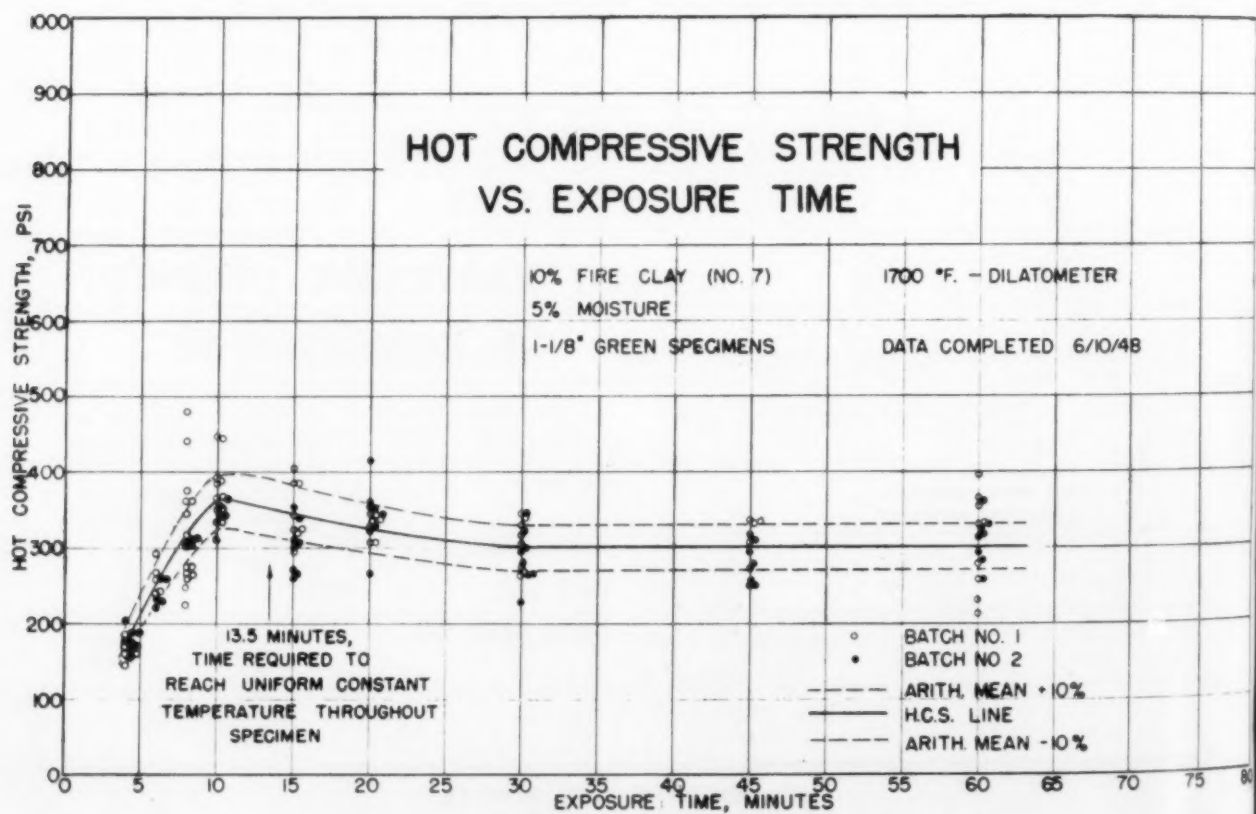


Fig. 16—Hot compressive strength vs. exposure time  
(10% fire clay 1700 F).



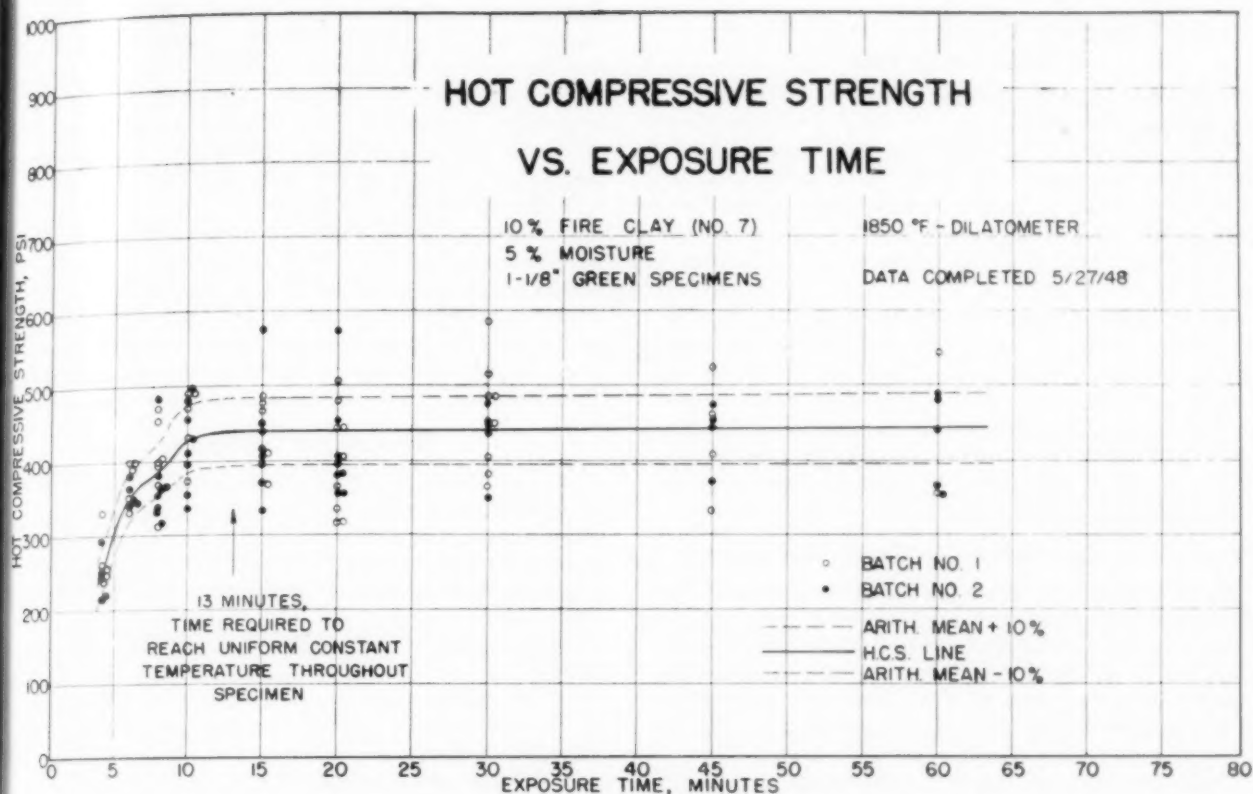


Fig. 17—Hot compressive strength vs. exposure time  
(10% fire clay 1850 F).

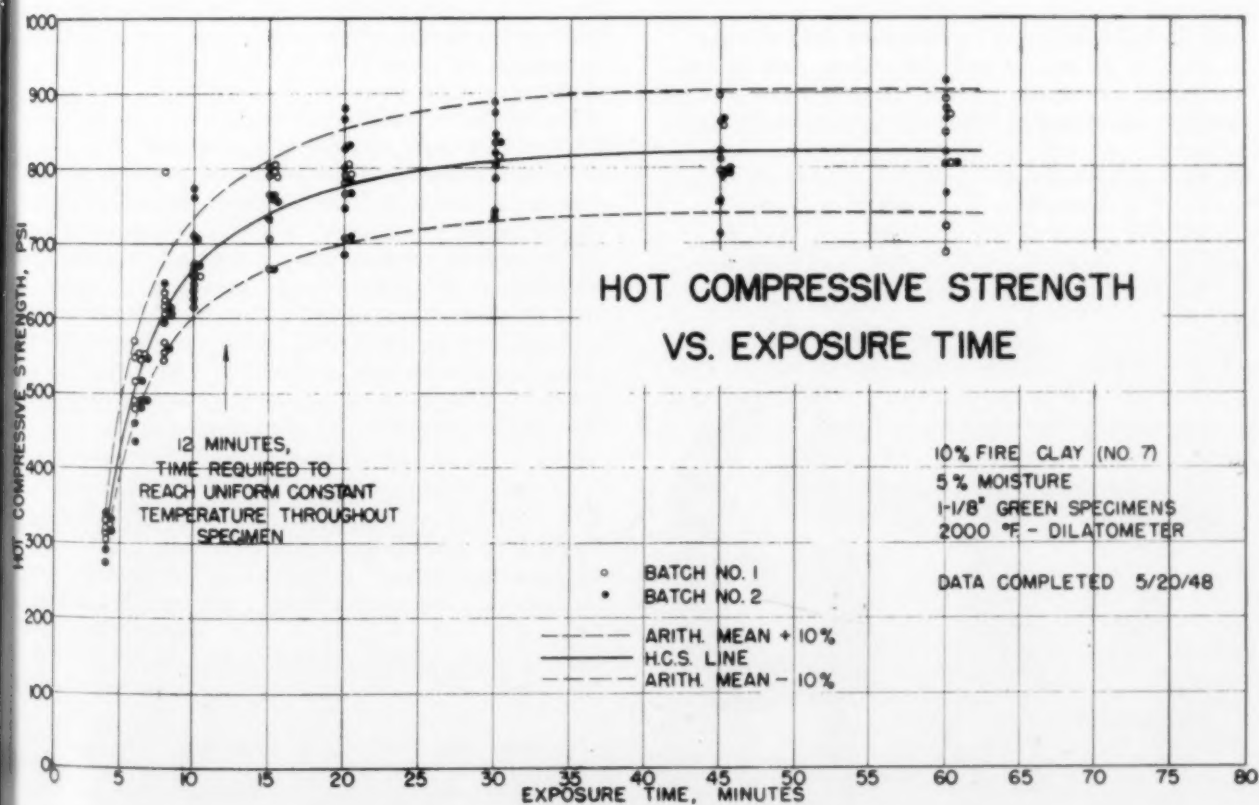


Fig. 18—Hot compressive strength vs. exposure time  
(10% fire clay 2000 F).

as closely as would be expected. The green compressive strength and permeability are shown in Fig. 8.

#### Hot Compressive Strength vs. Temperature at 20 Min Exposure Time

Before studying the effect of exposure time on hot compressive strength, it was thought desirable that the effect of temperature should be thoroughly investigated at an exposure time sufficiently long to allow the specimens to be heated to the test temperature throughout. The time chosen was 20 min which is the time used previously in strength vs. temperature studies.

Figure 9 shows these properties of the 4 per cent western bentonite mixture and Fig. 10 shows the same data for the 10 per cent fire clay mixture. For each set of data several batches were prepared in order to get the best average possible and to detect variations between batches. The number of test points falling within the  $\pm 10$  per cent of the mean value obtained is indicated along the inside of the plot at the bottom.

#### Effects of Exposure Time on Hot Compressive Strength

To date the test program has included studies of the effect of exposure time on the 4 per cent western bentonite mixture at 1300 F, 1520 F, 1550 F, 1800 F, and 1850 F and on the 10 per cent fire clay mixture at 1700 F, 1850 F and 2000 F. The results of these studies up to exposure times of 60 min maximum are shown in Figs. 11 to 18 inclusive.

These relationships fall into three types as follows:

- A. Strength increases rapidly with time to a maximum and then remains constant.
- B. Strength increases rapidly up to a certain time and increases slowly thereafter.
- C. Strength increases rapidly with time to a maximum, decreases slowly for awhile, and then remains constant.

For the mixtures tested, the relations are essentially as follows:

4% Western Bentonite	10% Fire Clay (No. 7)
1300 F — Type A	1700 F — Type C
1520 F — Type B	1850 F — Type A
1550 F — Type B	2000 F — Type B
1800 F — Type A	
1850 F — Type A	

For the conditions where the maximum strength is reached and remains constant thereafter (Type A) it is noted that the time to reach this maximum strength is greater than the time necessary to heat the specimen

throughout. For the Type B relationship, it appears that for the 4 per cent western bentonite mixture, the time to reach the point where the hot compressive strength starts to increase slowly is about the same as the time required to heat the specimen throughout, while for the 10 per cent fire clay mixture, this time is greater. In the one case where the strength drops after reaching a maximum, Type C, the time to reach maximum strength is less than the time required to heat the specimen throughout.

Further studies of these mixtures at other test temperatures and careful checking of the results at the test temperatures reported on should help in our understanding of the laws governing molding sand behavior under load at elevated temperatures. Until these additional studies are made, it will not be possible to state these laws and put them to practical use in the foundry.

#### Summary and Conclusions

The results of the work on hot compressive strength vs. temperature of the six sand mixes exposed 20 min before testing indicate that the 4 per cent western bentonite, 10 per cent fire clay and the 4 per cent illite mixtures have superior strength properties at elevated temperatures when 5 per cent moisture is used. At 3 per cent moisture, these mixtures decrease in maximum strength by amounts of 160, 130 and 200 psi respectively. The addition of 1 per cent corn flour to these mixtures containing 5 per cent moisture lowers the peak strength of the 4 per cent western bentonite mixture by 175 psi and increases the peak strength of the 10 per cent fire clay and 4 per cent illite mixture by 200 and 20 psi respectively. The peak strengths of the bentonite and illite mixtures occur at about 1800 to 1850 F, the fire clay mixtures at about 2000 F, while the halloysite mixture has no noticeable peak in most cases, but shows a slight one at about 2100 F when 1 per cent corn flour is added at the 5 per cent moisture level.

Studies of hot compressive strength vs. exposure time for the 4 per cent western bentonite and 10 per cent fire clay mixtures at 5 per cent moisture indicate that these mixtures each have different characteristics depending on the test temperature and the time required for the solid-solid reactions in the clay to proceed to completion. Further strength tests are necessary to establish the pattern of behavior for these mixtures, and additional verification of the reactions causing the strength variations is needed. These two studies are being carried on simultaneously and further progress will be reported at a later date.

# THE VALUE OF PRESSURE TESTS AND RADIOGRAPHS OF GUN METAL CASTINGS\*

By

W. H. Baer\*\*

## ABSTRACT

Cast gun metal plates 1 x 8 x 6 in. of composition 88 Cu-8 Sn-4 Zn, containing defective areas, were investigated by radiography and tested under hydrostatic pressure. After machining metal from both sides of the castings, radiographs were taken and the castings pressure tested at the new thicknesses. The plates were divided into four major classifications by the appearance of the radiographs of the 1-in. section. Radiographs of castings that were pressure-tight and castings which leaked are shown. It was concluded that castings intended for critical use should be radiographed as well as pressure tested.

To assure high quality of bronze pressure castings, it is customary to use the pressure test as the principal means of evaluation. The sensitivity of this test is considerably different when performed under altered conditions of testing. The pressure, the medium for transmitting the pressure, and the time of application are important factors. More often than not, a change in one or more of these factors makes the pressure test a highly specialized test even though it is almost universally accepted as one of the most practical means of measuring the degree of success attained in the manufacture of pressure castings.

The pressure test permits but limited interpretation of the cause or nature of a defect beyond the elemental fact that leakage takes place in a specified portion of the casting. In an attempt to remedy this failure of the pressure test a number of foundries have made initial investigations in the radiography of bronze castings. The success of the radiographic method of inspection has been hindered by the limited number of observations on which to base judgments. A few instances in which leakage has occurred in apparently radiographically sound bronze castings have raised considerable question of the reliability of this method. On the other hand, much can be pointed out regarding the pressure test that would make an occasional failure an expected occurrence under certain circumstances.

The limitations of both the radiographic and pressure testing methods of inspection stem from the

nature of bronze castings. If bronze castings have a superficially sound skin, the pressure test is a valid indication of expected service only insofar as the skin remains intact and undamaged. If fine capillaries of the metal are plugged with dirt during the pressure test and later become freed of the material, the pressure test as performed is valueless. The same factors that occasionally contribute to the inadequacy of the pressure test also occasionally affect the radiographic examination adversely. It is well known that fine cracks occurring transverse to the direction of the X-ray beam are unrecorded because they entail so little actual change in thickness. Interdendritic shrinkage may often be of this degree of fineness. From another point of view, the sound skin formation may be of such depth that serious defects revealed by radiography do not contribute to leakage in pressure tests.

Considerable experience in radiography of bronze castings was obtained in the previous portion of this investigation, which was reported as "Radiography of Gun Metal Castings" in *TRANSACTIONS of the American Foundrymen's Association*, vol. 55, pp. 153-159 (1947). It was shown that characteristic differences

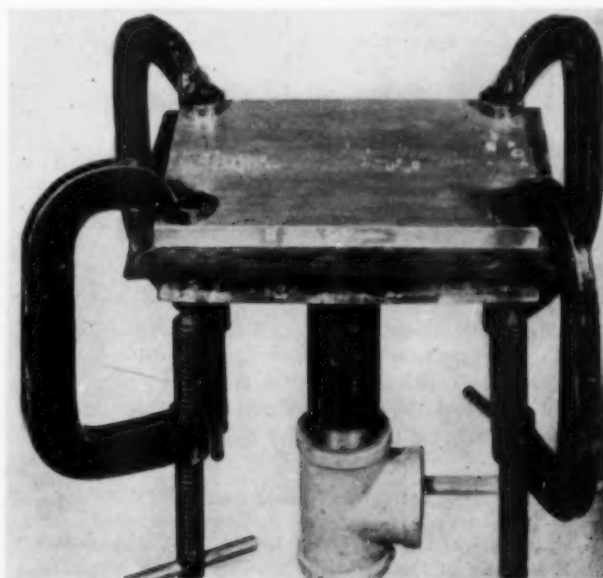


Fig. 1—Pressure testing apparatus.

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\*\* Non-Ferrous Section, Naval Research Laboratory, Washington, D. C.



TABLE 1—EFFECT OF LIGHT SHRINKAGE ON PRESSURE TIGHTNESS

Test Identity	Section Thickness							
	1 in.		$\frac{3}{4}$ -in.		$\frac{1}{2}$ -in.		$\frac{1}{4}$ -in.	
	X-ray Appearance	Pressure Tightness	X-ray Appearance	Pressure Tightness	X-ray Appearance	Pressure Tightness	X-ray Appearance	Pressure Tightness
R11-5 1	LS	HP	MS	HP	MS	HP	MS	L
R11-4 2	LS	HP	LS	HP	LS	HP	LS	HP
R7-4 3	LS & T	HP	MS	HP	MS	L	....	....
R4-3 4	LS	HP	LS	HP	MS	L	....	....
R5-4 5	LS	HP	LS	HP	LS	HP	LS	L
R11-1 6	LS & T	HP	LS & T	HP	LS & T	HP	LS	HP
R12-1 7	LS	HP	LS	HP	LS	HP	MS	HP
R12-4 8	LS	HP	MS	HP	MS	HP	MS	HP
R21-2 9	LS	HP	MS	HP	MS	HP	MS	HP
R29-2 10	LS	HP	LS	HP	LS	HP	MS	L
R10-2 11	LS & T	HP	LS & T	HP	LS & T	L	....	....
R11-2 12	LS	HP	CS	L	....	....	....	....
R11-3 13	LS	HP	LS	HP	LS	HP	LS	L
R23-1 14	LS & T	L	....	....	....	....	....	....

CODE: LS = Light Shrinkage  
MS = Medium Shrinkage

CS = Concentrated Shrinkage  
HP = Held Pressure

L = Leaked  
T = Tear

TABLE 2—EFFECT OF MEDIUM SHRINKAGE ON PRESSURE TIGHTNESS

Test Identity	Section Thickness							
	1 in.		$\frac{3}{4}$ -in.		$\frac{1}{2}$ -in.		$\frac{1}{4}$ -in.	
	X-ray Appearance	Pressure Tightness	X-ray Appearance	Pressure Tightness	X-ray Appearance	Pressure Tightness	X-ray Appearance	Pressure Tightness
R10-4 1	MS & T	L	....	....	....	....	....	....
R7-6 2	MS	HP	MS	L	....	....	....	....
R21-1 3	MS & T	L	....	....	....	....	....	....
R11-6 4	MS	HP	MS	HP	MS	HP	HS	L
R4-1 5	MS	HP	MS	HP	MS	L	....	....
R46-2 6	MS	HP	MS	HP	MS	HP	MS	L
R25-1 7	MS & T	L	....	....	....	....	....	....
R19-8 8	MS & T	HP	MS	HP	MS	HP	MS	HP
R15 9	MS	HP	MS	L	....	....	....	....
R7-5 10	MS	HP	CS	HP	CS	L	....	....
R7-3 11	MS	HP	MS	HP	MS	L	....	....
R-2 12	MS	HP	LS	HP	LS	L	....	....
R7-7 13	MS	L	....	....	....	....	....	....
R-1 14	MS	HP	MS	L	....	....	....	....
R-39 15	MS	HP	MS	HP	MS	HP	MS	L
R-28 16	MS & T	L	....	....	....	....	....	....
R-4 17	MS	HP	MS	HP	LS	HP	LS	HP
R18-1 18	MS	HP	HS	HP	HS	L	....	....
R-3 19	MS & T	HP	LS & T	HP	LS	HP	LS	HP
R42-2 20	MS	HP	MS	HP	MS	HP	MS	L

CODE: MS = Medium Shrinkage  
CS = Concentrated Shrinkage

HS = Heavy Shrinkage  
HP = Held Pressure

L = Leaked  
T = Tear

existed between radiographs of defects such as shrinkage, gas porosity, and tears. The report was also concerned with the metallurgical factors and foundry practices responsible for certain types of defects. No attempt was made to pressure-test the castings or assess the effectiveness of radiography as an inspection tool.

#### Purpose of Present Study

The purpose of the present investigation is to ascertain the value of radiography as a means of inspection of gun metal castings (nominal composition: Cu, 88 per cent; Sn, 8 per cent and Zn, 4 per cent).

#### Experimental Procedure

A. *Sequence of Operation.*—As a preliminary measure the plates were classified into four categories by the appearance of the radiographs of the 1-in. section. In each category the plates were numbered according to increasing severity of the radiographically detected non-uniformities. The four major classifications were: Light shrinkage (LS), Table 1; medium shrinkage (MS), Table 2; concentrated shrinkage (CS), Table 3, and gas porosity (GP), Table 4. Pressure tests were then made on the plates with the expectation that bronzes of low uniformity in structure would possess

TABLE 3—EFFECT OF CONCENTRATED SHRINKAGE ON PRESSURE TIGHTNESS

		Section Thickness							
		1 in.		$\frac{3}{4}$ -in.		$\frac{1}{2}$ -in.		$\frac{1}{4}$ -in.	
Test Identity		X-ray Appearance	Pressure Tightness	X-ray Appearance	Pressure Tightness	X-ray Appearance	Pressure Tightness	X-ray Appearance	Pressure Tightness
R10-1	1	CS & T	L	....	....	....	....	....	....
R5-3	2	CS & T	L	....	....	....	....	....	....
R6-1	3	CS & T	HP	CS & T	HP	CS & T	L	....	....
R10-3	4	CS	L	....	....	....	....	....	....
R6-2	5	CS & T	HP	CS & T	HP	MS & T	HP	MS & T	L
R6-3	6	CS & T	HP	CS & T	HP	CS & T	L	....	....
R5-1	7	CS & T	L	....	....	....	....	....	....
R12-3	8	CS	HP	CS	HP	CS	HP	CS	L
R6-5	9	CS	L	....	....	....	....	....	....
R7-1	10	CS & T	HP	CS & T	L	....	....	....	....
R7-2	11	CS & T	HP	CS & T	L	....	....	....	....
R5-2	12	CS & T	L	....	....	....	....	....	....
R3-4	13	HS	HP	MS	HP	MS	HP	MS	L
R17	14	CS & T	L	....	....	....	....	....	....

CODE: LS = Light Shrinkage  
MS = Medium Shrinkage  
HS = Heavy Shrinkage  
GP = Gas Porosity

HP = Held Pressure  
L = Leaked  
No Concentrated Porosity  
No Tears

TABLE 4—EFFECT OF GAS POROSITY AND SHRINKAGE ON PRESSURE TIGHTNESS

		Section Thickness							
		1 in.		$\frac{3}{4}$ -in.		$\frac{1}{2}$ -in.		$\frac{1}{4}$ -in.	
Test Identity		X-ray Appearance	Pressure Tightness	X-ray Appearance	Pressure Tightness	X-ray Appearance	Pressure Tightness	X-ray Appearance	Pressure Tightness
R23-2	1	MS & GP	HP	MS & GP	HP	MS & GP	HP	MS & GP	L
R27-2	2	MS & GP	HP	MS & GP	HP	MS & GP	HP	MS & GP	HP
R33-2	3	MS & GP	HP	MS & GP	HP	MS & GP	HP	MS & GP	HP
R24-2	4	MS & GP	HP	MS & GP	HP	LS & GP	HP	LS & GP	HP
R33-1	5	MS & GP	HP	MS & GP	HP	MS & GP	HP	MS & GP	HP
R24-1	6	MS & GP	HP	MS & GP	HP	MS & GP	HP	MS & GP	L
R29-3	7	MS & GP	HP	MS & GP	HP	MS & GP	HP	MS & GP	HP
R43-3	8	MS & GP	HP	MS & GP	HP	MS & GP	HP	LS & GP	HP
R23-3	9	MS & GP	HP	MS & GP	HP	MS & GP	HP	MS & GP	HP
R27-3	10	MS & GP	HP	MS & GP	HP	MS & GP	HP	HS & GP	L
R8-1	11	HS & GP	L	....	....	....	....	....	....
R42-4	12	MS & GP	HP	MS & GP	HP	MS & GP	L	....	....
R42-3	13	MS & GP	HP	MS & GP	HP	MS & GP	L	....	....
R9-1	14	HS & GP	L	....	....	....	....	....	....
R42-1	15	HS & GP	HP	MS & GP	HP	MS & GP	HP	MS & GP	HP
R40	16	MS & GP	HP	LS & GP	HP	LS & GP	HP	LS & GP	HP
R27-1	17	MS & GP	L	....	....	....	....	....	....
R43-2	18	HS & GP	HP	MS & GP	HP	MS & GP	HP	MS & GP	L
R44	19	MS & GP	HP	HS & GP	L	....	....	....	....
R24-3	20	MS & GP	HP	MS & GP	HP	MS & GP	HP	MS & GP	HP
R33-3	21	HS & GP	HP	HS & GP	HP	HS & GP	L	....	....
R38	22	HS & GP	HP	HS & GP	HP	HS & GP	L	....	....

CODE: LS = Light Shrinkage  
MS = Medium Shrinkage  
CS = Concentrated Shrinkage  
HP = Held Pressure  
HS = Heavy Shrinkage

L = Leaked  
T = Tear

least pressure tightness. When 100 per cent correspondence was not obtained in the 1-in. section, it was decided to investigate the value of radiography as an indication of internal quality. Equal amounts of metal were removed by machining both sides of the plates. Radiographs and pressure tests were made at thicknesses of three-quarters, one-half, and finally one-quarter of an inch. When the first leakage occurred further pressure tests and radiographs were not made.

The results of over 200 pressure tests and radiographic examinations are given in Tables 1 and 4 inclusive.

B. *The Pressure Tests.*—The special jig shown in Fig. 1 was made to hold the plates during the pressure test. With this apparatus, an hydraulic pressure of 250 psi was applied to one side of a plate and held constant for 15 min. This is more severe than most pressure tests made on bronze castings. A special precaution in testing was the avoiding of air pockets

immediately under the plate by holding the plate at an angle of three or four degrees to the horizontal to allow the air to leak through the rubber gasket before the clamps were tightened and maximum hydraulic pressure applied.

**C. Radiography.**—The radiographs of the 1-in. section were taken with 220 KV and ten milliamperes with an exposure time of five minutes and a 36-in. target-to-film distance. A lead filter 0.03 in. in thickness was used at the source and lead intensifying screens 0.005 in. in thickness were placed on both sides of Eastman Type F X-ray film. In order to obtain maximum contrast the voltage of the 220 KV unit was decreased with decreasing thickness so that it was only 180 KV at the thinner sections. No changes were made in the other conditions of the radiographic procedure, and with a density of 1.25 the sensitivity was usually about one per cent.

Figures 2 to 5 are illustrations of typical non-uniformities found in the 1-in. sections in pressure tight castings. With the exception of casting R6-2 shown in Fig. 4, all of the castings held pressure in the one-quarter inch section. This particular casting (R6-2) leaked only when the thickness was reduced to one-quarter inch.

Non-Uniformity	Plate No.
Light Shrinkage	Fig. 2, R11-4
Medium Shrinkage	Fig. 3, R-3
Concentrated Shrinkage	Fig. 4, R6-2
Shrinkage and Gas Porosity	Fig. 5, R24-3

Every radiograph that is described in Table 1 to 4 inclusive has been kept for reference and is available for examination at the Naval Research Laboratory. Since wide variations existed in each of the four classifications of non-uniformities there is a tendency for some types to overlap.

#### Discussion of Results

Considerable difficulty was encountered in producing a single type of non-uniformity in the plates. As will be noted in the tables, light and medium shrinkage were frequently associated with tears. Similarly gas porosity often had shrinkage as an accompanying non-uniformity.

**A. Light Shrinkage.**—Several important facts regarding radiography of light shrinkage are shown in Table 1. It will be noted that only one of the castings leaked when tested in the 1-in. section. Considering the total castings that leaked at various thicknesses, approximately two-thirds of them were not pressure tight even though the shrinkage was very light. The interpretation of radiographs showing light shrinkage in the 1-in. section is difficult because the shrinkage may be concentrated, as is shown by the change in description to medium shrinkage in many of the radiographs of the thinner sections. The interpretation is made more difficult by the fact that such concentration is not necessarily indicative of failure in the pressure test. Of the five quarter-inch sections that held pressure, three did so in spite of medium shrinkage.

**B. Medium Shrinkage.**—The general behavior of the 1-in. castings showing medium shrinkage might have been anticipated from the results obtained on plates showing light high shrinkage. Only five of the plates in this group gave initial leakage. After the castings had been machined to thinner sections, the soundness was much more impaired. Only three held pressure at one-quarter of an inch in thickness and in two of these the shrinkage was classified as light. It can be said qualitatively, by comparing Tables 1 and 2, that the more severe the shrinkage, the less the pressure tightness.

**C. Concentrated Shrinkage.**—Seven of the fourteen 1-in. castings showing concentrated shrinkage were not pressure tight (Table 3). The seven remaining castings failed at some thinner section. This indicates that concentrated shrinkage is the most serious kind of shrinkage. For castings with this type of non-uniformity, radiography appears to be a desirable method of inspection.

**D. Shrinkage and Gas Porosity.**—Several outstanding characteristics of gas porosity are evident from Table 4. First, leakage occurred in the 1-in. plates to about the same extent as it did with light shrinkage. Second, a higher proportion of the plates held pressure in the quarter-inch section than in any other set of experiments. Third, there was fairly uniform distribution of the gas porosity and accompanying shrinkage. In very few instances did the intensity of the porosity change from the designation in the 1-in. section. This category of non-uniformity is distinguished by the absence of tears and concentrated shrinkage.

**E. Tears.**—One of the most serious type of defect is the hot tear. This defect almost always occurs with other defects and cannot easily be separated into a special category. Although as shown in Table 2, tears may occasionally be superficial, it is believed that even minute tears present a strong basis for rejection of gun metal castings.

**F. Significance of Radiography of Gun Metal Castings.**—If radiographs of gun metal castings are free of any indications of non-uniformities the castings should be pressure tight. If, however, there are indications of non-uniformities, even though they are slight, the possibility of leakage exists. The fact that a radiograph indicates a slight non-uniformity is not a valid reason for assuming that the damage to the casting is in proportion to the severity of the non-uniformity, as will be understood by an inspection of Fig. 2 to 5. The type of non-uniformity is very important.

#### Conclusions

1. Gun metal castings showing light shrinkage in a radiograph are of questionable acceptability.
2. Radiographs showing medium shrinkage or gas porosity in combination with shrinkage indicate a serious condition and greater possibility of leakage under pressure.
3. There is little question about radiographs of castings that have concentrated shrinkage or tears. These are the most dangerous types of non-uniformities and are sufficient reason for rejecting castings.
4. Castings intended for critical use should be radiographed as well as pressure tested.



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Fig. 2—Light shrinkage; pressure tight.



Fig. 3—Medium Shrinkage; pressure tight.

Exographs reduced  $\frac{1}{4}$  in reproduction.

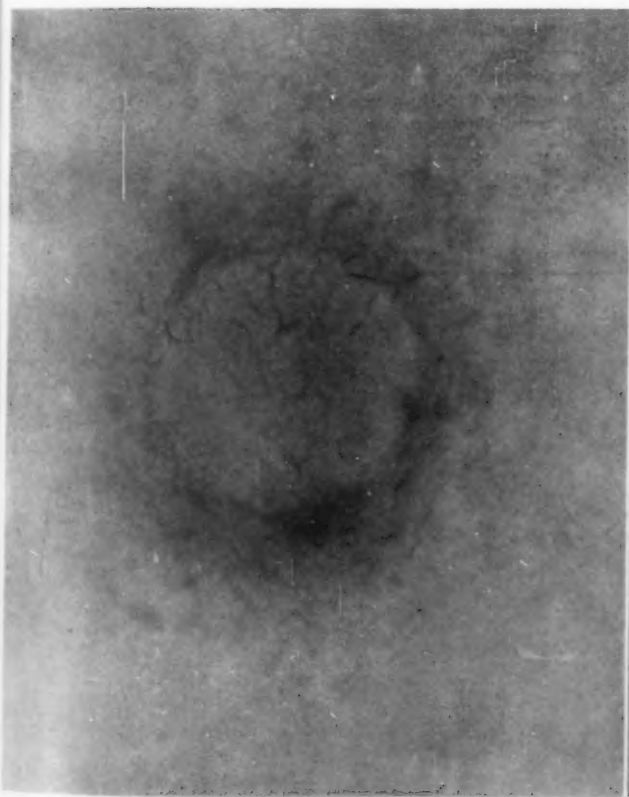


Fig. 4—Concentrated shrinkage; leaked at  $\frac{1}{4}$  in.

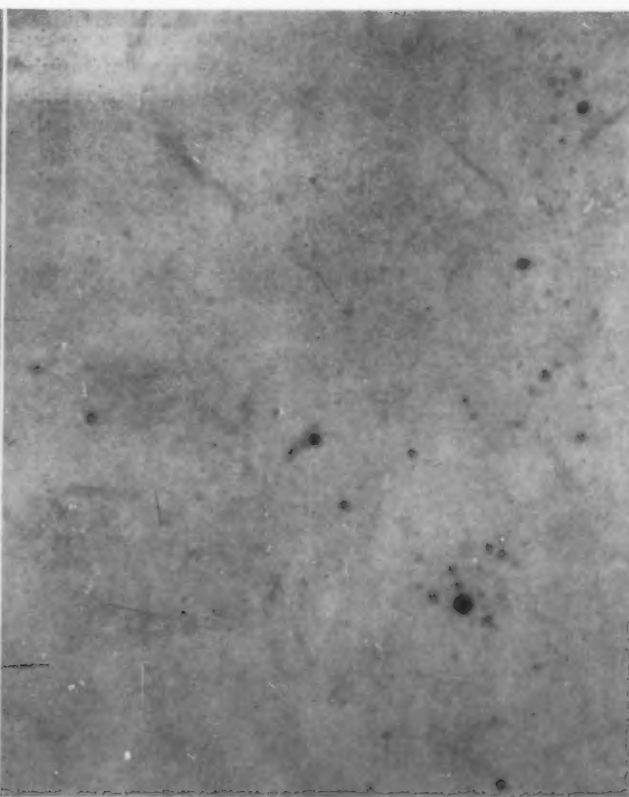


Fig. 5—Shrinkage and gas porosity; pressure tight.

Exographs reduced  $\frac{1}{4}$  in reproduction.

## APPENDIX

## Mechanical Properties of Some Defective Bronze Castings

Selected plates that leaked in the 1-in. section were used for standard 0.505-in. diameter tensile specimens machined from sections containing defects. The test specimens were selected so that each specimen showed a difference in the degree of severity of the defect. The positions from which tensile test specimens were taken are shown on the radiographs, Fig. 6 to 9.

Non-Uniformity	Plate No.
Light Shrinkage	Fig. 6, R4-4
Medium Shrinkage	Fig. 7, R7-7
Concentrated Shrinkage	Fig. 8, R6-5
Shrinkage and Gas Porosity	Fig. 9, R9-1

These are four typical radiographs corresponding to the four classifications given in the main part of this report. The test coupons were taken over the greatest gradation of non-uniformity found. When the machining of the tensile test specimens were completed, they were subjected to visual examination with the naked eye and also with a 32 mm microscopic lens. The tensile specimens were pulled at a rate of 0.05 in. per min. The percentage of discoloration of the fracture and the type of fracture was noted with each specimen. The results of the mechanical tests are given in Table 5.

As would be expected, mechanical properties of unsound bronze castings are far below the requirements for gun metal. The results given in Table 5 are typical of data taken from ten plates that were investigated. The tensile strength of a sound plate might be 4000 to 5000 psi below specification, but certainly not 20,000 psi below, as was found in many of the castings. It was not possible to find characteristic effects of the various types of non-uniformities on the

mechanical properties. There is fairly good agreement however between severity of defect estimated by radiography and the damage to mechanical properties.

## DISCUSSION

Chairman: R. W. PARSONS, Ohio Brass Co., Mansfield, Ohio  
Co-Chairman: A. K. HIGGINS, Allis-Chalmers Mfg. Co., Milwaukee

K. A. MIERCKE: <sup>1</sup> Did you have some specific reason for checking Type F film for this work? Did you try Type A or M for greater contrast and finer detail?

MR. BAER: If you will refer to "Radiography of Gun Metal Castings" which I had published in TRANSACTIONS, A.F.A., vol. 55, pp. 153-9 (1947) you will find the reason why we used Type F film for this work. It was mainly because of the fact that Type F film is used in industry more than practically any other type of film. When we started this problem, in order to cut down on variables, we decided to use the same type of film all the way through.

R. G. TOBEY: <sup>2</sup> At the present time, there seems to be a growing trend away from the use of such films as Kodak Type F with lead intensifying screens in exposures of this type. The finer graininess and higher contrast of Kodak Type A and Type M films yield greater radiographic sensitivity, which probably explains the current preference for materials of these types.

CO-CHAIRMAN HIGGINS: In addition to the author's work on this subject, the Bureau of Ships has been interested in the same problem. When critical castings have been radiographed, what proportion of the commercial castings examined have been such that they would be accepted without hesitation?

C. L. FREAR: <sup>3</sup> First of all, it is almost impossible to find any cases where radiography has been used as an acceptance medium on bronze castings. We hope in the near future to develop radiographic techniques, interpretation and standards of acceptance whereby we will be able to inspect castings by radiography. To date we have not accomplished too much but the present paper will be a distinct aid in this direction.

We have been doing some work in the Bureau in trying to correlate the radiographic images with the properties of the castings and their resistance to leakage under hydrostatic tests. But we still have not obtained much data. I would say, if we were going to start using radiography right now on bronze castings, rejections would be about 95 per cent. The position is not hopeless, however, as the chief difficulty lies in our lack of knowledge as to just what effect any discontinuity will have

<sup>1</sup> U. S. Gypsum Co., Chicago

<sup>2</sup> Eastman Kodak Co., Rochester, N. Y.

<sup>3</sup> U. S. Navy Dept., Bureau of Ships, Washington, D. C.

TABLE 5—EFFECT OF SHRINKAGE ON MECHANICAL PROPERTIES OF GUN METAL

Test Identity	Position of Test Bar in Plate	Tensile Strength, d.s.i.	Elongation, Per Cent	Fracture Appearance
Light Shrinkage				
R5-4	1	35100	29	100% Woody
R4-4	2	23900	12	50% Woody, 50% dark oxide
	3	24000	10	50% Woody, 50% dark oxide
	4	27250	15	80% Woody, 20% dark oxide
Medium Shrinkage				
	1	28500	27	35% Woody, 65% dark oxide
R7-7	2	29400	B.O.	35% Woody, 65% dark oxide
	3	31500	20	50% Reddish-grey, 50% dark oxide
	4	28750	20	85% Grey, 15% reddish-yellow
Concentrated Shrinkage				
	1	11400	..	5% Woody, 95% black oxide
R6-5	2	10200	..	5% Woody, 95% black oxide
	3	20500	15	25% Woody, 75% black oxide
	4	28000	21	50% Woody, 50% black oxide
Heavy Shrinkage and Gas Porosity				
	1	32700	22	50% Light Brown, 50% grey
R9-1	2	21100	..	40% Woody, 60% light red oxide
	3	22200	..	30% Grey, 70% red oxide
	4	26500	15	20% Woody, 80% dark oxide



Fig. 6—Light shrinkage; leaked at  $\frac{1}{4}$  in.



Fig. 7—Medium shrinkage; leaked at 1 in.

Exographs reduced  $\frac{1}{4}$  in reproduction.

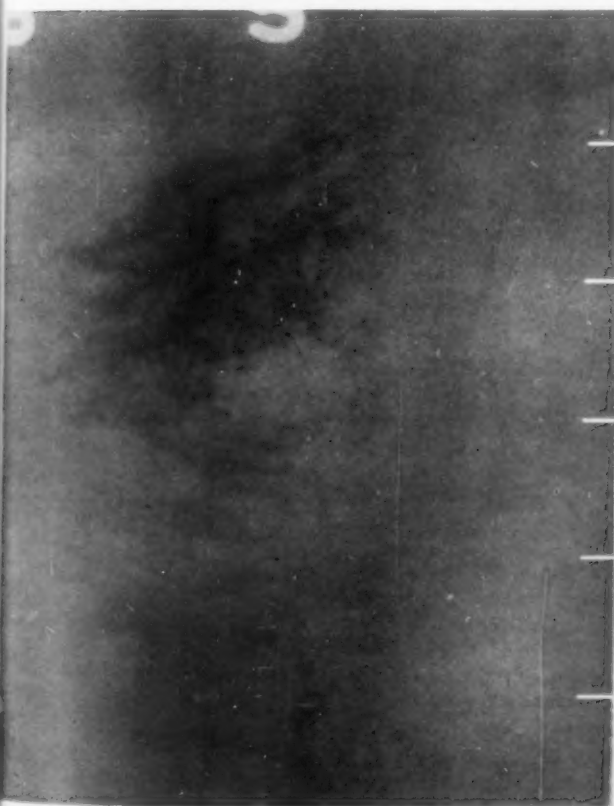


Fig. 8—Concentrated shrinkage; leaked at  $\frac{1}{2}$  in.

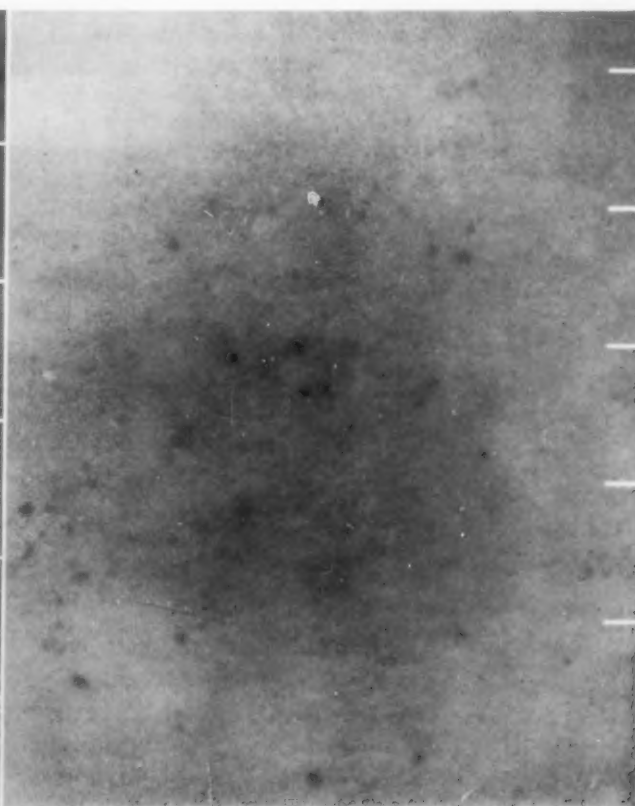


Fig. 9—Shrinkage and gas porosity; leaked at 1 in.

Exographs reduced  $\frac{1}{4}$  in reproduction.



Of course, developments in foundry technique will be a great step forward in this direction.

J. F. EDNIE: <sup>\*</sup> I would like to ask Mr. Baer if he has any additional data to supplement that shown in the paper regarding the tensile properties of the good plates that were pressure tight. In our opinion, most of the values shown are quite low and not very uniform. Would this suggest that the plates are not substantially sound? Would it not be possible to cut a series of tensile bars out of one of the good plates after it has passed the pressure test and thereby ascertain the quality and uniformity of the plate, as a casting?

MR. BAER: The plates that were radiographically sound and did hold pressure were machined down, and as a result I did not have enough to make a tensile bar after that. When I did select my plates for tensile properties, I did not necessarily try to select the best ones. I tried to select one that would compare radiographically to the previous four that I had shown that did hold pressure. I was working more on the pressure tightness rather than the tensile properties. The tensile properties were taken to determine the correlation, if any, between tensile properties and radiographs.

CO-CHAIRMAN HIGGINS: Radiographically sound sections of Navy "G" bronze castings, made commercially with, I believe some inspection from Mr. Frear did show properties that were in excess of those required by specification, namely, 40,000 psi tensile and 25 per cent elongation. Unsound sections from the same casting were considerably below that.

MR. FREAR: Before we can use radiography as a method of inspecting castings, I believe we are going to require considerable standardization. I am thinking of a set of bronze plates, one plate especially, which I borrowed from Mr. Baer and had it radiographed with radium and X-rayed using two million volts, one million volts, 250 kv and 140 kv. The radium did not show discontinuities in this particular plate, neither did the two million, nor the one million volt X-rays. The 250 kv X-ray showed considerable shrinkage and the 140 kv indicated considerably more shrinkage than actually existed. The 140 kv started to show up heterogeneity in the form of flow lines in the metal. Therefore, we are not only going to have to standardize on the type of film but also the equipment as well as the

electrical input of the X-rays which we use in the radiographic examination.

CO-CHAIRMAN HIGGINS: Does not the standard penetrameter show the sensitivity sufficiently well?

MR. FREAR: It will to a certain extent, but the smallest hole in the standard penetrameter is much larger than the size of the defect, especially the individual voids due to interdendritic shrinkage, which is probably the chief cause of leakage in the tin bronzes. In this connection, most of these interdendritic voids are smaller than one per cent of the wall thickness of the casting. One per cent sensitivity is about the limit for X-rays even when the most exacting technique is used.

R. F. HOLSTE: <sup>\*</sup> What was the thickness of these plates?

MR. FREAR: The thickness of these plates was 1 in.

MR. HOLSTE: It is known to radiographers that there is an optimum voltage to use for every thickness of material to be examined. I do not think that trying to make comparisons at all voltages is the best thing to do. Of course, I appreciate that we do have to select a certain voltage for the particular thickness of plate that we are going to examine, and I think this merely brings out all the more, that this selection is going to be necessary. Now, as far as the 140 kv machine showing something that was not in the plate, I do not think an X-ray machine will do that.

CO-CHAIRMAN HIGGINS: We have found it not uncommon to find shadows of large crystals occurring in castings that appear to be defects and are not.

MR. HOLSTE: That is true, you can have that sort of an indication.

CO-CHAIRMAN HIGGINS: These pseudo-defects can of course be recognized as different from real faults.

J. I. MEDOFF: <sup>\*</sup> Is not that 0.030 in. filtration you are using at the tube unusually heavy? We more commonly use 0.10 or 0.12 in. most of the time.

MR. BAER: It may be a little heavy, but we seem to get very good results.

MR. MEDOFF: Do you use that as standard practice?

MR. BAER: Yes, we used that as standard practice for this problem. An answer to a previous discussion, we did reduce the kilovoltage as the thickness of the plate was reduced, so that for some of the radiographs we were down to 180 kv.

<sup>\*</sup> Duquesne Smelting Corp., Sub. of The American Metal Co., Ltd., Pittsburgh

<sup>\*</sup> General Electric X-ray Corp., Chicago

<sup>\*</sup> Metallurgical Testing Services, New York

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# HEAT TRANSFER

A.F.S. Committee Report

## REPORT OF THE COMMITTEE CHAIRMAN

Dr. H. A. Schwartz\*

ACTIVITY of the Heat Transfer Committee for 1947 has been largely in rounding out and elaborating work reported in 1946. Thus there is a report from Battelle Memorial Institute dealing with Heat Transfer in a granular material of mixed grain sizes, which supplements last year's work on three separate grain sizes.

The Heat Flow Analyzer Laboratory at Columbia University has completed studies correlating the bleeding tests made last year on aluminum and cast iron, with the results to be expected from the best available information as to thermal properties.

On aluminum the results are quite satisfactory. On cast iron the concordance between computation and observation is not all that it might be. It is suspected that this is, in part, due to the difficulty of making good bleeding tests on a material having a considerable freezing range, and therefore a fairly wide zone in which the material is only partially frozen.

Dr. V. Paschkis has also compared the results reported many years ago by Briggs and Gezelius\*\* on the freezing rate of spheres, with the results which can be predicted on the heat flow analyzer, and secured pretty good corroboration within the times which these observers used.

It is believed that we now have ample evidence that the heat flow analyzer produces useful results, both in the case of unidirectional heat transfer and in the case of heat transfer in three directions.

It is proposed in the following year to make a study at Columbia University on the relative rates of freezing on the simple shapes—the slab, cylinder and the sphere—as a function of dimension, in the hope of developing isochrones, that is, lines of equal time marking boundary between completely liquid and partly frozen metal. There is considerable reason to hope that it will be easier to develop such an attack on relative freezing rates than to proceed in each case with an attempt to develop the absolute times.

It seems appropriate to emphasize in this report that the work at Columbia on spheres and slabs does not confirm the conclusion of Chvorinov that the freezing time of a body could be determined as a function of the ratio of its surface area to its volume.

The technical details of this report will be found in the series of papers forming a part of this program. They are as follows:

"Thermal Conductivity of a Sand Mixture" by C. F. Lucks, O. L. Linebrink and K. L. Johnson,

"Studies on the Solidification of Aluminum Castings" by Dr. V. Paschkis,

"Studies on Solidification of White Iron Castings," by Dr. V. Paschkis, and

"Study on Solidification of Steel Spheres," by Dr. V. Paschkis.

\* Manager of Research, National Malleable and Steel Castings Co., Cleveland.

\*\* "Studies on Solidification and Contraction in Steel Castings." TRANSACTIONS, A.F.A., vol. 43, p. 274 (1935).

## THERMAL CONDUCTIVITY OF A SAND MIXTURE

By

C. F. Lucks, O. L. Linebrink and K. L. Johnson\*

### Summary

THE THERMAL CONDUCTIVITY of a mixture consisting of 60 per cent 20-30 sand (Ottawa Silica Co.),

and 40 per cent No. 7 sand (American Graded Sand Co.) was determined over a temperature range of 750 to 2250 F. The specimen tested had a unit weight of 111 lb. per cu. ft.

The thermal conductivity of the sand mixture in-

\* Battelle Memorial Institute, Columbus, Ohio.

creased with temperature. The thermal conductivity, in  $\text{Btu hr}^{-1}\text{ft}^{-2}\text{in.}^{\circ}\text{F}^{-1}$ , at 750 F is 2.6 and at 2250 F is 6.5.

### Specimens

The sand mixture tested was made up from the 20-30 sand and No. 7 sand, described in our report entitled "The Thermal Conductivities of Three Sands" in *TRANSACTIONS OF A.F.A.*, vol. 55, p. 62 (1947). Sixty pounds of sand consisting of 60 per cent 20-30 sand and 40 per cent No. 7 sand were placed in a cylindrical container. The sands were dry mixed by rolling for 2 hr. at approximately one revolution each 25 sec. Five per cent water was then added to the sand mixture and the rolling continued for 3 hr. The slight amount of sand adhering to the container walls, at the conclusion of the latter rolling, was mixed thoroughly into the mixture by hand.

A specimen of the wet sand mixture was prepared by jolting the mixture in the specimen container as described on pp. 63, 65 and 66 of the previously mentioned report. A total of 40 jolts on a foundry mold jolting machine (as described in above mentioned report) were used to pack the specimen. The decrease in specimen thickness after 12 jolts was about  $\frac{1}{2}$ -in. and after 28 additional jolts approximately  $\frac{1}{8}$ -in.

The specimen and its container were heated in an oven for 5 hr. with a gradual increase in temperature from 150 to 210 F. The drying was then completed by heating the unit to 400 F and holding it at this temperature for 17 hr. The dried specimen was used for the thermal-conductivity measurements. The unit weight of the specimen as tested was 111 lb. per cu. ft.

### Method and Apparatus

The method and apparatus described in our report entitled "The Thermal Conductivities of Three

Sands" published in *TRANSACTIONS, AFA*, vol. 55, were used to determine the thermal conductivity of the present specimen.

### Results

Observed temperatures at each thermocouple position for each equilibrium condition are given in Table 1.

TABLE 1—OBSERVED TEMPERATURES AT VARIOUS POSITIONS FOR SAND MIXTURE (60 PER CENT 20-30, 40 PER CENT No. 7)

Position of Thermocouple From Bottom, In.	Temperature, $^{\circ}\text{F}$ Equilibrium State		
	1	7	3
20/64	585	771	896
45/64	858	1154	1325
62/64	1018	1361	1554
110/64	1414	1818	2043
136/64	1596	2020	2258

The thermal-conductivity values calculated in the temperature range of 750 to 2250 F at 250 F intervals, using tangents to temperature-thermocouple position curves for temperature gradients at the selected temperatures, are given in Table 2. The data of this table are shown plotted in Fig. 1.

A theoretical treatment of the thermal-conductivity of dry sands has been prepared by Dr. J. C. Bell, and is presented in Appendix I of this report. Thermal-conductivity values obtained by use of the theory and

Fig. 1—Thermal conductivity vs. temperature for mixed sand (60% 20-30, 40% No. 7)

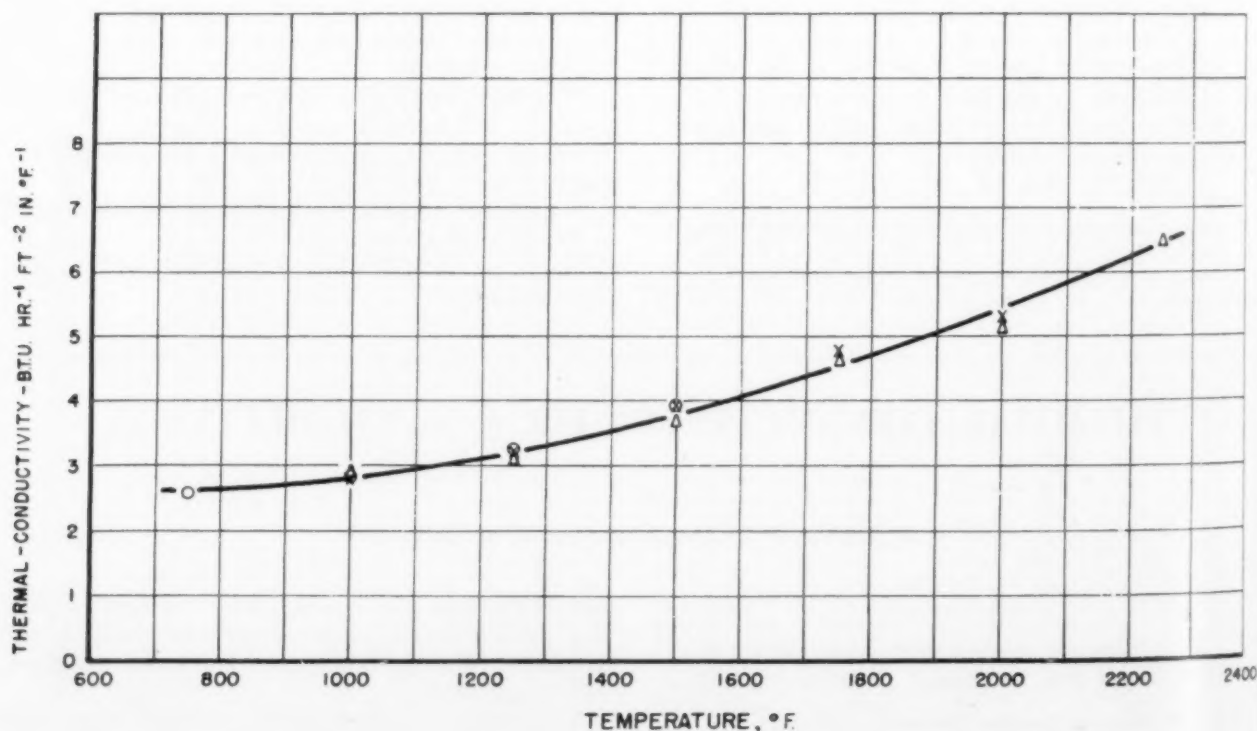




TABLE 2—THERMAL CONDUCTIVITIES OF SAND MIXTURE (60 PER CENT 20-30, 40 PER CENT No. 7) AT VARIOUS TEMPERATURES

Temperature °F	Thermal Conductivity* Equilibrium State			
	1	2	3	Average
750	2.60			2.6
1000	2.81	2.78	2.88	2.8
1250	3.28	3.23	3.13	3.2
1500	3.96	3.92	3.68	3.9
1750		4.82	4.68	4.8
2000		5.39	5.17	5.3
2250			6.52	6.5

\* Btu hr-1ft-2in °F-1

those obtained experimentally are summarized in Table 3 for the three sands covered in our previously mentioned report and the present sand mixture. The agreement between the theoretical and experimental values is certainly sufficient to verify that the experimental values found in the mixed sand are reasonable.

The theoretical and experimental data indicate that filling the voids of a coarse sand with a fine sand results in a thermal-conductivity value for the mixture which is lower than would be anticipated, based upon the increased unit weight. With fine sand in the voids of a coarse sand, the increased number of paths per unit length for radiation is more effective in reducing the thermal conductivity than the increase of solid material is in increasing the thermal conductivity.

## Appendix I

### Thermal Conductivity of Dry Sands

By  
J. C. Bell

Dry sand is normally a mixture of solid particles and air. Since the thermal conductivities of both these substances can be determined experimentally, it is reasonable to try to use such measured values to predict what the conductivity of the mixture will be. An approximate theory suitable for this purpose has been described by H. W. Russell.\*

One complication involved in using Russell's theory is that the conductivity of air due to radiation ( $K_r$ ) depends on the thickness of the air space. For a random array of sand particles, the air space thickness evidently varies greatly. However, to get some reasonable value to use, we may proceed as follows:

Assume that the sand consists of spheres. In order to allow somewhat for variation of sizes, assume that in a unit volume of the mixture there are  $N$  spheres of radius  $R$  and  $n$  spheres of radius  $r$ . Now suppose that the heat traverses the mixture by moving in but one direction. Take a cube of unit volume with two faces normal to this direction, and think of it as being traversed by a large number  $m$  of equally spaced lines in this direction. A sphere of radius  $R$  will be intersected by  $m\pi R^2$  of these lines, while one of radius  $r$

TABLE 3—SUMMARY OF THEORETICAL AND EXPERIMENTAL VALUES FOR THREE SANDS AND A SAND MIXTURE

Sand	Density, lb./ft <sup>3</sup>	Temperature, °F	Conductivity, Btu hr-1ft-2in °F-1	
			Theoretical	Experimental
No. 7	91	1000	2.3	2.2
		1500	2.9	2.4
		2000	3.8	3.1
50-70	102	1000	2.8	2.9
		1500	3.6	3.2
		2000	4.8	4.4
20-30	108	1000	3.8	3.6
		1500	5.4	5.0
		2000	7.5	7.2
Mixture 60 per cent 20-30 40 per cent No. 7	111	1000	3.1	2.8
		1500	4.2	3.9
		2000	5.4	5.3

will be intersected by  $m\pi r^2$ . The total number of intersections of lines with spheres is  $Nm\pi R^2 + nm\pi r^2$ . Thus, the average line will intersect

$$N\pi R^2 + n\pi r^2$$

spheres. Now the fraction of such an average line which lies in the air spaces in  $P$ , where  $P$  is the porosity. Thus, the average length of these lines between encounters with spheres is

$$L = \frac{P}{N\pi R^2 + n\pi r^2}$$

Since  $P = 1 - \frac{4}{3}N\pi R^3 - \frac{4}{3}n\pi r^3$ , and the fraction of a unit volume occupied by spheres is  $1-P$ , we find

$$L = \frac{\frac{4rP}{3}}{(1-P) \left[ 1 - f \frac{R-r}{R} \right]}$$

where  $f$  is the fraction of total sphere volume (or weight) consisting of spheres of radius  $R$ . This distance  $L$  will be taken as the air space thickness in estimating the thermal conductivity ( $K_r$ ) of air by radiation.

Russell's theory is now to be applied to certain sands whose conductivities—as mixtures with air—were determined experimentally. The purpose is to decide in a qualitative way whether the measured values were reasonable ones, and in particular whether the variation among them is similar to the theoretical variation. For the conductivity ( $K_s$ ) of the solid particles of sand, we use\*\*

Temperature, °F	$K_s$
	Btu hr-1ft-2in °F-1
1000	29
1500	26
2000	24

(For the density of the solid, take 165 lb. per cu. ft.)

\*\* J. B. Austin, *Symposium on Thermal Insulating Materials*, p. 20, ASTM, Columbus Regional Meeting, March 9, 1939.

\* Journal, American Ceramic Society, vol. 18, pp. 1-5, (1935).

In addition to the notation already introduced, let  $K_e$  be the true thermal conductivity of air,  $K_a = K_e + K_r$  the total conductivity of air, and  $K_b$  the conductivity of the air-sand mixture. Then it follows from Russell's Eq. 1 that

$$K_b = \frac{(1-P)^{2/3} + [1 - (1-P)^{2/3}] K_a / K_s}{[(1-P)^{2/3} - 1 + P] K_s / K_a + [2 - (1-P)^{2/3} - P]}$$

This equation was used to determine theoretical values for the following sands whose conductivities were also determined experimentally.

Type Sand	Density, lb/ft <sup>3</sup>	P	Assumed Radius (in.)	L (in.)	Temp., °F	Conductivity, Btu hr-1ft-2in°F-1				Theo. Kb	Exp. Kb
						Ks	Kc	Kr	Ka		
No. 7	91	0.45	0.0030	0.0033	2000	24	0.51	0.34	0.85	3.8	3.1
					1500	26	0.45	0.16	0.61	2.9	2.4
					1000	29	0.38	0.07	0.45	2.3	2.2
50-70	102	0.39	0.0050	0.0043	2000	24	0.51	0.44	0.95	4.8	4.4
					1500	26	0.45	0.21	0.66	3.6	3.2
					1000	29	0.38	0.09	0.47	2.8	2.9
20-30	108	0.35	0.0140	0.0100	2000	24	0.51	1.03	1.54	7.5	7.2
					1500	26	0.45	0.50	0.95	5.4	5.0
					1000	29	0.38	0.21	0.59	3.8	3.6
60%	111	0.33	60%	0.0037	2000	24	0.51	0.38	0.89	5.4	5.3
20-30			0.0140	0.0037							
40%			40%		1500	26	0.45	0.19	0.64	4.2	3.8
No. 7 Mixture			0.0030		1000	29	0.38	0.08	0.46	3.1	2.8

## STUDIES ON THE SOLIDIFICATION OF ALUMINUM CASTINGS

By  
Dr. Victor Paschkis\*

### 1. Purpose of Experiment

PREVIOUS REPORTS covered a comparison between bleeding tests and tests by the electric analogy method on the solidification of steel and of white cast iron. The present report deals with a similar comparison for the solidification of pure aluminum.

There are two reasons why such an additional comparison is desirable. First, it is the first comparison covering non-ferrous metals; secondly, it is the first time that the electric analogy method is used to study the solidification of a metal which freezes at one temperature rather than over a range. This behavior, common of course to all pure metals, calls for a change in technique in the analogy method.

### II. Selection of Properties

The thermal conductivity of the solid aluminum was assumed to follow Eq. (1):

\* Technical Director of Heat and Mass Flow Analyzer Laboratory, Department of Mechanical Engineering, Columbia University, New York.

$$k = 2.03 \cdot (1 - 0.275 t \cdot 10^{-3} - 0.34 t^2 \cdot 10^{-6}) \dots (1)$$

In this equation  $k$  is expressed in Joules/cm, sec, C. Temperature,  $t$ , is expressed in degrees C. This equation is taken from *International Critical Tables*, vol. 5, p. 220.

For liquid aluminum the conductivity was assumed to follow Eq. (2):

$$k = 0.90 - 0.288 \cdot 10^{-3} \cdot (t - 650) \dots (2)$$

(*International Critical Tables*, vol. 5, p. 220).

The specific heat was assumed to follow Eq. (3):

$$c = 24.7 + 11 \cdot 10^{-3} t \dots (3)$$

Here the specific heat  $c$  is expressed in Joules/gm atom, C, and the temperature  $t$  in degrees C. This equation is taken from page 92, vol. 5, of *International Critical Tables*.

For the liquid state, values were taken from the table transmitted by letter by Dr. Schwartz. In this table for 657 C, a specific heat of 0.2502 cal/gm, C, was given; and for 700 C a specific heat of 0.2523. Thus for a temperature of 677 C, a specific heat of 0.251 was found.

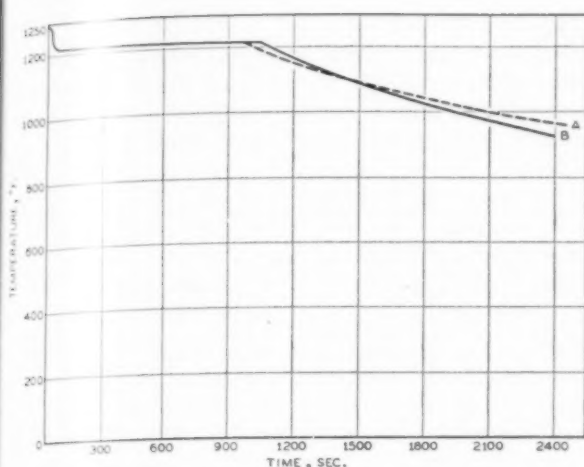


Fig. 2—Comparison of the center temperatures as determined by the electric analogy method (Curve A) and by bleeding tests (Curve B).

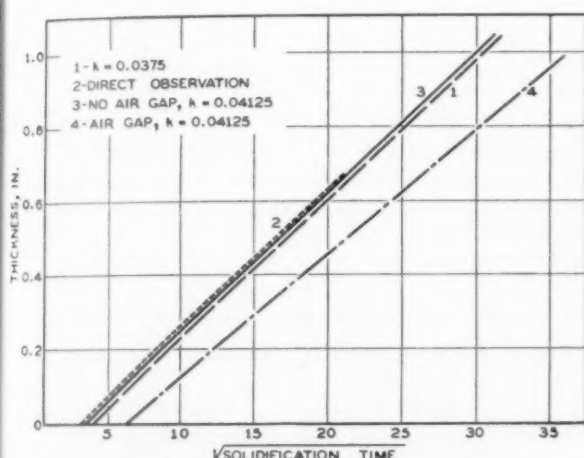


Fig. 3—Comparison of solidified thickness for different time, in seconds.

The density of the liquid was found by interpolation between the values for 658 C and 700 C, transmitted by letter from Dr. H. A. Schwartz. The value for 677 C was found to be 2.377 gm/cm<sup>3</sup>.

The density for the solid aluminum was found by using the third power of the coefficient of the linear expansion. This coefficient is shown on p. 55 of the U. S. Bureau of Standards, Circular No. 346, to be  $33.5 \cdot 10^{-6}/C$ .

The temperatures are as follows:

Pouring Temperature	— 1300 F.
Solidification Temperature	— 1200 F.
Cooling, continued to the lowest solid temperature	— 700 F.

The average temperature for the liquid state is 1250 F (677 C), and the average temperature for the solid state is 950 F (510 C).

The latent heat of fusion was indicated in a communication from Dr. Schwartz to be 93 cal/gm.

For the sand, the experiments were started with

exactly the same properties as those used for white cast iron and for steel. These properties were:

Thermal Conductivity, 0.075 Btu/in.hr.F.  
Specific Heat, 0.28 Btu/lb.F.

After some preliminary experiments the conductivity of the sand was changed. The basis for the change is to be seen in the fact that the pouring temperatures of the aluminum are very much lower than those of white cast iron and steel.

With reference to Table 2 of "Thermal Conductivity of a Sand Mixture," by C. F. Lucks, O. L. Linebrink and K. L. Johnson of Battelle Memorial Institute, the following relationship holds:

Average Conductivity at 2250 F, 6.5  
Average Conductivity at 1250 F, 3.2

These values are given in Btu/hr.sq.ft, F/in. The values hold for a sand mixture of 60 per cent 20-30, and 40 per cent No. 7. In view of this relationship, it was decided to run the final tests on aluminum with half the conductivity used in the steel and cast iron experiments, namely, a conductivity of only 0.0375 Btu/in.hr.F.

### III. Comparison of Results

The experiments were carried out for a casting consisting of a 2-in. thick slab, long and wide enough so that the end effects could be neglected. In the work by the electric analogy method an infinite slab of 2 in. was assumed. The castings made at the Cleveland Research Division of the Aluminum Company of America were slabs 2 x 12 x 12 in. and it was hoped that no end effects would occur.

In the report "Solidification Rates of Aluminum in Dry Sand Molds"\* H. Y. Hunsicker reports in Fig. 36 the time-temperature curve of the casting, and in Fig. 34 the average solidified thickness plotted against the square root of time.

Figure 2 of this report shows the center temperature of the aluminum plotted against time. Curve A shows the values found on the Heat and Mass Flow Analyzer; Curve B the values plotted from the above mentioned paper. It may be seen that the two curves coincide closely. Figure 3 shows the solidified thickness plotted against the square root of time. This chart contains four curves. Curve 1 shows the values obtained with the correct sand conductivity. Curve 2 shows the values found by Hunsicker. Curves 3 and 4 will be discussed later. Again it may be seen that Curves 1 and 2 check well, although Curves 3 and 2 check better.

### IV. Variation of Properties

Because the sand properties are not too accurately known, it was considered worth while to vary the sand properties slightly in the electric analogy experiments to see what effect it might have.

Moreover, it was desired to find out what influence formation of an air gap would have. Finally it was desired to determine what influence the time lag, caused by a thermocouple, would have.

\* TRANSACTIONS, AFA, vol. 55, p. 68 (1947).



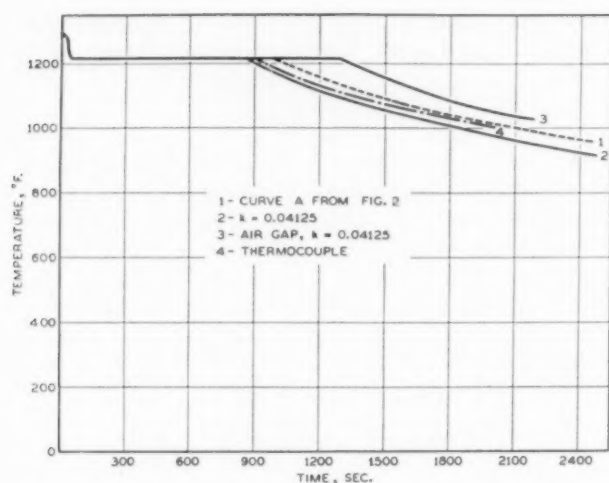


Fig. 4—Comparison of temperatures in the aluminum for various conditions.

(a) Change of Thermal Conductivity of the Sand.

In Fig. 4 the aluminum temperatures are plotted against time. A total of four curves is shown: Curve 1 repeats the Curve A from Fig. 2, conductivity of 0.0375 Btu/in,hr,F. Curve 2 holds for a thermal conductivity of the sand of  $k = 0.04125$ . This value was found by adding 10 per cent to the original conductivity value. In Fig. 3, Curve 3 shows the solidified thickness plotted against the square root of time, for a value of  $k = 0.04125$ .

It will be seen that the influence on the solidification time is relatively small. Total solidification time changes from 940 sec to 900 sec. The influence on the center temperature is somewhat more marked. Thus, the aluminum temperatures seem to check better with the values observed in bleeding tests if a conductivity of 0.0375 is assumed; but the solidification times check better with bleeding tests if a conductivity of 0.04125 is assumed. In either case the differences are not very large.

(b) Influence of an Air Gap

As a next step the influence of an air gap, formed immediately upon pouring at the aluminum-sand interface, was investigated. Heat flow across the air gap was assumed to take place by radiation. The corresponding curve for aluminum temperatures is shown in Fig. 4 as Curve 3; the progress of solidification is shown as Curve 4 in Fig. 3. It will be noted that the influence of the air gap is quite noticeable. In comparing this result with that reported upon in the TRANSACTIONS A.F.A., vol. 53, p. 90, Fig. 2 (1945), it should be kept in mind that the radiation at the lower temperatures, in the case of aluminum, makes for a much higher resistance than radiation at the high temperatures, in the case of steel. Therefore, a relatively larger influence of the air gap in the case of aluminum casting, is to be expected compared with the steel casting. The influence might be slightly decreased if the thermal conduction across the air gap would be taken into consideration in addition to the radiation. Such additional heat flow would also tend to bring Curve 4 in Fig. 3 closer to Curve 1.

(c) Thermocouple

It was thought worthwhile to investigate the possible influence of the time lag of the thermocouple. For this purpose a ceramic protection tube was assumed to have the following dimensions and properties:

Outside Diameter	— 1/2 in.
Thickness	— 1/8 in.
Thermal Conductivity	— 0.075 Btu/in,hr,F.
Specific Heat	— 0.28 Btu/lb,F.
Density	— 0.0811 lb/cu in.

It was assumed that the temperatures in the aluminum might be influenced up to a distance of 1 in. from the thermocouple.

Obviously all these assumptions are arbitrary, but it was felt that the results might yield some information as to the order of magnitude of error which can be expected from using the thermocouple. Curve 4, Fig. 4, shows the results, and shows that the influence is not very large.

## V. Sand Temperature

Temperatures at various points in the sand were observed by the electric analogy method, and are plotted in Fig. 5. In the paper by Hunsicker previously mentioned no sand temperatures are shown for the aluminum casting. Therefore, no direct comparison is possible. However, some rough estimates can be made by comparing the temperatures observed for the aluminum alloy.

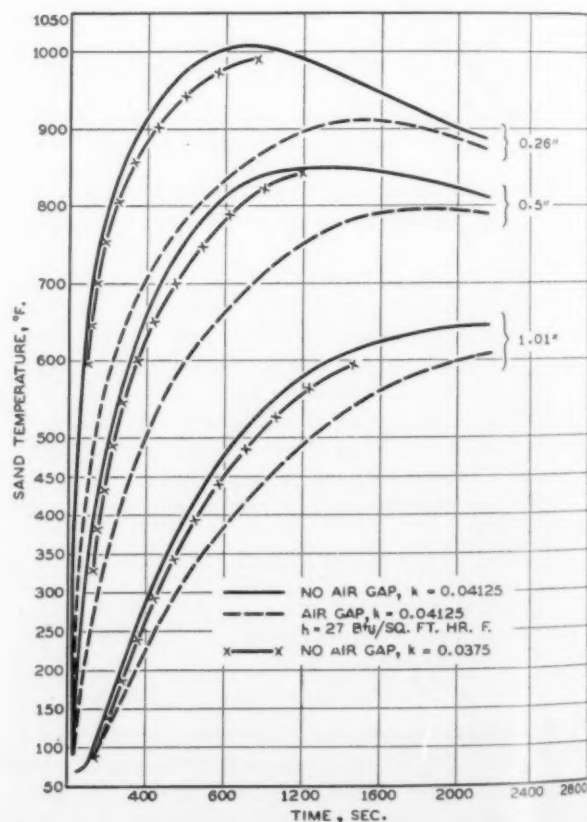


Fig. 5—Sand temperature for aluminum casting. Thickness of casting, 2 in.; thickness of sand, 4 in.

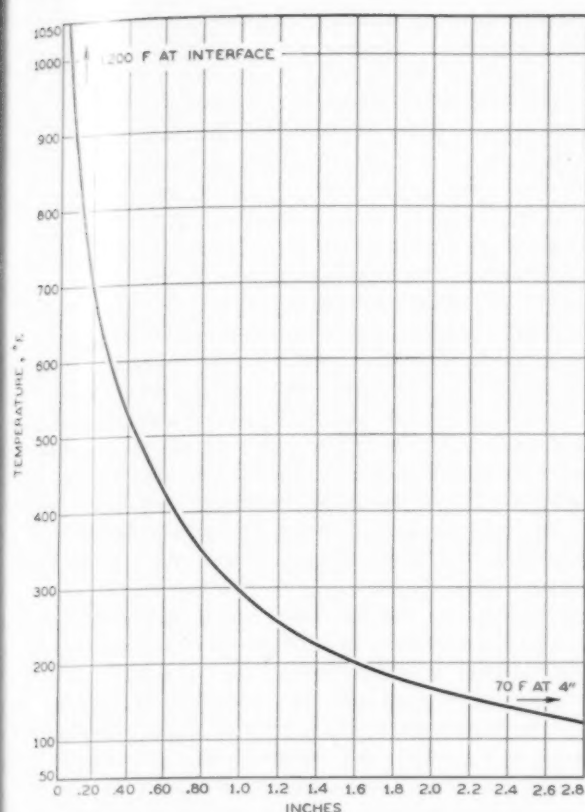


Fig. 6—Temperature distribution in sand, 600 sec. after pouring. Aluminum alloy casting 2 in. thick; sand, 4 in. thick.

Figure 5 comprises three sets of curves, two for the two thermal conductivities used, and one for the air gap.

In order to compare with the aluminum alloy tests, the following figures are mentioned:

In Hunsicker's paper, Fig. 30: Values for a  $\frac{1}{4}$ -in. distance are shown as  $A_1$  in the following table. Values for a distance of 0.26 in. from the interface found from the electric experiments on pure aluminum, with a sand conductivity of  $k = 0.0375$ , are shown in the table as  $E_1$ . Similarly, values for a  $\frac{1}{2}$ -in. distance are shown as  $A_2$  and  $E_2$ . Values for the aluminum alloy of 1-in. distance are shown as  $A_3$ , and those for the electric analogy method, for a distance of 1.01 in. from the interface, as  $E_3$ .

Time (Sec.)	$A_1$	$E_1$	$A_2$	$E_2$	$A_3$	$E_3$
200	360	765	225	450	130	130
400	535	885	375	630	220	265
600	640	942	475	720	295	370
800	685	980	535	785	355	450
1000	705	995	580	822	400	510

#### VI. Analysis of the Aluminum Alloy Casting

In Hunsicker's paper curves are shown for the sand temperature in an alloy casting having dimensions  $2 \times 8 \times 8$  in. The paper also shows the progress of solidification with time.

It appeared desirable to duplicate on the Heat and Mass Flow Analyzer the solidification of the aluminum alloy casting. In this connection a tentative heat balance was set up based upon the temperatures and figures shown in Hunsicker's paper. This heat balance was carried out for each of the following two assumptions.

(1) All heat flow to be in a direction perpendicular to the long sides of the casting.

(2) Two-dimensional heat flow from the casting to the sand, including the corner effects in the casting.

In both investigations heat flow to the top and bottom was considered to be negligible in view of the riser and the pouring gate.

The heat balance was made for the moment 600 sec. after pouring. At this time the temperature curve in the aluminum alloy begins to drop, indicating that the superheat and the heat of fusion have been given off at this time and the entire mass of the aluminum alloy is at 1200-F in a solid state. In order to determine the heat given off up to this moment, it was assumed that during the cooling from the pouring temperature to solidification temperature the alloy is at "liquid density," and that during the solidification it is at a density equal the arithmetic mean of the solid and liquid densities. With the properties shown in Section

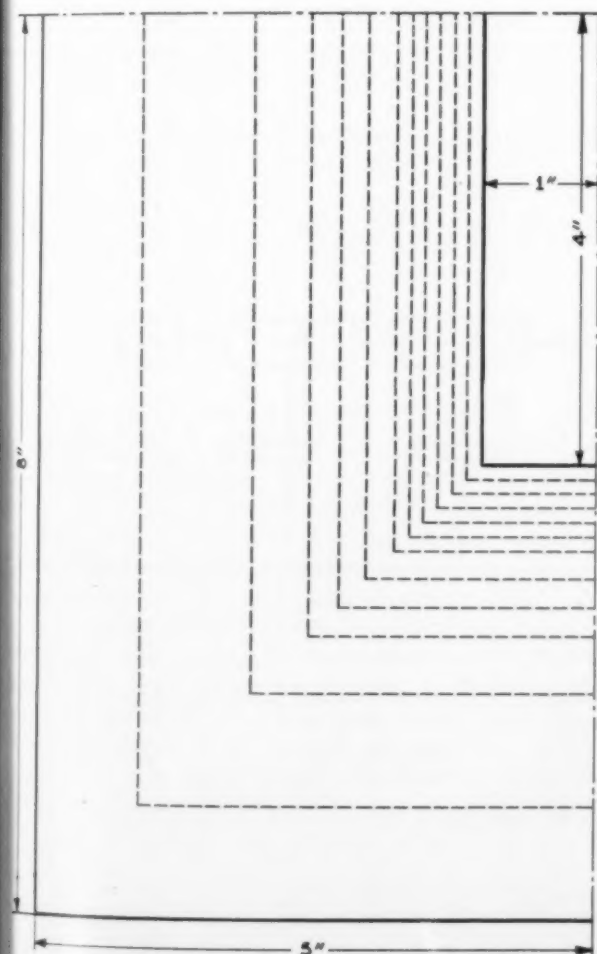


Fig. 7—Cross section of one-fourth of the casting.

11 of this report (assuming that they hold also for the aluminum alloy) the heat given off by the steel can be calculated as follows:

- (1) Loss of Heat Content = sensible heat + latent heat of fusion.
- (2) Loss of Heat Content = cu.in. x [(Btu/lb, F) x F x [lb/cu in.] + Btu/lb x lb/cu in.]
- (3)  $16 \times (0.251 \times 100 \times 0.0855 + 167.4 \times 0.08925) = 273.4$  Btu/in. height of casting.

This is the heat absorbed by the sand. From the curves in Hunsicker's paper the temperature distribution 600 sec. after pouring was determined and is shown in Fig. 6. Figure 6 shows only 2.8 in. of the sand and the curve would have to be extended from 2.8 to 4 in. at which point the sand is assumed to be at 70 F.

By integrating the area under this curve an amount of 712.63 degrees-in. has been found. Multiplying this figure by 16 in. for the two 8-in. faces of the casting, and by the specific heat and density of the sand as given in the report, the total heat content of the sand of 158.4 Btu/in. height of casting may be found. This figure is obviously less than 60 per cent of the heat lost by the casting.

The second heat balance, which is two-dimensional and thus includes the corners, may be understood from Fig. 7, which shows one-quarter of the casting and the surrounding sand. Of course the casting is symmetric about its axes, therefore the part shown

would have to be multiplied by 4. In the drawing, the individual sections in the sand are shown closely spaced near the surface where the temperatures are high and wider spaced towards the outside. The calculation was carried out as follows: The average temperature for any one section (for example 940 F for the first increment of  $\frac{1}{8}$  in. from the aluminum) was multiplied with the cross-sectional area of that section (for example,  $8.25 \times 2.25 = 8 \times 2$  for the first increment next to the aluminum.) The sum of "deg. x sq. in." was in turn multiplied by the specific heat x density. In this way the entire heat content was determined. The heat contained in the sand, assuming two-dimensional heat flow, is found to be 202.7 Btu/in. height of the casting. In this determination it was assumed that the temperature of the sand is uniform all over the circumference of the casting. Actually the temperatures in the corners will be lower and thus the true figure will be lower than 262.7. On the other hand, there is also heat lost in the third dimension, although less than in the two dimensions considered in the calculations. The heat flow in the third direction will be less because there is some heat gain from the riser and the gate.

The close proximity of the values 262.7 and 273.4 of the heat contents of sand and aluminum under the specified assumptions indicates that the  $2 \times 8 \times 8$  in. casting was not big enough to eliminate corner effects, at least as far as sand temperatures are concerned. Therefore, no basis for checking results from the Heat

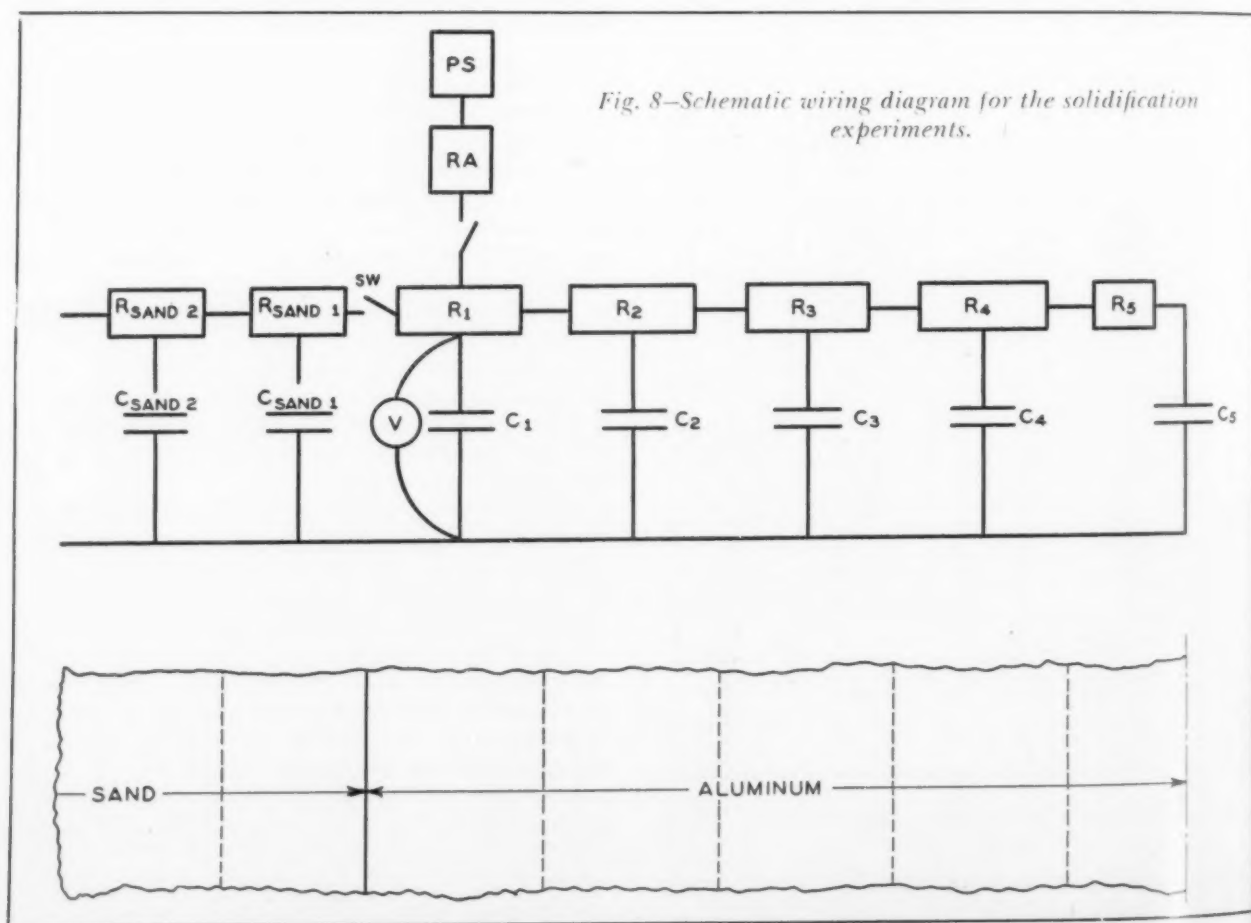


Fig. 8—Schematic wiring diagram for the solidification experiments.



and Mass Flow Analyzer is available and no tests for the aluminum alloy were carried out.

#### VII. Analyzer Technique for Studying Solidification at Constant Temperature

In the Heat Transfer report published in *TRANSACTIONS, A.F.A.*, vol. 53, p. 90 (1945), the technique of the electric analogy method was described in some detail. The experiments covered in that report deal with the solidification of steel, which freezes over a finite temperature range. However, aluminum, as any pure metal, solidifies at a constant temperature. Instead of representing the heat of solidification by an electric condenser, as in the solidification of steel, the heat of solidification was now represented by a power supply with constant voltage. Figure 8 shows part of the circuit used.  $R_1$ ,  $R_2$ , etc. represent the thermal resistance of the various sections of the casting, the casting being shown on the diagram as part of an infinite slab. The individual sections are marked in the lower part of the figure by dotted lines. For the sake of simplicity, the resistors are shown as constant resistors, whereas actually they were changed from the liquid to the solid state to take into account the change in thermal conductivity upon solidification. The electric condensers  $C_1$ ,  $C_2$ , etc., represent the thermal capacities of each section. The thermal capacitance is the product of specific heat  $\times$  density  $\times$  volume of each section. Again only one condenser is shown for each section whereas actually several condensers were used to represent the change of specific heat from liquid to solid. To the left of the aluminum circuit is shown the beginning of the sand circuit, which again consists of resistors and condensers to

represent the thermal capacity and thermal resistance of the various sand sections. At time 0 (the pouring time) the switch  $Sw$  was closed. Prior to this closing of  $Sw$  the entire aluminum circuit had been loaded to a voltage equivalent to the pouring temperature of 1300 F. Upon closing  $Sw$ , the voltage (equivalent to the temperature) of the "casting circuit" dropped and the temperature in the first section was observed by means of a voltmeter  $V$ . As soon as the temperature had reached the solidification point of 1200 F, a power supply  $PS$  was connected to this point. The power supply was so set as to maintain a constant voltage equivalent to the solidification temperature of 1200 F. From the power supply current was flowing into the "casting circuit" and this current was reported on the recording milliammeter marked  $RA$  in Fig. 8. The current-time curve was integrated and the power supply was disconnected as soon as enough millicoulombs (milliampere seconds) had been transferred into the casting circuit. Millicoulombs in the electric analogy are equivalent to Btu and the amount of millicoulombs had been so determined as to correspond to the heat of fusion of this section. Obviously the next section could not drop in temperature to the solidification point before the first section had absorbed the entire heat of fusion. Upon absorbing the entire heat of fusion, the power supply and the recording milliammeter were switched to the second section, and the procedure repeated.

It should be noted that notwithstanding the fact that only one section at a time can give off the heat of fusion, the findings of Hunsicker were confirmed that there is no measurable temperature difference during solidification between the first and the last section.

## STUDIES OF SOLIDIFICATION OF WHITE IRON CASTINGS

By  
Dr. Victor Paschkis\*

### I. Purpose of Experiment

IN PREVIOUS YEARS the work of this laboratory was concentrated mainly on steel castings. It has been shown that for steel castings the freezing rates obtained in this laboratory by the electric analogy method check well with results obtained from bleeding tests. Such a close check may be expected however only if for the materials under consideration and the thermal properties used in calculating the circuits for the electric experiments are known. It therefore appears desirable to investigate how closely properties for other materials than steel are known. If the properties are known with sufficient accuracy, then the bleeding tests should result in the same solidification

curves as those found on the analyzer by the electric analogy method.

The work was again carried out by the electric analogy method and reference is made to previous publications.<sup>1</sup>

The present program comprises a study of the solidification of casting white cast iron. Reference is made to the report of Dr. H. A. Schwartz, "Freezing Rate of White Cast Iron in Dry Sand Molds," *TRANSACTIONS, A.F.A.*, vol. 55, p. 66 (1947).

### II. Dimensions and Temperatures

As may be seen from this report, the casting which was investigated by Dr. Schwartz was a slab 4 x 8 x 8 in.

In the electrical experiments the center plane of the identical slab was tested; it was assumed that no end effect occurs and that therefore the center of the slab may be considered as part of a semi-infinite slab. The

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thickness of the sand on either side of the casting was 10 in.

The report by Dr. Schwartz shows in Table 16 amongst other items the pouring temperature. The average for the eleven castings is 2524 F. In previous correspondence with Dr. Schwartz this laboratory was instructed to base experiments on a casting temperature of 2500 F. In view of the higher temperature mentioned in the aforementioned report of Dr. Schwartz, two experiments were run, one with a casting temperature of 2500 F and the other one with 2524 F.

### III. Rates of Solidification

In Table 17 of Dr. Schwartz's report, frozen thicknesses are shown. Although the half-thickness is 2 in., no values are shown beyond 0.88 in. The difficulties of bleeding iron castings account for this omission and are described in detail in the above mentioned report. Consequently, for thicknesses beyond 0.88 in. no values are available to compare the solidification obtained on the analyzer with those obtained by bleeding tests.

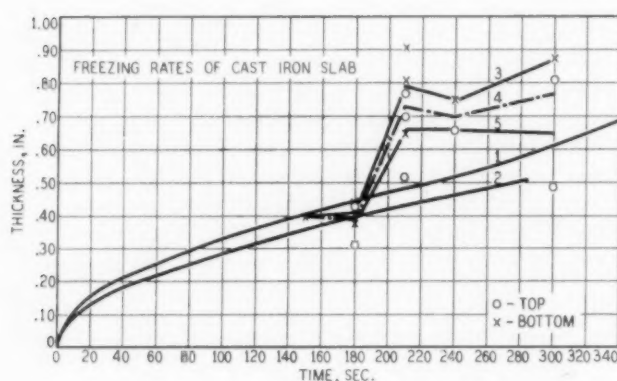


Fig. 9—Freezing rates of cast iron slab.

In Fig. 9 the solidified thickness is plotted against time. Five curves are shown, numbered one to five, the numbers having the following meaning:

- (1) Values obtained on the Analyzer for 2500 F pouring temperature.
- (2) Values obtained on the Analyzer for 2524 F pouring temperature.
- (3) Results obtained by bleeding tests, top.
- (4) Results obtained by bleeding tests, bottom.
- (5) Results obtained by bleeding tests, average between top and bottom.

In the figure the individual measurements, as tabulated in Dr. Schwartz's report in Table 16, are shown. The wide scattering of the points is obvious. The ordinate scale in Fig. 9 extends only to 1 in., because no values beyond 1 in. were obtained by bleeding tests. This chart allows therefore a close comparison between the electric tests and the bleeding tests. The complete electric tests are shown in Fig. 10; the curve for top freeze from the bleeding tests is repeated because this curve seems to check close with the electrical experiments.

### IV. Comparison of Results

Dr. Schwartz's previously mentioned report states that probably the top curve is the true solidification curve. The "electric" points seem to fit reasonably well and only in a few points seem to lie lower than the "top" curves from the bleeding tests. The number of bleeding tests is not sufficient to permit a precise comparison between electric and bleeding tests. The shapes of the solidification curves from the bleeding tests are not consistent with physical facts, and it is highly improbable that a curve of this shape would occur if the bleeding tests could be made with sufficient accuracy. On the other hand, the freezing rates at the top are probably more nearly correct than those at the bottom. In view of the casting difficulties only a statistical averaging of the results from many molds could give accurate results.

### V. Properties

The outcome of freezing rates by the electrical analogy method depends to a considerable extent on the properties. The analyzer determines with reasonable accuracy the solidification rates for any property that is fed in, independent of the correctness of this property for any specific metal, for example for cast iron. The following properties were selected from the sources indicated in the tests:

#### A—Cast Iron

##### (1) Thermal conductivity:

Liquid 8.5 Btu/ft hr F

Solid 17 Btu/ft hr F

The value for the "solid" was taken from E. Soehnchen, *Archiv F.D. Eisenhuettenwesen*, vol. 8, p. 223 (1934-35). The values given by Soehnchen (Fig. 8) for 1 per cent Si and 3 per cent C were extrapolated for a temperature of 1000 C. The liquid conductivity was assumed to be  $\frac{1}{2}$  the value of the solid, in accordance with the property of steel.<sup>2</sup>

##### (2) Specific Heat:

Liquid 0.20 Btu/lb. F

Solid 0.165 Btu/lb F

The value for the "solid" was taken from Umino, *Science Reports, Tohoku Imperial University*, vol. 15, p. 331 (1926). The reasons for selecting the liquid to 0.200 are the same as those mentioned in (1).

##### (3) Density:

Liquid 449 lb/cu ft

Solid 0.165 lb/cu ft

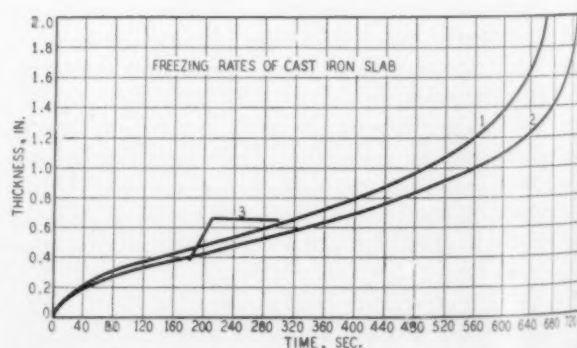


Fig. 10—Freezing rates of cast iron slab.

These values were taken from Vol. II, *The Alloys of Iron and Carbon* by F. T. Sisco, p. 270 (McGraw-Hill, 1937). The values are interpolated between a curve for 3.10 per cent C, 1.69 per cent Si, and another for 2.1 per cent C, 1.24 Si.

(4) Heat of Fusion: 99 Btu/lb

For pure iron, H. E. Cleaves and J. G. Thompson in *The Metal Iron*, (McGraw-Hill 1935), recommend, in accordance with Austin as most probable, a value of 65 cal/g.

Umino (*Proc. ASTM*, vol. 33, p. 161 (1933), gives 47 cal/g for a carbon content of 4.22 per cent. Assuming that the heat of fusion is proportional to the carbon content, the heat of fusion for 2.5 per cent C is  $\frac{2.5}{4.2} (65 - 47) = 58$  cal/g or 104 Btu/lb.

(5) Liquidus Temperature 2282 F (1250 C)

Solidus Temperature 2088 F (1142 C)

indicated by Dr. Schwartz by letter to the author.

B—Sand

Inasmuch as in the bleeding tests the same sand was used as in the NRL tests for steel castings, the properties of the sand were assumed to be identical with those used previously in the analysis of the steel casting tests.<sup>2</sup> Properties were as follows:

Thermal Conductivity: 0.9 Btu/ft hr F  
Specific Heat 0.28 Btu/lb F  
Density: 93.6 lb/cu ft

Summary

Tests on freezing of cast iron made on the electric analyzer check reasonably with bleeding tests carried out by the National Malleable and Steel Casting Company.

The latter were carried out only to a solidified thickness of roughly 35 per cent of the half-thickness of the slab. In view of the experimental difficulties of the bleeding tests, the curves obtained from the latter are too irregular to permit a definite statement of the validity of properties used in the electric tests.

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1. Victor Paschkis, "A Method for Determining Unsteady-State Heat Transfer by Means of an Electrical Analogy," *Transactions*, ASME, vol. 64, p. 105 (1942).

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"Heat Flow Problems in Foundry Work," *TRANSACTIONS*, American Foundrymen's Association, vol. 52, p. 649 (1944).

2. Victor Paschkis, "Studies on Solidification of Castings," *TRANSACTIONS*, American Foundrymen's Association, vol. 53, p. 90 (1945).

## STUDY ON SOLIDIFICATION OF STEEL SPHERES

By

Dr. Victor Paschkis\*

### I. Purpose and Scope of Experiments

Previous work<sup>1, 2</sup> with the Heat and Mass Flow Analyzer (HMFA) on solidification of castings was limited to slabs which were so big that the end effect could be neglected. In the first study,<sup>1</sup> as well as in work not connected with steel castings<sup>3, 4</sup> experiments on the HMFA were always compared with bleeding tests, carried out specifically for the purpose of such comparison.

For two reasons it appeared desirable to study the solidification of steel spheres by comparing HMFA experiments with experiments carried out by Briggs and Gezelius, henceforth referred to as B & G.<sup>5</sup>

First, it is important to establish cooling curves for spheres, and secondly, it is of interest to carry out a comparison with experiments published relatively long ago, and not made specifically for the purpose of comparison with HMFAL (Heat and Mass Flow Analyzer Laboratory) Tests.

B & G investigated several spheres and made various generalizations. The HMFA experiments reported hereafter were concentrated on 4½-in. and 6-in. spheres. In view of the wealth of material produced in the B & G experiments, the publication<sup>5</sup> was necessarily limited in size and did not bring complete data. For reasons explained in Section II it was necessary to go back to the original values, which Mr.

C. H. Briggs put at the disposal of Dr. H. A. Schwartz. As far as necessary, parts of these original data will be quoted hereafter.

### II. Briggs and Gezelius Solidification Values

In order to arrive at a simple presentation, B & G averaged results from a number of tests made under rather different conditions.

They neglect the corner effect (Section 38) and thus treat a "two-dimensional" or even "three-dimensional" problem as if it were of the "one-dimensional" type: one, two or three dimensions refers to the directions of heat flow from the center of the casting. Inasmuch as this report does not deal with rectangular castings, this simplification will not be analyzed here.

Steels of different compositions were used and the results mixed (Section 42). This is probably permissible,<sup>2</sup> but differences of from 10 to 15 per cent in solidification times must be expected.

Finally the results of castings made at different temperatures were averaged (Section 44). Because the validity of such averages is open to question, copies of the original tables were secured from Mr. Briggs. The tables are numbered I-VI, and 9-12. All tables except number 12 refer to 6-in. spheres and refer to a bleeding time, indicated in the head of the table. Table 12 refers to 4½-in. spheres and contains data for various bleeding times.

Each table contains the following information: Heat Number, Analysis, Thickness (In.), Weight (lb.),

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Lapsed Time (Min:Sec), Fluidity (In.), Temperature ( $^{\circ}$ F).

In the present comparison only the two italicized items deserve special attention.

From previous work<sup>2</sup> it is suspected, that the pouring temperature has particular influence on the solidified thickness. Therefore the new results are correlated with those of B & G using the actual pouring temperatures. B & G used a large variety of pouring temperatures, varying from 2650 F to 3010 F. Inasmuch as experiments on the HMFA were carried out for only a limited number of different pouring temperatures, not every casting of B & G could be used for comparison. Those with pouring temperatures equal or close to the ones used on the HMFA are listed in Table 4 of this paper.

### III. Selected Properties

The HMFA may be considered to be a calculating machine for the solution of heat flow problems. As in any calculating machine the answers are no better than the properties fed in. In a previous report,<sup>2</sup> the relative importance of various properties had been established; based on these findings, the heat of fusion of the steel and the thermal conductivity, specific heat and density of both the steel (solid and liquid) and the sand were selected and not changed in any of the experiments. Pouring temperature and solidification range, however, have considerable influence on the solidification rate and therefore were changed systematically.

The following values were used:

Thermal conductivity (Btu/ft, hr, F)

steel (liquid) 9.2

steel (solid) 18.4

sand 0.9

Specific heat (Btu/lb, F)

steel (liquid) 0.200

steel (solid) 0.165

sand 0.28

Density (lb/cu ft)

Steel (liquid) 450

steel (solid) 468

sand 93.6

Heat of fusion (Btu/lb) 126.2

Range of solidification and pouring temperatures used in the experiments are listed in Table 5.

No pouring temperature below 2740 F and no liquidus temperature below 2700 F were introduced in the electric experiments, because they would be in definite contradiction to the iron-carbon diagram. It is not readily understandable how B & G could operate in some instances with pouring temperatures of as low as 2640 F.

### IV. Experimental Results

All experiments were carried out on the HMFA. The technique used has been described previously<sup>1, 2, 6</sup>. Modifications of the technique for the analysis of spheres are described in the appendix. In the case of spheres it is not possible to obtain the center temperature by the method of electric analogy, but only

TABLE 4—CORRELATION OF B & G AND HMFA TEST DATA

Point Number in Present Report	B & G		Pouring Temp. ( $^{\circ}$ F)	Solidified Thickness (In.)
	Heat Number	Table		
1	98	11	2900	1.77
2	142	9	2890	0.74
2a	142	9	2900	0.75
3	101	10	2890	1.55
4	78	11	2910	0.500
6	68	VI	2900	1.36
7	144	12	2750	0.41
8	144	12	2760	0.49
9	144	12	2730	0.52
10	144	12	2720	0.55
11	144	12	2800	0.67
12	146	12	2800	1.59
13	144	12	2810	0.65
14	144	12	2820	0.52
15	146	12	2820	1.55
16	144	12	2910	0.39
17	144	12	2905	0.31
18	89	II	2720	0.378
19	77	III	2760	0.550
20	142	9	2750	0.75
21	98	10	2760	1.59
22	98	11	2750	1.83
23	89	II	2790	0.368
24	77	III	2790	0.515
25	86	III	2790	0.507
26	78	III	2800	0.553
27	88	V	2810	0.985
28	79	V	2810	1.005
29	142	9	2800	0.75
30	101	10	2800	1.50
31	98	10	2805	1.55
32	89	II	2835	0.378
33	78	III	2815	0.535
34	82	VI	2820	1.21
35	98	10	2835	1.53
36	83	III	3010	0.456

TABLE 5—RANGE OF SOLIDIFICATION AND POURING TEMPERATURE DATA

Pouring Temperature	Ranges of Solidification
2740	2732-2597; 2700-2600
2800	2732-2597; 2700-2600
2820	2732-2597; 2700-2600
2900	2732-2597; 2700-2600; 2822-2732
3010	2732-2597; 2700-2600; 2822-2732

a point a finite distance from the center. Experimental difficulties increase as the magnitude of this finite distance decreases. In the present experiments the last observed points are  $0.105r$  distance from the corner, where  $r$  is the radius of the sphere.

Presentations of the results are grouped according to pouring temperature as follows:

Pouring Temperature ( $^{\circ}$ F)	Figure Number for:	
	6-In. Sphere	4½-In. Sphere
2740	11	16
2800	12	17
2820	13	18
2900	14	19
3010	15	20

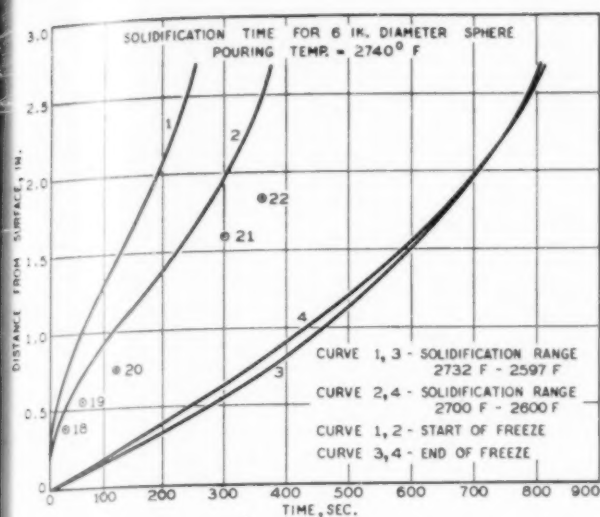


Fig. 11—Solidification time for 6-in. diameter sphere.  
Pouring temperature, 2740° F.

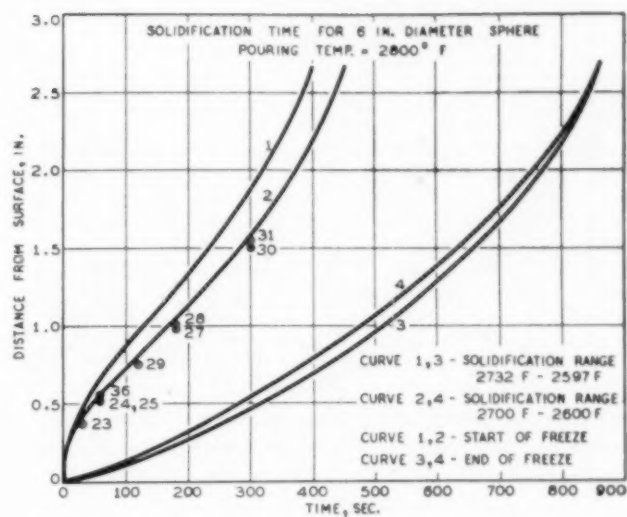


Fig. 12—Solidification time for 6-in. diameter sphere.  
Pouring temperature 2800° F.

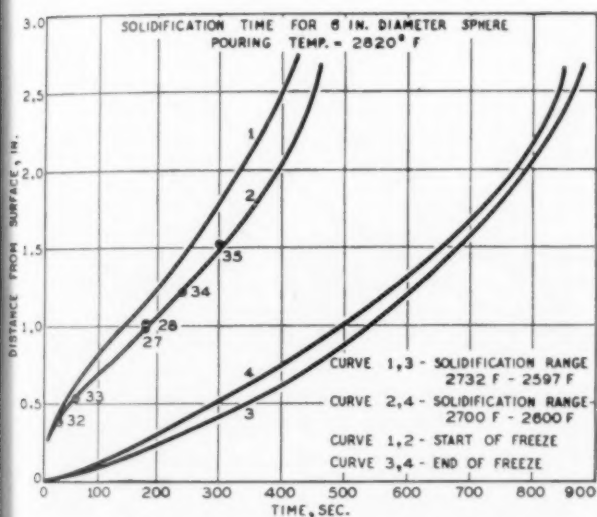


Fig. 13—Solidification time for 6-in. diameter sphere.  
Pouring temperature, 2820° F.

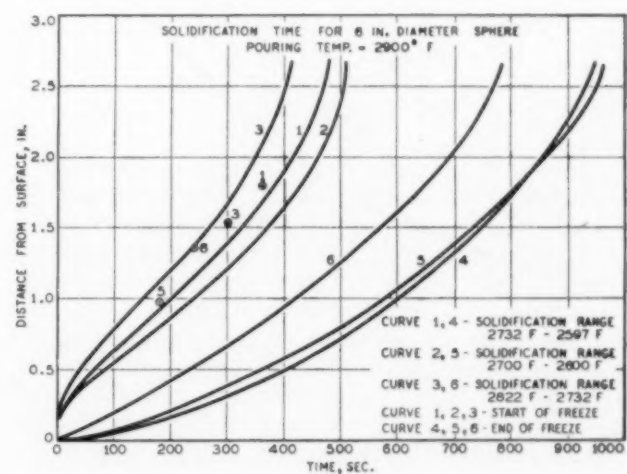


Fig. 14—Solidification time for 6-in. diameter sphere.  
Pouring temperature, 2900° F.

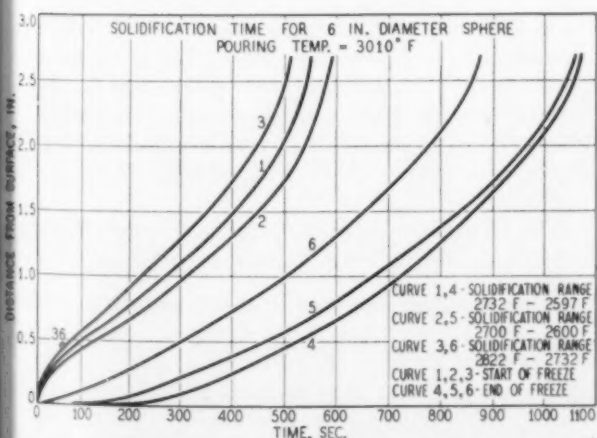


Fig. 15—Solidification time for 6-in. diameter sphere.  
Pouring temperature, 3010° F.

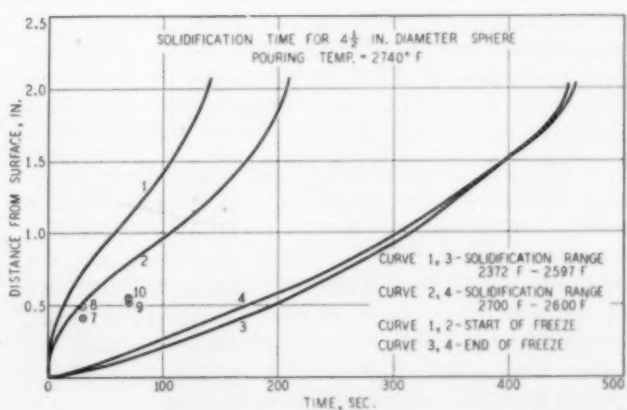


Fig. 16—Solidification time for 4 1/2-in. diameter sphere.  
Pouring temperature, 2740° F.

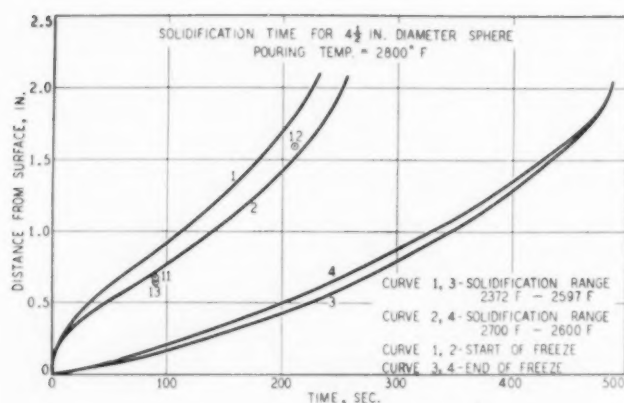


Fig. 17—Solidification time for  $4\frac{1}{2}$ -in. diameter sphere. Pouring temperature, 2800° F.

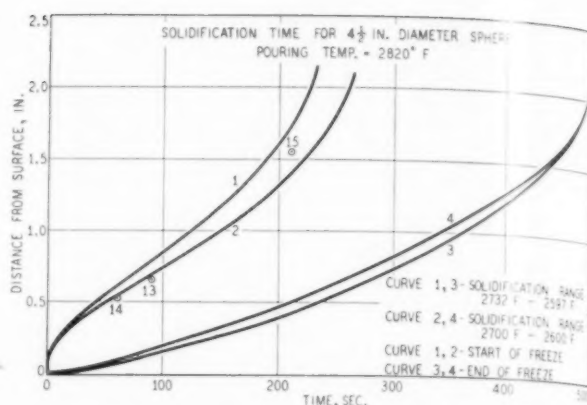


Fig. 18—Solidification time for  $4\frac{1}{2}$ -in. diameter sphere. Pouring temperature, 2820° F.

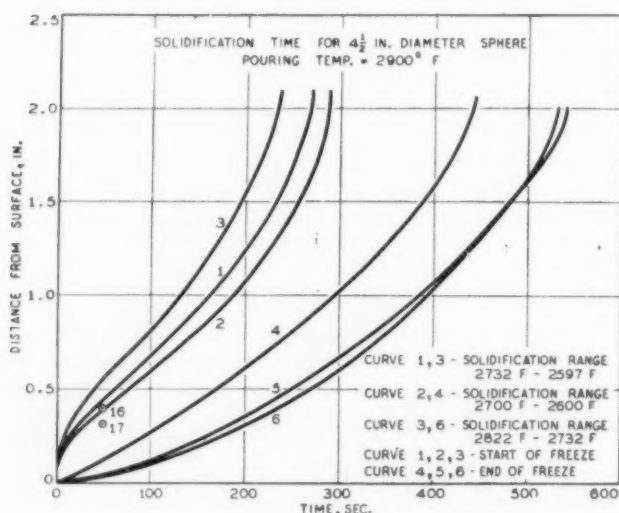


Fig. 19—Solidification time for  $4\frac{1}{2}$ -in. diameter sphere. Pouring temperature, 2900° F.

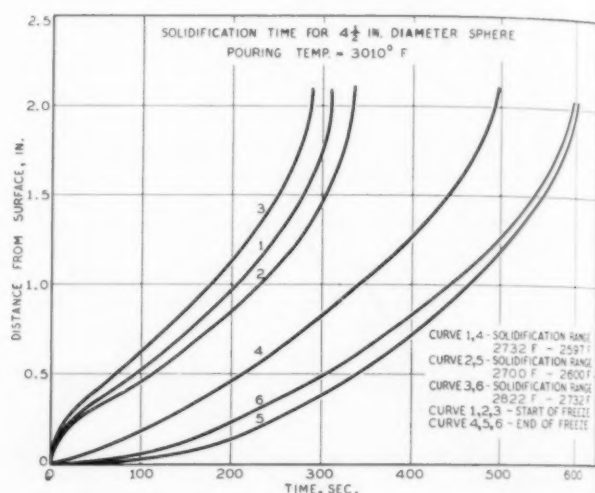


Fig. 20—Solidification time for  $4\frac{1}{2}$ -in. diameter sphere. Pouring temperature, 3010° F.

On each chart, two groups of curves are shown, one marked as "start of freeze" the other as "end of freeze." The former indicates the time (abscissa) required at various values of thickness (ordinate) to reach the liquidus line; the latter ("end of freeze") the corresponding times to reach the solidus line. Previous experience shows that the times to reach the liquidus line correspond to those obtained from bleeding tests. Inasmuch as the experimental time required to reach the last points of the "end of freeze" curves are in some instances quite long, these curves were not always carried to the end.

In each group on all charts several curves are shown; each curve holds for a different combination of solidification range and pouring temperature. Inasmuch as not all ranges start at the same (liquidus) temperature, the curves for the different ranges do not follow in order of range (e.g. in Fig. 14 the 135 F range, from 2732 to 2597 results in solidification times between those of the 90 F range from 2822 to 2732 and that of the 100 F range from 2700 to 2600).

Points from B & G tests are recorded by numbers, the latter referring to Column 1 in Table 4.

Except for the 2740 F temperature, the values of B & G seem to check very well with those from the HMFA. Why those at 2740 F do not check so well cannot be said with certainty.

One reason may be that the temperature measurement was inaccurate. The temperature of 2740 F is quite close to the liquidus temperature, and small inaccuracies in measurement might result in considerable change in time. Another explanation is that the composition of the steel was appreciably different from that poured at other temperatures possible by segregation in the ladle during pouring the preceding castings.

The curves can be used for any size sphere by applying the following scale factors to those of the 6-in. sphere. Multiply the solidified thickness by  $\frac{6}{d}$ , where  $d$  is the diameter of the new sphere to be examined and multiply the abscissa (time) by  $\frac{d^2}{36}$ . Thus the solidification of a 9-in. sphere to a depth of 6 in. may be found by reaching the ordinate scale for  $6 \times \frac{6}{9} = 4$



in. Read the abscissa value for the desired  $d$  curve and multiply the time by  $\frac{81}{36} = 2.25$ .

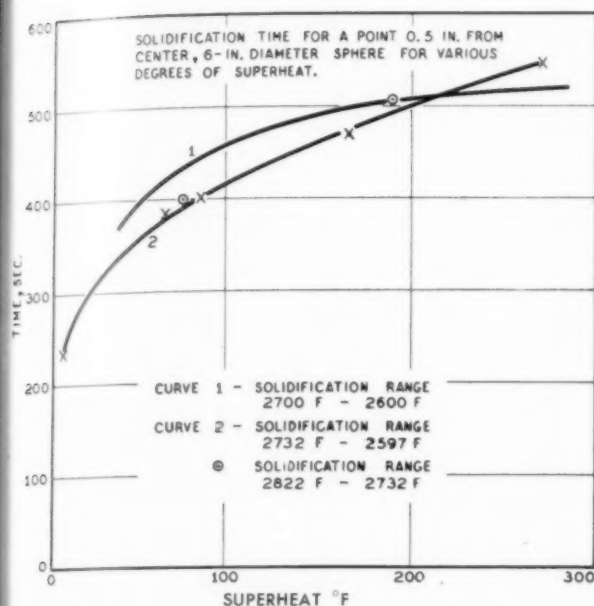


Fig. 21—Solidification time for a point 0.5 in. from center, 6-in. diameter sphere for various degrees of superheat.

In Fig. 21 the solidification times are plotted against degrees superheat; curve 1 holds for the range 2700 F, and curve 2 for the range 2732-2597 F. The curves show the time to solidify to a depth of 2.5 in. on the 6-in. sphere. Two points, indicated by circles, show the solidification times for the 2822-2732 range. It appears as if one could draw a curve for solidification time vs. superheat, almost independent of the range of solidification.

#### V. Comparison Between Sphere and Slab

It has been suggested repeatedly that a correlation between solidification rates of various shapes should be possible based on some characteristic figure including volume and surface area. With data now available for slabs (large; no end or corner effects) and spheres, a preliminary examination was made. A slab of thickness  $d$ , cooled on both sides and having the same surface as a sphere with a diameter  $d$ , has a volume three times as large as the sphere.

The solidification times for various degrees of superheat and a solidification range of 2700 F to 2600 F are tabulated in Table 6 for sphere and slab. The values hold for  $d=4.5$  in. For the slab the values had to be taken from a previous publication<sup>2</sup> and corrected by multiplying with  $(4.5/4)^2 = 1.265$ .

The first column contains the pouring temperature, the second the solidification time for the slab, the third that for the sphere, and the fourth the ratio of the two times. Obviously the ratio is not constant. A consideration of the heat transfer phenomenon involved confirms that this should be so.

TABLE 6—COMPARISON OF SOLIDIFICATION OF SLAB AND SPHERE

Pouring Temperature (F)	Solidification Time (sec) for Center of:		Ratio of Solidification Times Slab/Sphere
	Slab	Sphere	
2740	316*	215	1.47
2800	630	261	2.41
2820	695	270	2.57
2900	890	298	3.06
3010	1105	338	3.27

\* Estimate

Therefore it does not seem possible to reduce the cooling characteristics of the large slab and the sphere to one simple expression.

## APPENDIX

### Modification of Electric Analogy Method for Representation of Spheres

In setting up the analogous electric circuit for a heat flow problem the bodies (casting and mold) are thought of as divided into sections or lumps. The thermal resistance of each section is represented by an electric resistor, the thermal capacitance by an electric condenser.

In the case of a slab, sections of even thickness result in equal size of resistors and condensers respectively in each section. In the case of cylinders and spheres, however, the resistance and capacitance values change from section to section. With equal thickness of the sections, the capacitance of each section decreases and the resistance of each section increases towards the center. In fact, in either case the innermost resistance, starting at the center, would be infinite. Jackson and co-workers<sup>7</sup> have shown a method to overcome the handicap for cylinders. An analysis of spheres<sup>8</sup> based on the method shown by Jackson and co-workers proves that a similar approximation is not feasible for spheres. Therefore only hollow spheres can be investigated on the HMFA. Theoretically the hole of the hollow sphere can be made as small as desired, and therefore the approximation to the solid sphere can be made as close as necessary.

Practical experimental considerations require, however, a finite inside diameter. In the present experiments the last measured point is at a distance of approximately ten per cent of the radius. Extrapolation of the curves to the center is feasible, and was used on setting up Table 6.

#### Acknowledgments

The author is indebted to Dr. H. A. Schwartz for his interest in and support of the work and to Mr. C. W. Briggs for making available the original test data.

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## DISCUSSION

Chairman: E. C. TROY, National Engineering Co., Philadelphia

Co-Chairman: H. F. TAYLOR, Massachusetts Institute of Technology, Cambridge, Mass.

### Discussion on "Thermal Conductivity of a Sand Mixture"

DR. PASCHKIS (*Written Discussion*): It is known that theoretically it is not permissible to determine the thermal conductivity of mixtures based on the rule of mixtures. Nevertheless the writer tried this approach using the figures presented in a paper "Thermal Conductivities of Three Sands" by the same authors published in *TRANSACTIONS, A.F.A.*, vol. 55, p. 62 (1947) and in the present report. It is interesting that at least for the two sands under investigation the law of mixtures holds very closely as can be seen from the following table which was computed from the present and previous reports.

Temp. °F	A	B	C	D
2250	5.3	1.28	6.58	6.5
2000	4.33	1.24	5.57	5.3
1750	3.48	1.12	4.60	4.8
1500	3.00	0.96	3.96	3.9
1250	2.52	0.92	3.44	3.2
1000	2.17	0.88	3.05	2.8
750	1.80	0.84	2.64	2.6

Column A contains for each temperature the conductivity values for the sand 20-30 taken from Table 6 of the previous paper. The original values were multiplied by 0.6, because in the present mixture 60 per cent of this sand is used.

Column B contains for each temperature the conductivity values for sand No. 7 taken from Table 8 of the previous paper. The original values were multiplied by 0.4, because 40 per cent of this sand is used.

Column C contains the sum of the figures of Columns A and B for each temperature.

Column D repeats the values of thermal conductivity for the mixture from Table 2 of the present paper.

By comparing Columns C and D it can be seen that the conductivities based on averages (law of mixtures) is very close to the observed conductivities. In fact, with one exception (1,000 F), the deviation of the calculated from the observed value is 5 per cent or less, and more than 50 per cent of the values deviate 2 per cent or less.

Of course the study of one mixture is not conclusive but it

might be worth while to look into this matter, because this method of calculation is simpler than that proposed in the paper and checks more closely with observed values.

### Discussion on "Studies on the Solidification of Aluminum Castings"

H. E. GRAVLIN:<sup>1</sup> Would Dr. Paschkis say there was no appreciable air gap between sand and metal in the solidification test of aluminum?

DR. PASCHKIS: That is right. We conclude that from an indirect check which we make in the following manner: We first run a test of temperature solidified thickness vs time assuming "no air gap" and then repeat the test, assuming existence of an air gap. We find out, empirically, that the former condition closely checks the values observed in bleeding tests. Thus we conclude that "no air gap exists." This condition prevailed also in sand-casting of steel.

MR. GRAVLIN: Was the No. 4 line (Fig. 3) based on an assumed air gap?

DR. PASCHKIS: That is right.

MR. GRAVLIN: Of approximately what dimension?

DR. PASCHKIS: The dimension of the air gap is not too important because the main heat transfer across the air gap is by radiation which does not depend on the width of the gap. But, in addition to the radiation, heat is transferred by conduction across the air gap. It seems that in connection with the air gap, the time when the air gap is formed is of importance. Probably it does not exist immediately after pouring, but starts a short time later.

### Discussion on "Studies of Solidification of White Iron Castings"

W. K. BOCK:<sup>2</sup> The bleeding tests were run on those castings which were reported, I believe, by Dr. Schwartz last year. The manipulation alone at the plant was quite a chore. We thoroughly agree with Dr. Paschkis that it is not simple to get good consistent data.

DR. PASCHKIS: It is quite obvious that if it is so difficult to make bleeding tests and that if we can use our heat and mass flow analyzer, which may be considered as a calculating machine, we can free the foundrymen from jobs that are unpleasant and difficult. The Heat Transfer Committee of A.F.S. considers it one of the purposes of its work to prove if it is possible to use the analyzer because if and when foundrymen really become convinced that this method is equivalent to long drawout, exact, and annoying bleeding tests, answering of solidification problems can be taken out of the shop.

F. G. SEFING:<sup>3</sup> If a body of metal is freezing and there is some frozen metal and remaining liquid, I gather from what you stated about white cast iron that the temperature of the frozen metal is at the solidus and that the metal in the center was still coming down to the liquidus. Is that correct?

DR. PASCHKIS: Let us consider temperature curves in Fig. 22.

MR. SEFING: The temperature of the surface is at the solidus. It is moving across the horizontal (solidus) whereas the temperature of the liquid center of the casting at the same time is

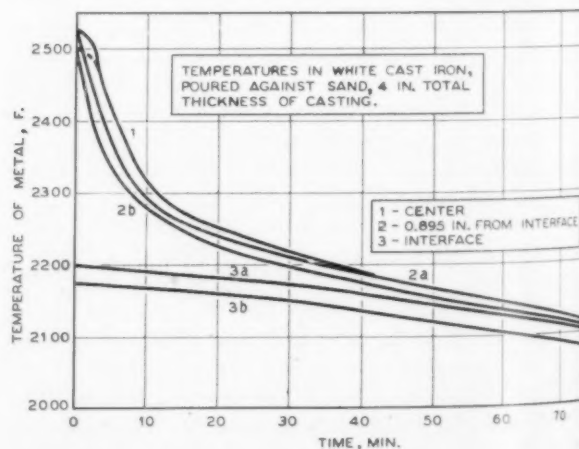


Fig. 22

<sup>1</sup> Ford Motor Co., Dearborn, Mich.

<sup>2</sup> National Malleable & Steel Castings Co., Cleveland

<sup>3</sup> International Nickel Co., Inc., New York

approaching the liquidus, according to your interpretation. In other words, from your interpretation, as half of the casting is frozen, the temperature of the liquid metal is still above the liquidus.

DR. PASCHKIS: How much above depends on the degree of superheat.

MR. SEFING: According to Dr. Hazlett, I think it was, once freezing has begun the temperature of the remaining liquid metal in the center of the casting is down at the liquidus. The outside, being solid, of course, is below the liquidus in the center and the difference in time between the freezing of the center and the outside is merely the time to remove the latent heat of fusion.

DR. PASCHKIS: I am not familiar with the work of Dr. Hazlett but I would not agree with his findings as explained by Mr. Seffing. They are contradictory to the measurements and I think they are also contradictory to what you would expect in the way of heat transfer.

MR. SEFING: Would you expect that a casting of copper with higher conductivity than iron might come nearer to the conditions that Dr. Hazlett was describing?

DR. PASCHKIS: Copper is a pure metal which solidifies at one given temperature and does not have a solidification range. Therefore its temperature curves would be of a character different from those shown in Fig. 22. The latter applies only to alloys solidifying over a range. In Fig. 2 of "Studies on the Solidification of Aluminum Castings" the temperatures first drop from the value of the pouring temperature, then remain constant for awhile and finally start dropping again. As mentioned previously we cannot detect any different temperature for the aluminum casting between surface and center. Copper, of course, being a pure metal, freezes following a similar temperature line. If Dr. Hazlett's work was on copper it would explain, I think, the conditions you mentioned.

MR. SEFING: As I recall, Dr. Hazlett's work was on pure zinc.

DR. PASCHKIS: As long as it is pure metal which solidifies at one temperature, the general character of the curve would be unchanged.

MR. SEFING: So the discrepancy in thought here depends upon having a constant freezing temperature, is that correct?

DR. PASCHKIS: Yes.

MR. SEFING: Then in freezing an iron casting we do not expect the remaining liquid in the center of the casting to be down to the temperature ready to freeze (liquidus)? It may be higher than the liquidus according to your findings?

DR. PASCHKIS: Yes, it should be higher.

R. W. TINDULA:<sup>4</sup> In a paper on "Solidification Characteristics of Gray Iron" by J. H. Schaum and J. E. Fifield (see p. 382), data was presented to show that when gray iron is about 20 per cent solidified, it will no longer bleed. At that stage solidified dendrites bridge over and hold the remaining liquid. These tests were run on cylindrical sand castings with thermocouples placed at the centers on three different analyses of gray iron. At the time liquid bleeding would stop the center was 20 per cent solidified.

DR. PASCHKIS: What do you mean by 20 per cent solidification?

MR. TINDULA: According to the constitutional diagram at the recorded temperature, only 20 per cent of the material should be solid.

#### Discussion on "Study of Solidification of Steel Spheres"

J. T. MACKENZIE:<sup>5</sup> I do not understand the difference between the liquidus and the solidus curves as to the amount of metal frozen. It is either frozen or it is not frozen. You have two curves with an amount of metal frozen against time. One of them you said is liquidus iron and the other one is solidus. If you can prove by bleeding tests, how can you tell whether it is liquidus or solidus?

DR. PASCHKIS: We can best discuss this by reference to Fig. 12. I showed on the curve a number of bleeding tests at 2800 F which lie directly on the liquidus line. You ask what does the solidus line mean if the casting is frozen already at the liquidus line. Our analyzer determines the time-temperature history of the casting. In determining the time-temperature history we do not have an actual casting before us. We know the size of the casting, its properties, and the pouring temperature. Then we

determine the time required by different parts of the casting to reach various temperatures. The temperature values with which we make this determination are arbitrary. Obviously, we select as one value the liquidus temperature, as another the solidus temperature. We might take a third curve at 2500 F and another at 2400 F. These lines indicate how long it takes any given thickness to reach the temperature indicated by the line.

CHAIRMAN TROY: Would pure metal be one line?

DR. PASCHKIS: No. Also for pure metals we can take such curves at any arbitrary temperature value. The difference is only that for alloys, we write on one curve "liquidus", on the other "solidus" and for pure metal we write on one curve "solidification temperature." Now to the decision, which curve to use in comparing with bleeding tests. When we first started with bleeding tests, we expected that the bleeding tests would coincide with the time-space curve for the solidus temperature. We found it is not so and we have found consistently that the bleeding test checks with the liquidus line. The solidus line is included only for completeness of record, but has no physical significance for our comparison.

CHAIRMAN TROY: We will not be too interested in the solidus line in that case. Are you satisfied, Dr. MacKenzie?

DR. MACKENZIE: I am on that point. It looks to me like our constitutional diagrams are wrong if I understood Dr. Paschkis correctly, because when you freeze a 2 per cent carbon iron for example, my understanding always has been that the temperature of the whole mass drops at the same time, but at the center, the melting point drops and consequently you first freeze out austenite with a melting point of 1420 C, then you continue to freeze that particular carbon content for the particular temperature until you get down to the eutectic at 1140 C. We made many freezing tests on hypo-eutectic irons that were as much as could be at 1250 C, but you could still work a thermocouple in them until you got right down to 1130 C. I do not know what size you could pour from, but I am sure we could not shut down a centrifugal casting at 1150 C and have it all stick on the wall.

CO-CHAIRMAN TAYLOR: There are no curves for gray iron.

DR. PASCHKIS: I should like for Dr. MacKenzie to explain his question more in detail.

DR. MACKENZIE: Let us consider the iron carbon constitutional diagram where temperature is plotted against carbon content. See p. 77, CAST METALS HANDBOOK, 3rd ed. At 1530 C pure iron freezes. Now the liquidus temperature drops to 1130 C when the carbon content is 4.3 per cent in pure iron-carbon alloys and then rises rapidly, whereas the solidus line drops to 1.7 per cent carbon and then goes straight across, so that at any given temperature you have a metal of certain carbon content freezing out and a liquid. When you got down into white irons you would be in the range of 2 per cent carbon. So your metal with this carbon content begins to freeze at 1420 C, continues to freeze with higher and higher carbon at lower and lower temperatures until we have 1.7 per cent carbon freezing at 1130 C and the remaining liquid would freeze at 4.3 per cent carbon and 1130 C.

CO-CHAIRMAN TAYLOR: You will not have very much.

CHAIRMAN TROY: That is the factor that I believe was brought up here in the 20 per cent. That is what you were trying to get across, that your material stops flowing.

MR. TINDULA: The frozen dendrites act like a sponge to hold the liquid by capillary action. If you could centrifuge the mixture, liquid probably would still come out, but without external force it will no longer flow when about 20 per cent solid has formed.

CHAIRMAN TROY: We are overlooking the fact that the flow of heat from the surface of the casting outward into the sand is so much faster than the flow of heat from the center to the surface in the metal body itself that we do not have equilibrium conditions at all. So our equilibrium diagram will not hold.

DR. MACKENZIE: That is what we are working on, to find out how much that effect is.

CHAIRMAN TROY: That is exactly what we are trying to measure.

DR. MACKENZIE: We did our work with the Bureau of Standards on the freezing point of these different cast irons. We poured a 6-in. cylinder, 1 ft. deep. We put a thermocouple in and kept it stirred.

CHAIRMAN TROY: What did you pour it in?

<sup>4</sup> Naval Research Laboratory, Washington, D. C.

<sup>5</sup> American Cast Iron Pipe Co., Birmingham



DR. MACKENZIE: We poured it into a sand mold. You get a definite freezing point and it makes no difference how much you stir it so long as you do keep it stirred.

DR. PASCHKIS: Of course, stirring it means that you make the temperature uniform artificially, but that is not what is happening in regular casting.

DR. MACKENZIE: But as long as the metal is completely liquid it keeps stirring itself. The cold metal keeps going to the bottom and the hot metal rises to the top.

DR. PASCHKIS: In pouring ingots it was found that for killed steel which has relatively little convection we could duplicate the solidification of an ingot reasonably well. As soon as you get to rimmed steel, with an appreciable amount of convection the electric analogy does not hold as well any more. Was your cast steel rimmed steel or killed steel?

DR. MACKENZIE: It was killed steel.

DR. PASCHKIS: It has been proven repeatedly by experience that the electric analogy holds for killed steel and neglecting natural convection, your casting could be represented on the analyzer.

J. B. CAINE:<sup>6</sup> Will Dr. Paschkis elaborate on why he thinks the Chvorinov theory regarding the relation between surface area and volume and freezing time is in error?

DR. PASCHKIS: As yet we have only the proof that it does not work. The ratio of solidification time of a given sphere to that of a given slab varies with pouring temperature. That is contrary to Chvorinov's statement because if the surface volume ratio does not change with the pouring temperature and, therefore, the ratio of solidification times should be independent of the pouring temperature.

MR. CAINE: I do not understand that either.

DR. PASCHKIS: There is no reason why Chvorinov should be right. I am not aware that Chvorinov has in any way proven his contention except by making relatively few castings and then establishing his rule saying "Well, it looks that way empirically." If his rule is only an assumption made on the basis of a limited number of experiments and then by fairly systematic work it is proven that the few experiments were in error then it is difficult to answer the question why his assumption is incorrect.

MR. CAINE: I am sure that Chvorinov's original work used Briggs and Gezelius' results. You are using them now to disprove his work.

DR. PASCHKIS: I do not know but assume that Chvorinov had only the published paper by Briggs and Gezelius in hand. Now these authors average their results obtained with different pouring temperatures and draw all conclusions from averaged values.

MR. CAINE: I will have to go over Chvorinov's original work, because I checked some of the surface area-volume relations of their spheres against their solidification times.

DR. PASCHKIS: Chvorinov's?

MR. CAINE: No, Briggs and Gezelius' spheres that Chvorinov used and I checked the point on the curve. Chvorinov's curve shows individual points. I think his points check against the actual data from the American publication of Briggs and Gezelius.

DR. PASCHKIS: The American publication uses averages in all its conclusions. I do not know if it is worthwhile but we could look at actual values, for example, at 2740 F and 2820 F.

CHAIRMAN TROY: Briggs and Gezelius' paper presented their averaged findings. The paper by Chvorinov used these averaged results. Dr. Paschkis got original data from Briggs and Gezelius and took from the original data the specific values at different temperatures.

MR. CAINE: Then I would qualify my question as follows: At a given pouring temperature would Chvorinov's theory regarding the relation of surface area and volume to freezing time regardless of shape apply?

DR. PASCHKIS: From the present paper I can only say I do not know. Experiments to be reported next year show that it does not check.

DR. MACKENZIE: The effect of superheat is just to heat the mold, is it not? A well known formula for fluidity of a molten metal is: temperature of the metal minus the freezing point over the freezing point minus the mold temperature, so that the superheat in this case goes into raising the mold temperature. After that then you start all over again. A fluidity test is nothing more than the determination of freezing rate. If you heat

your mold to the freezing point of the metal it will never quit running.

CHAIRMAN TROY: That is true if you transfer no heat from the metal.

DR. MACKENZIE: No, if you heat the mold at that temperature.

CHAIRMAN TROY: In other words, you are establishing the condition under equilibrium conditions.

DR. MACKENZIE: Yes.

CHAIRMAN TROY: We are trying to deal with something that is not quite an equilibrium condition and we do not know which way the factors are going.

DR. MACKENZIE: I am simply talking about the superheat. It is doing nothing but heating the mold. Then your freezing curves take over and your heat transfer problem comes in.

CHAIRMAN TROY: If you poured metal into a body that maintained a temperature equal to the solidifying temperature of the metal you are pouring, you would be right. You would drop only the superheat, but that is not the condition that we are trying to deal with. We are dealing with a mold continually extracting heat.

DR. MACKENZIE: That will not explain the difference between 2800 and 2700 F. That is just heating up the mold that extra 100 degrees.

DR. PASCHKIS: You say the difference between 2700 and 2800 F goes into the mold. Actually the mold-metal interface reaches a temperature between the initial mold temperature and the pouring temperature of the steel. This value remains constant for some time, unless, as in pouring against a metal surface or a lubricant an additional heat transfer resistance develops. Heat flows from the metal to the mold under the influence of the temperature difference between the metal (initially pouring temperature, slowly decreasing) and the interface. Therefore there is no justification in saying the heat content of the range 2800 and 2700 F goes into the mold—the entire heat content above this interface temperature flows to the mold. You mentioned fluidity. Do you mean fluidity expressed in inches?

DR. MACKENZIE: Anything you want to use. Fluidity is the inverse of viscosity.

DR. PASCHKIS: Yes, but in our test we are not dealing with viscosity.

DR. MACKENZIE: No, you are not. I am citing this as a parallel case.

CHAIRMAN TROY: We have reached that point where the percentage of solid is so great that it has the physical strength to endure when we dump it over. I feel that the last crystals forming were probably washed out with the liquid. In bleeding we start measuring at that point where the solids are attached strongly enough to the mold wall to endure disturbance. So it is not truly liquidus or truly solidus that we are measuring. Apparently we are closely approaching the liquidus temperatures if our figures are correct. If you cross section the bleed mold wall and etch you will find that the percentage of voids is rather small. At least such was true in steel.

DR. MACKENZIE: Try that on the other side of the iron-carbon diagram. Take a hyper-eutectic cast iron with 6 or 7 per cent carbon and make your studies on that side. In that case your solid constituent will not be getting in the way.

On the point about mold temperature, anyone who pours from a ladle knows that the freezing rates that are being discussed, do not obtain in a hot ladle at all. You can hold a ladle of steel for 6 to 8 hr if you get it hot enough to start with. Blast furnace metal is commonly held for 20 to 24 hr.

CO-CHAIRMAN TAYLOR: I believe our thinking is confused somewhat by not defining our endpoints. When Dr. Paschkis attempted to get confirmation of his results in gray iron it was difficult and I think even on white iron he adds a statement that there are certain physical observations that make it impossible to check the results accurately. When he got to other materials, like steel for example, it was easier to measure actual values in making these comparisons. The reason for this is that when you bleed steel castings you get a well defined solidified layer. The dendrites are not long. The directional growth of dendrites is relatively short in steel. However, as Dr. MacKenzie has pointed out, if you try to bleed castings to the right of your eutectic, say with 3.5 to 4 per cent carbon in the iron, then you find that the dendrites actually grow out as much as 2 or 3 in. into the liquid. If you bleed such a casting the liquid in be-

<sup>6</sup>Sawbrook Steel Castings Co., Lockland, Cincinnati

tween the dendrites flows out for a distance of 2 in., leaving these solid dendrites sticking right out into the liquid.

Mr. TROY feels that these dendrites are washed off in the bleeding experiment. Interestingly enough, I do not believe they are because you can actually prick your finger on them. These dendrites are as sharp as a needle, and furthermore they do not lean in the direction of the bled metal. They stand right out as though they were as stiff as concrete. They are a definite solid. One should think of a mushy metal with reservation. Part of it is solid and part of it is more liquid than it was before. This is why the iron we are talking about is difficult to measure in bleeding experiments. The measurements that Dr. Paschkis made on gray iron or white iron are not measurable with accuracy except by careful quantitative means. So that what we are pointing to in the main is confirmation of Dr. Paschkis' results on the metals which we can measure with accuracy because of the low directional growth of the dendrites and the uniform wall from which the metal is bled out. Then we accept the accuracy of his data to indicate an average measurement for the metals where we get a less accurate experimental measurement of interface. Does that make sense, Dr. Paschkis?

DR. PASCHKIS: It certainly makes sense to me. It would be highly desirable to get a temperature curve for steel similarly to the one shown for aluminum. Put a thermocouple in the

center of the casting and compare the temperature curve in solidifying steel with our prediction.

G. J. GROTT:<sup>7</sup> I should like to have clarified what Prof. Taylor said concerning dendrites. As I understand it, dendrites are a matter of segregation. The more carbon you have in the iron the greater the segregation would be and the longer the dendrites would be. Was the range covered by Dr. Paschkis' experiment that where the dendrites were very short and did not trap anything?

CO-CHAIRMAN TAYLOR: No. That is the one that we can check. For the others we have to assume that his methods are correct and that they measure a somewhat different quantity.

MR. GROTT: If the dendrites were very short then the metal that went out would be anything right up to that liquidus line.

CO-CHAIRMAN TAYLOR: I think that we have to take Dr. Paschkis' measurements in gray iron or in white irons rather reservedly perhaps at this time. The endpoints are not clearly defined in our own minds, are they?

DR. MACKENZIE: I should like to see you try it in hyper-eutectic gray iron. You would not have any dendrites at all. It will freeze out as graphite.

CHAIRMAN TROY: Do you mean to make freezing tests?

DR. MACKENZIE: No, you could not make freeze tests on hyper-eutectic iron. You could probably analyze it though.

<sup>7</sup> Massachusetts Institute of Technology, Cambridge, Mass.

# SOLIDIFICATION CHARACTERISTICS OF GRAY CAST IRON\*

By

J. E. Fifield † and J. H. Schaum ‡

## ABSTRACT

*Solidification studies herein described were conducted by (1) bleeding partially solidified gray iron castings and (2) obtaining cooling curves at various points within castings which were identical to the bled castings and which were poured simultaneously but allowed to solidify completely.*

## Introduction

MANY OF THE PROPERTIES of castings are affected by the manner in which solidification takes place. Data on solidification characteristics of metals are therefore of great interest and value to foundrymen. Only a few studies have, however, been made to obtain such data. One was by Briggs and Gezelius<sup>1</sup> on steel, which consisted of pouring castings, allowing solidification to proceed for varying lengths of time, dumping the remaining liquid, and making observations on the solidified skin. They concluded that:

- a. As the surface area is increased for a given volume of casting, the rate of solidification is increased.
- b. Variations in the common refractory mold materials cause little change in the velocity of solidification.
- c. The degree of superheat modifies the rate of solidification; with higher superheat the rate of solidification is lowered.
- d. If the temperature of the mold is higher than normal, the velocity of solidification is lower than normal.

The work reported here covers similar experiments on cast iron.

## Experimental Work

The best way to obtain practical information regarding characteristics of solidification is to hold all conditions, except one, as constant as possible and observe the process of solidification as a function of time. The solidification studies described in this report were con-

ducted by (1) bleeding \* partially solidified castings and (2) obtaining cooling curves at various points within castings which were identical to the bled castings and which were poured simultaneously but allowed to solidify completely.

The heats were melted in a high-frequency, lift-coil induction furnace in clay-graphite crucibles. The raw materials were Armco iron, graphite, and ferro-alloy. Each heat was inoculated 30 sec before tap with 0.5 per cent silicon as 50 per cent ferro-silicon.

The pyrometric studies were made with a Brown high-speed temperature recorder. Measurements were made by inserting thermocouples into the sides of the mold cavity at the center of the vertical axis. Temperature measurements near the mold-metal interface were made by having the protection tube make a short chord across the cylinder with the hot junction against the far wall of the mold.

## Technique Used

The first experiments were made on a 2 x 6 x 8-in. rectangular block with a riser of 3 x 7 x 2-in. placed on top of one of the 2 x 6-in. faces. This block was placed with the head up, top poured through the riser, and bled through the head by inverting the casting. It was also cast in the inverted position with the heavy section in the drag, open through the cope, gated at the parting line, and bottom bled at the parting line by lifting the cope from the drag. The latter method was found to be more satisfactory for the casting received less jarring action than when it was inverted. Because cast iron is extremely weak at the solidifying temperatures, the casting tears in half at the junction of the two sections when the cope is lifted from the drag and the remaining liquid drains easily from the cope. Forty-eight block castings were poured, but it was found that these blocks were not suitable test castings because the side walls sagged together and welded so that the results were obscured.

The block type casting was therefore abandoned in favor of cylindrical castings. A group of four solid-

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\* Bleeding is the term used in this report to indicate the action of removing the remaining liquid metal from a partially solidified casting by overturning it or by other means. Some foundrymen use the term, "slush-casting," to describe this procedure.



cylinders, 1½, 2, 2½, and 3 in. in diameter, all 8 in. high, with a heavy section 1 in. larger in diameter than the cylinder and 1½ in. deep attached to the bottom of each cylinder, were bottom-poured through horn gates. They were each bottom bled, by lifting the cope from the drag. About 350 of these castings were poured from 77 heats of metal. A photograph of one of these castings is shown in Fig. 1. The mold was made with a common drag but with a separate cope for each set of cylinders. One set was bled and pyrometric data were obtained from the undisturbed set.

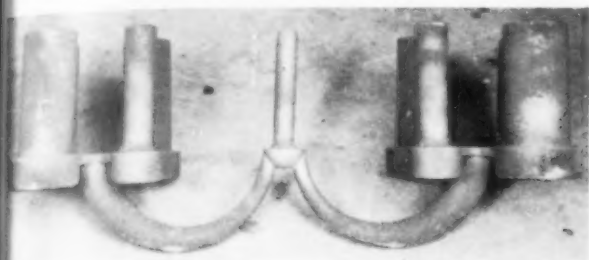


Fig. 1—Completely solidified test casting of the type used for bleeding experiments.

The patterns were molded in fine synthetic sand, the composition and properties of which were:

Washed and dried silica sand (135 gfn)	96 per cent
Western Bentonite	2 per cent
Dextrine	1½ per cent
Corn flour binder	½ per cent
Moisture	5-6 per cent of dry weight
Green Permeability	38-48 A.F.A. units
Green Strength	5½-6 psi
Hardness	70-78 A.F.A. units
Flowability	75-83 per cent
Dry Permeability	52-65 A.F.A. units
Dry Strength	180-230 psi

All molds were air-dried for at least 24 hr before they were used. However, from preliminary experiments, it was determined that no marked difference was caused by 4 to 96 hr variation in the drying time.

In each case the pouring temperature was measured by a black-body, immersion-type, optical pyrometer. The maximum pouring time was 14 sec and the average time, 8 sec. The elapsed time before bleeding was measured from the time the mold was half filled until bleeding was started. Observations of wall thickness were made by sectioning the cylinders transversely at the center of the long axis and photographing them.

Sets of cylinders were poured from three different compositions of the following approximate chemistry:

- Low carbon: 2.5 C, 2.0 Si, 0.6 Mn, 0.01 P, 0.03 S.
- Medium carbon: 3.0 C, 2.0 Si, 0.6 Mn, 0.01 P, 0.03 S.
- High carbon: 3.5 C, 1.5 Si, 0.6 Mn, 0.01 P, 0.03 S.

A constant pouring temperature (2600 F) was used for all compositions and all cylinders were bled the same time after pouring, namely 2 min. Figures 2, 3, and 4 show the results obtained. Only one bar in each set was hollow and therefore suitable for skin thickness determination. Smaller bars were solid and larger ones just yielded thin, irregular-shaped shells at the top of the mold cavity.

Although the results indicate a variation in solidification rate with chemistry, it could not be concluded that this effect was due entirely to chemistry since there was a simultaneous variation of superheat. An attempt was therefore made to differentiate between chemistry and pouring temperature by making one run in which the chemistry was held constant and the pouring temperature varied, and another in which the pouring temperature was 300 F above the solidification temperature and the chemistry varied. The results of the first run are given in Fig. 5 showing the results of pouring cylinders at six different temperatures from the same ladle of metal and bleeding 2 min after pouring. (The castings which did not bleed were omitted from the picture). The chemical analysis of this iron was 2.88 per cent C, 2.13 per cent Si, 0.51 per cent Mn, 0.078 per cent S, 0.071 per cent P. Since composition and bleeding time were constant, a comparison of the castings with the same diameters in each horizontal row indicates that the rate of solidification decreases with an increase in pouring temperature. As would be expected, the rate of increases as the size of the casting decreases. The second run consisted of duplicate castings of two different compositions but poured at 300 F above their respective liquidi, namely 2590 F and 2660 F and bled 3½ min later. As shown in Fig. 6 the higher



Fig. 2—Solidified sections of cast iron (3.59 C, 1.68 Si) cylinders bled 2 min after pouring at 2600 F.

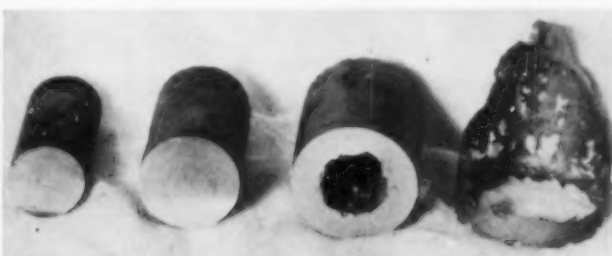


Fig. 3—Solidified sections of cast iron (3.05 C, 1.95 Si) cylinders bled 2 min after pouring at 2600 F.



Fig. 4—Solidified sections of cast iron (2.57 C, 1.87 Si) cylinders bled 2 min after pouring at 2600 F.

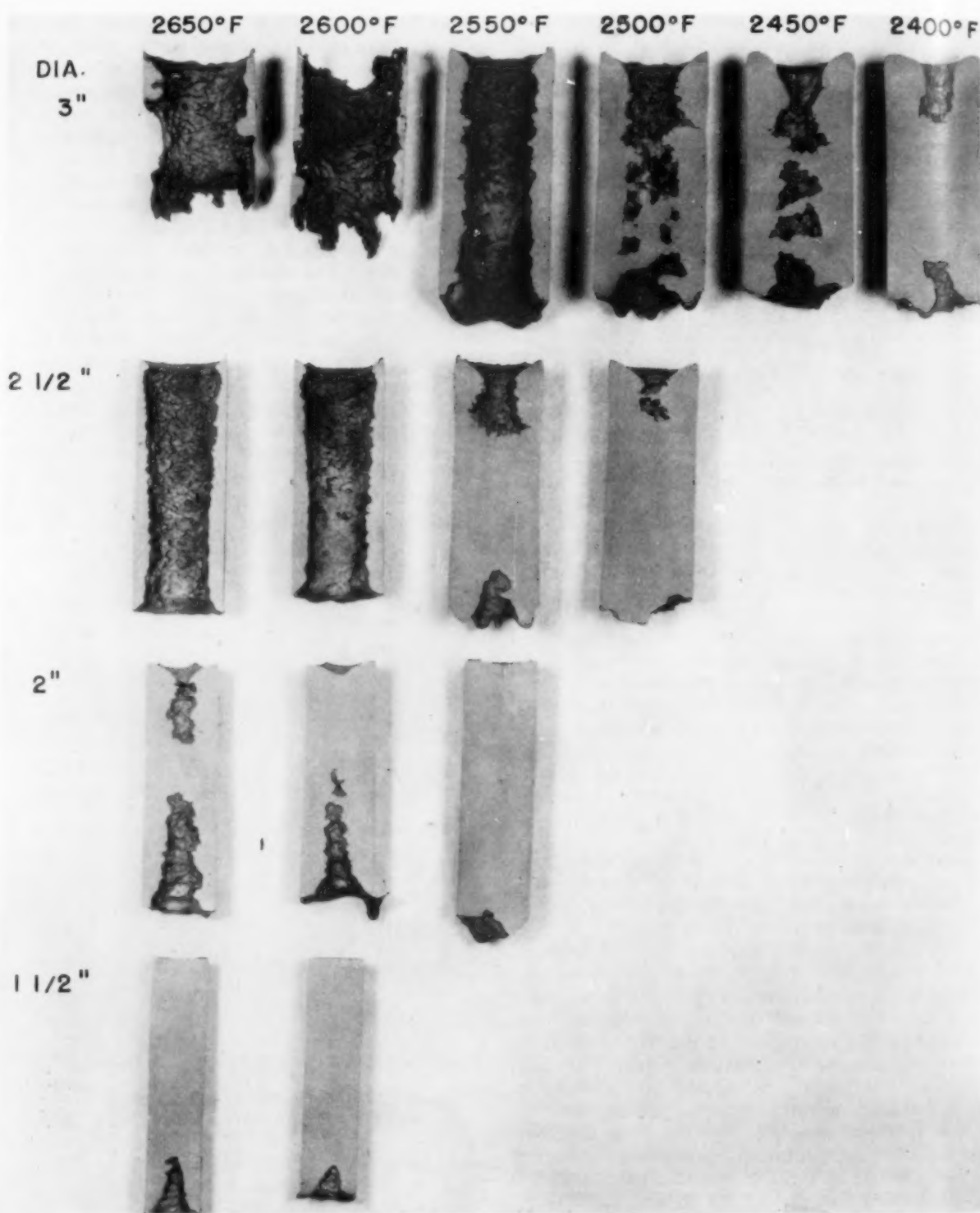


Fig. 5—Sectioned cylinders which were poured from the same ladle of metal at the indicated temperatures and bled 2 min later.

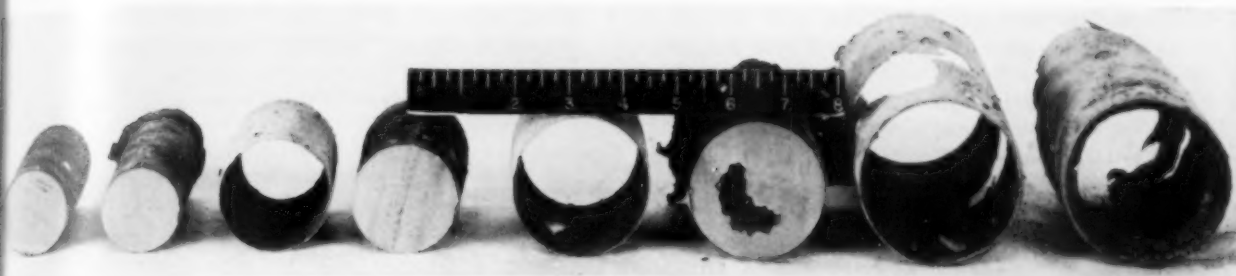


Fig. 6—Effect of composition on bleeding characteristics.

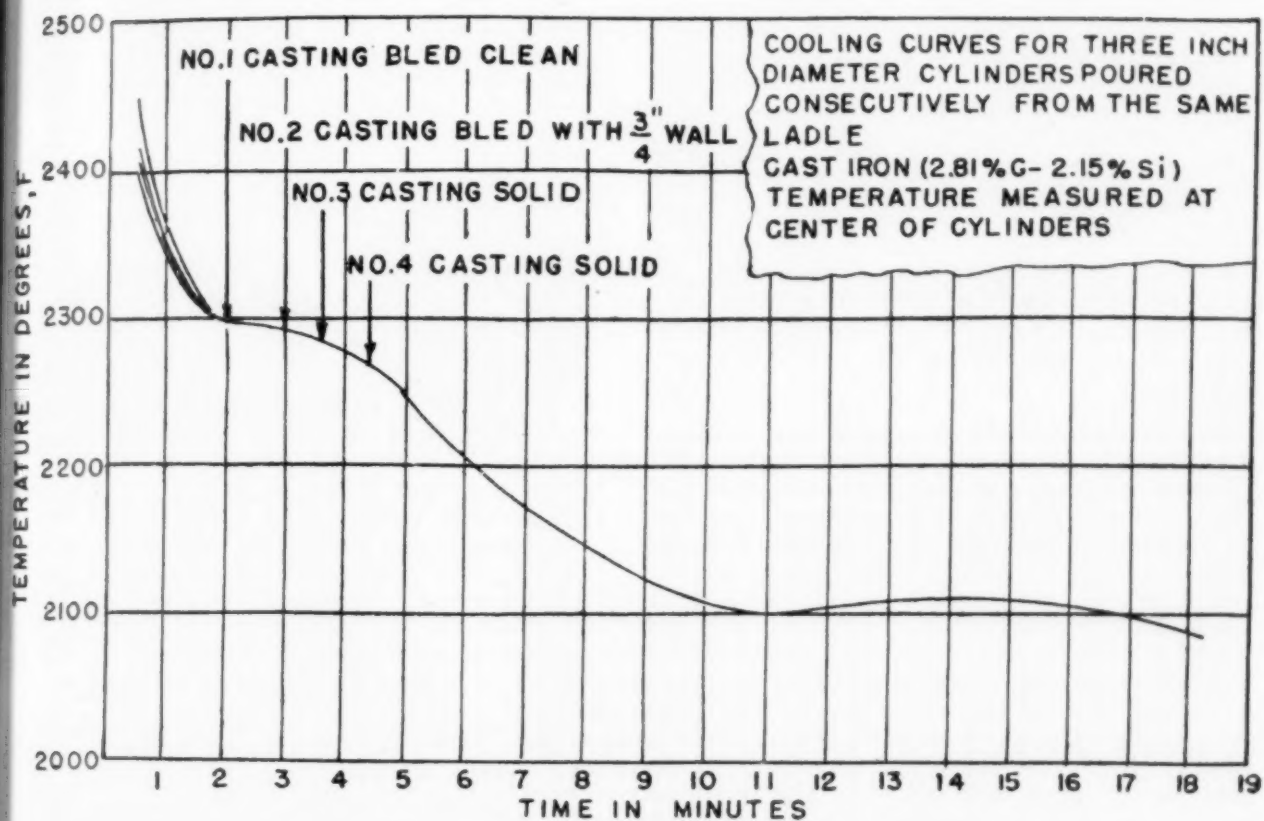


Fig. 7—Series of cooling curves from data obtained at centers of four 3-in. diameter cast iron bars.



Fig. 8—Solidified remains of cast iron (2.81 C, 2.15 Si) cylinder bled 2 min after pouring at 2600 F. Left is drag. Right is ring at top of cope.



Fig. 9—Solidified remains of cast iron (2.81 C, 2.15 Si) cylinder bled 3 min after pouring at 2600 F. Left is drag. Right is hollow cylinder of cope.



carbon iron bled to a greater extent than the lower carbon iron, although each was poured with the same increment of temperature above its liquidus. The left of each pair of castings (of the same diameter) analyzed 3.89 per cent C, 1.87 per cent Si, 0.61 per cent Mn, 0.026 per cent S, 0.005 per cent P and the right of each pair analyzed 3.05 per cent C, 1.87 per cent Si, 0.61 per cent Mn, 0.026 per cent S, and 0.005 per cent P. This experiment shows that chemistry also affects the rate of solidification. While it might be argued that this result could be due to the change in cooling rate caused by the difference in pouring temperatures, it is felt that this factor is too small to produce such a great difference in volume solidification.

A correlation of the process of solidification with temperature was effected by comparing cooling curves with thickness of skin at various times after pouring. Figure 7 is a series of cooling curves drawn from temperature data obtained at the centers of four 3-in. diameter cast iron bars poured one after another from the same ladle. The curve is marked when a cylinder was bled. Figures 8, 9, 10 and 11 are photographs of the bled castings. Figure 8 shows the remains of the casting bled at 2298 F, 2 min after pouring; there was just a small ring at the top of the bar, no sidewalls, only some evidence of solid metal in the bleedings. Figure 9 shows the casting which was bled at 2295 F, 3 min after pouring. This casting had a solid wall which was torn in bleeding with part dropping on the liquid bleedings. Figure 10 is a photograph of the casting which was bled at 2285 F, 3 min and 40 sec after pouring. This casting was almost solid for it showed a semi-crystalline fracture. When air-pressure was applied to the head of it in an attempt to force out all liquid

metal, a portion of the wall was blown to the floor and deformed slightly. The wall appeared to be a plastic mass barely able to support its own weight. Figure 11 shows that the last casting acted as a solid. It was

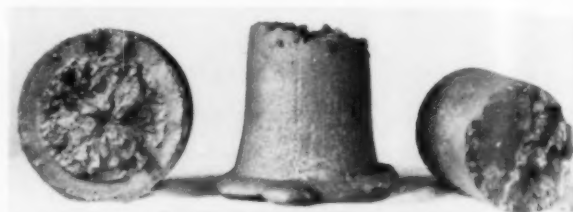


Fig. 10—Solidified remains of cast iron (2.81 C, 2.15 Si) cylinder bled 3 min 40 sec after pouring at 2600 F. Left is drag. Center is bottom half of cope after falling 4 in. to floor. Right is top half of cope.



Fig. 11—Solidified remains of cast iron (2.81 C, 2.15 Si) cylinder bled 4 min 21 sec after pouring at 2600 F. Left is drag. Right is cope. Note crystalline tear.

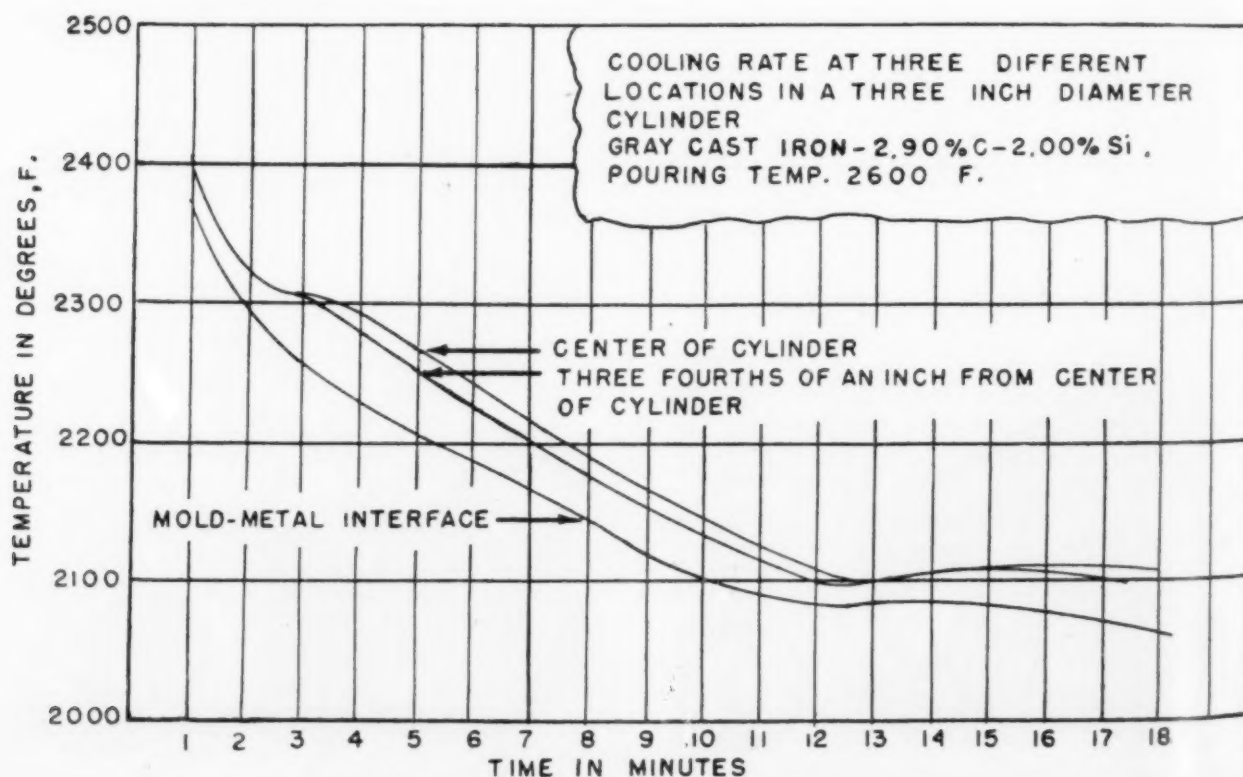


Fig. 12—Cooling curves made at center and outside of 3-in. diameter bar.

Fig. 13—  
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difficult to break the cope from the drag and the casting had a solid crystalline tear, even though it was torn at 2268 F, only 4 min and 21 sec after pouring, and 168 F above the solidus. This experiment was duplicated four times for verification of the results.

To assist in accounting for these results, cooling curves were made at the center and outside of a 3-in. diameter bar. These curves are given in Fig. 12, which show that the outside reaches the liquidus about 2 min after pouring and the center about 1 min later.

It is believed that the results shown in Fig. 7, 8, 9, 10, and 11 may be explained as follows. The casting shown in Fig. 8 was bled before enough dendrites had formed to give any strength to the wall and all the solid metal was swept out with the liquid as it drained from the mold. In Fig. 9 the dendrites had thickened enough to form a strong wall. The third casting, Fig. 10, had cooled far enough below the liquidus so that the dendrites were able to support approximately seven-eighths of the metal as liquid. (This proportion was calculated from the iron carbon diagram, using the lever arm principle). The last casting, shown in Fig. 11, acted as a solid even though the temperature near the surface was approximately 2175 F or 75 F above the solidus.



Fig. 13—Solidified section of cast iron (3.00 C, 1.72 Si) cylinder bled 3 min 19 sec after pouring at 2600 F. Note droplets of eutectic metal squeezed from wall.

That the walls of a bled specimen of cast iron consist of dendrites with liquid entrapped in them, is shown by Fig. 13 which is a photograph of the cross section of a bled cylinder showing small droplets of metal clinging to the inner wall. The formation of these droplets may be observed on the inner wall of the cylinder shortly after bleeding. First a bright spot appears and then gradually grows into a drop of metal which becomes large enough to break loose and fall to

the floor. As many as five or ten droplets of metal may exude from any one spot before the metal becomes cold enough to freeze at the point of exudation. These metal globules are commonly observed on the tops of risers as shown on the sectioned castings appearing in Fig. 14.

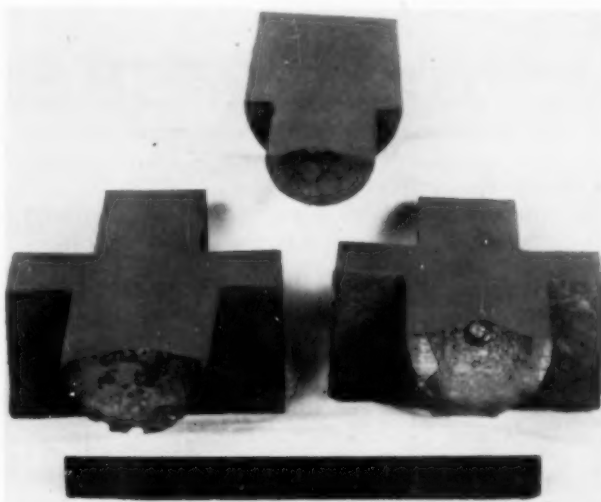


Fig. 14—Three castings with exudations on top of their risers.

Chemical compositions of five of these exudations from different heats indicated equivalent carbons (per cent carbon plus  $\frac{1}{3}$  of the per cent silicon) of 4.49, 4.47, 4.41, 4.12 and 4.20 per cent, while the equivalent carbon of the base metal in the castings ranged from 3.6 to 4.0 per cent. As a specific example, the parent metal of one casting had a composition of 3.04 per cent carbon and 2.18 per cent silicon but the exuded metal analyzed 3.40 per cent carbon and 2.37 per cent silicon. Since the exuded metal approaches eutectic composition, it has a low fusing temperature and high fluidity. It is believed that the metal was exuded by the pressure created from the increase in volume accompanying the formation of graphite and from thermal contraction of the solid portion of the casting.

Since all the experiments indicated that thin cast iron walls at the time of solidification have little strength, it was felt that this was the reason that blind risers do not function on iron castings when the casting is of greater height than the riser. For a blind riser to feed a casting higher than the riser, the pressure acting on the molten metal in the riser must exceed that acting to force the metal from the casting to the riser. Since the ferrostatic pressure is greater in the casting than in the riser, the riser will feed only if the difference in atmospheric pressure exceeds the ferrostatic pressure difference. This can be accomplished by reducing the pressure within the casting while keeping the metal in the riser at atmospheric pressure.

The reduced pressure within the casting which is produced as a result of the normal shrinkage on solidification will be maintained only if the wall of the casting is strong enough to resist collapse from atmospheric pressure. An experiment was carried out to determine if this collapse does occur. It consisted of pouring two 6 x 12 x 1-in. castings with the 12-in. dimension in the vertical position and with the riser attached to the

bottom. The risers for both castings had the same volume, but one was 16 in. high and open to the atmosphere, while the other was a blind riser only 5 in. high. The casting made with the tall riser was sound and true to shape, while the one made with the blind riser had a bad tuck or draw near the top indicating collapse of the wall.

### Conclusions

1. The rate of solidification of gray iron in a given section size varies inversely with the temperature above solidification temperature at which the metal is poured.
2. The rate of solidification increases with decrease in cross-sectional area.
3. The rate of solidification varies with chemistry; up to the eutectic composition, the higher the carbon, the lower the rate.
4. Cast iron acts essentially as a solid after its temperature falls slightly below the liquidus; the tightly interlaced dendrites are able to hold their shape and support considerable weight of liquid metal.
5. Blind risers do not function effectively on cast iron because of its weak skin strength.

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### DISCUSSION

**Chairman:** R. G. McELWEE, Vanadium Corp. of America, Detroit

**Co-Chairman:** E. C. JETER, Ford Motor Co., Dearborn, Mich. W. W. AUSTIN, JR.:<sup>1</sup> What type of graphite structure was observed in the synthetic iron used in this work?

**MR. FIFIELD:** All of the irons used in this work were inoculated. Every iron examined microscopically had random flake graphite.

**CO-CHAIRMAN JETER:** What influence does temperature have on this type of defect? What has the author found out about influence of pouring temperatures on shrinkage?

**MR. FIFIELD:** We feel that these experiments indicate that cast iron acts as a solid when it has cooled a short way below the liquidus temperature. It seems improbable that gray iron can be fed when the dendrites are so tightly interlaced that they will support seven-eighths of the metal weight as liquid. Prob-

ably risers feed only the shrinkage in gray cast iron which occurs in the liquid state and during the progress of initial solidification. That shrinkage which remains thereafter as microshrinkage may be compensated for by the formation of graphite, which forces the remaining eutectic metal into the interdendritic shrinkage areas. We have no definite conclusions on which we can base this statement. It is more a theory at the present time than a proven fact.

**V. A. CROSBY:**<sup>2</sup> The authors state in conclusion No. 5 that blind risers do not function effectively in cast iron because of cast iron's weak skin strength. I have been informed that some foundrymen are using blind risers quite effectively and I am wondering if the authors have thoroughly proven that blind risers are of little avail in practice?

**MR. FIFIELD:** I agree. I know foundrymen are using blind risers on cast iron and they claim that they are avoiding difficulties through their use. However, in a laboratory, in a controlled experiment, you cannot make blind risers act the same in cast iron as they do in steel. In steel, in a controlled experiment you can feed metal up a vertical height of approximately 30 to 50 in. If you do the same experiment and try to duplicate it in cast iron, inevitably you will have a tuck or a draw in the gray iron casting which has not been fed by the atmospheric blind riser. I cannot explain the difference between practical and laboratory conclusions.

**MR. SCHAU:** I think it depends somewhat on the definition of a blind riser. There are blind atmospheric pressure types of risers and there are blind risers. I have seen the latter type on castings where the riser is attached to a section down inside the mold and it does not have to feed uphill. It is feeding laterally or perhaps feeding a part of the casting below the location of this so-called blind riser. We really should qualify our reference to a blind riser as one functioning with the aid of atmospheric pressure where feeding is meant to be above its level. I am distinguishing here between the two types of blind risers. You can have the one type that feeds downward by the static pressure of the metal, or the other type in which it is a matter of atmospheric pressure forcing metal uphill or having it rise upward because of a vacuum being created.

**CO-CHAIRMAN JETER:** In reality both co-authors say the risers will not do much good on cast iron. Is that correct?

**MR. FIFIELD:** I do not want to say that risers are ineffective on cast iron, because I know very well you have to use them, but to a certain extent it seems hard to believe that the types of irons which were discussed in the paper are certainly not representative of all kinds of irons which are poured today. It seems hard to believe that risers for those irons do much more than compensate for the shrinkage which occurs in the liquid state and in the initial stages as it passes from the liquid to the solid state.

**MR. FIFIELD (Authors' Closure):** With regard to Mr. Jeter's last comment and the discussion concerning shrinkage, I should like to clarify my last statement by saying that risers must be used to feed gross shrinkage, such as piping or large voids, but these risers must act before the dendrites have completely traversed any intervening section. Risers are believed to be ineffective for feeding micro or interdendritic shrinkage.

<sup>1</sup> Southern Research Institute, Birmingham

<sup>2</sup> Climax Molybdenum Co., Detroit



# GATING SYSTEMS FOR METAL CASTINGS

By

W. H. Johnson \* and W. O. Baker \*

## ABSTRACT

This investigation was undertaken to observe the effects of gate design upon the flow and turbulence of liquid steel as a casting is poured. Gating systems and pouring techniques such as top pouring, whirl gates, riser gates, finger gates, horn gates, step gates and bottom side gates were studied. The results showed that many gating systems did not function as commonly supposed. The behavior of each system is described in the text.

To observe the discharge of molten steel into the mold cavities use was made of motion pictures. High speed photography at 1,000 frames per sec, was used in a study of flow of molten steel through sprues. The most interesting scenes of flow and turbulence produced by the various gating systems have been assembled into a 16 mm sound film.

FOUNDRY TECHNIQUES for "gating" molten metals into molds are generally developed by trial and error methods. The need for the study of the behavior of molten metal within gating systems is widely recognized. Most of the information on fluid flow which is available in the literature is concerned with steady flow conditions and little can be applied to the filling of a mold.

Studies of flow in models of gating systems have been reported by S. L. Fry<sup>1</sup> and by R. B. Fisher.<sup>2</sup> Fry recorded the behavior of the low melting nonferrous metals lead and aluminum by cine-radiography. He placed plate molds between an X-ray source and fluorescent screen and photographed the image with a 35-mm cine camera. Fisher poured Wood's metal at 250 C into molds faced with transparent lucite and photographed the flowing metal through the lucite with a 16-mm cine camera. His work dealt with the flow characteristics produced by various designs of gates and different rates of pouring.

In our experiments molten steel at 3000 F was poured by means of various gating systems into open mold cavities. From directly over the mold 16-mm kodachrome motion pictures at 64 frames per second recorded the flow and appearance of the metal. Studies were made of top pouring, whirl gates, riser gates, finger gates, horn gates, step gates and bottom side

gates. Several variations were used for each of these gating systems.

The sprues and ingates in all gating systems were 1 in. in diameter except where a deviation from this dimension is noted below. In the studies of top pouring, and of step and bottom side gates, a casting 5 x 5 x 18 in. high was used with a riser 7 x 7 in. Whirl gates, riser gates, finger gates, horn gates and other step gates were used on 6 x 6 x 12-in. casting without a riser. This mold was placed in a horizontal position except in the case of step gating when it was vertical.

Turbulence in the castings was revealed by the swirling and rolling of the molten metal in the mold.

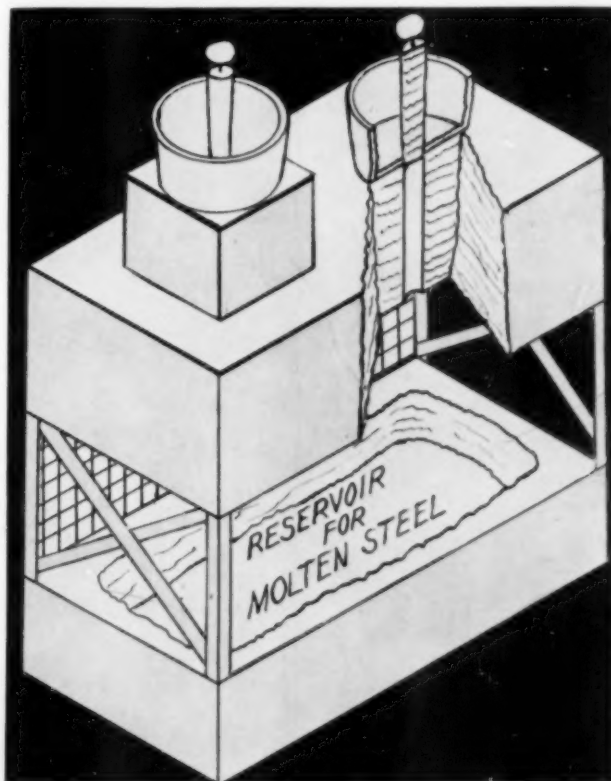


Fig. 1—Unattached sprues used for high-speed photographic studies of metal streams

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\*The authors are on the staff of the U. S. Naval Research Laboratory of the Office of Naval Research, Washington, D.C.

This was best observed and was of most interest in finger gating systems. When the combined area of the fingers entering the mold was twice that of the sprue area, a swirling turbulence occurred; when the areas were equal, a rolling turbulence was observed. Swirling turbulence was identified by a clockwise or counterclockwise movement or even both on the surface of the metal in the mold. The axis of rotation was vertical and perpendicular to the plane of the fingers. This form of turbulence was produced by unequal feeding through the several fingers. Rolling turbulence was characterized by a uniform flow on the surface of the metal from one side to the opposite. Its axis of rotation was parallel to the plane of the fingers and to the runner. This type of turbulence was associated with approximately equal feeding from each of the fingers.

In order to examine the effects of variations in reservoir level and in slag elimination with open sprue cups, some comparative runs were made with plugged sprue cups. The appearance of the stream emerging from a sprue without a runner was photographed. The mold

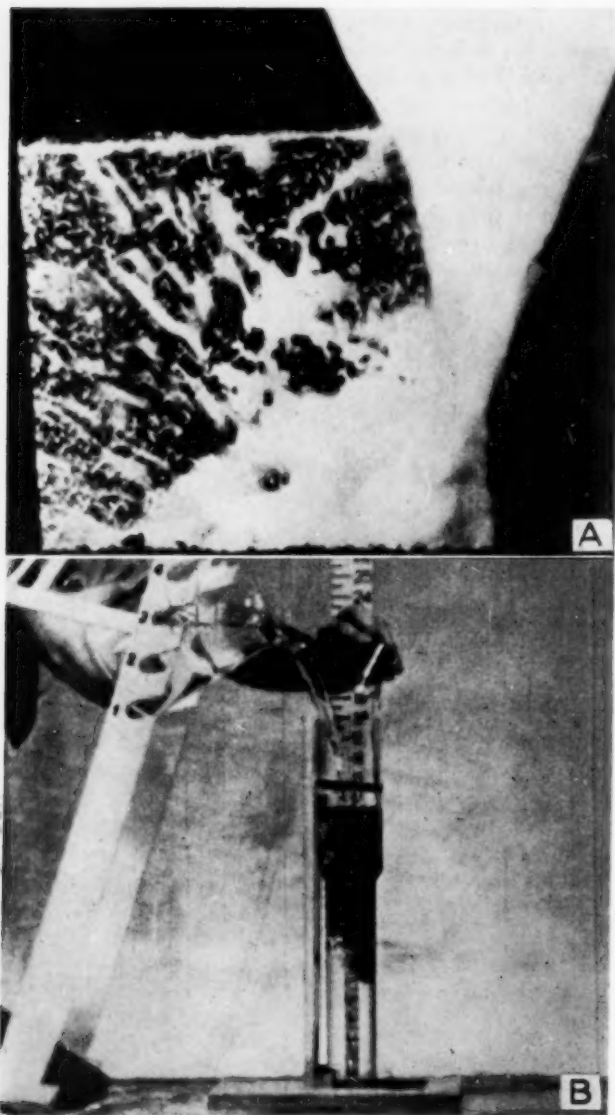


Fig. 2A—Top pouring and Fig. 2B—Top pouring, water and dye

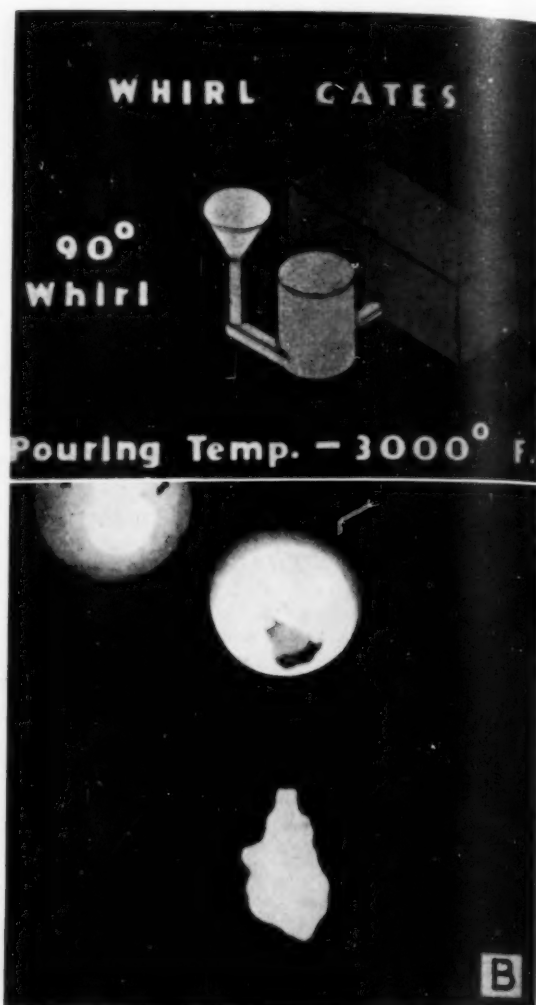


Fig. 3—Whirl gates

arrangement used for these studies is shown in Fig. 1. The stream from an open cup when photographed at 1000 frames per second and projected at 24 frames appeared very turbulent for some time after pouring started, but later became steadier and streamlined. The flow from a plugged sprue cup at high speed photography appeared to be streamlined during the entire pouring time, although there was frequent variation in cross-section of the stream. A plugged sprue cup appeared to increase the constancy of metal flow when compared with open cup behavior.

The results observed with the different methods of pouring and different types of gating systems are described below.

#### Top Pouring

Although direct pouring of molten metal into the top of a mold, as shown in Fig. 2A, is not commonly used in the foundry industry, steel ingots are cast by this method. Severe spattering occurred immediately and continued through the pouring cycle. Considerable turbulence was noted. A water-in-glass model, Fig. 2B, showed that even when the model was nearly full, approximately two-thirds of the liquid was agi-

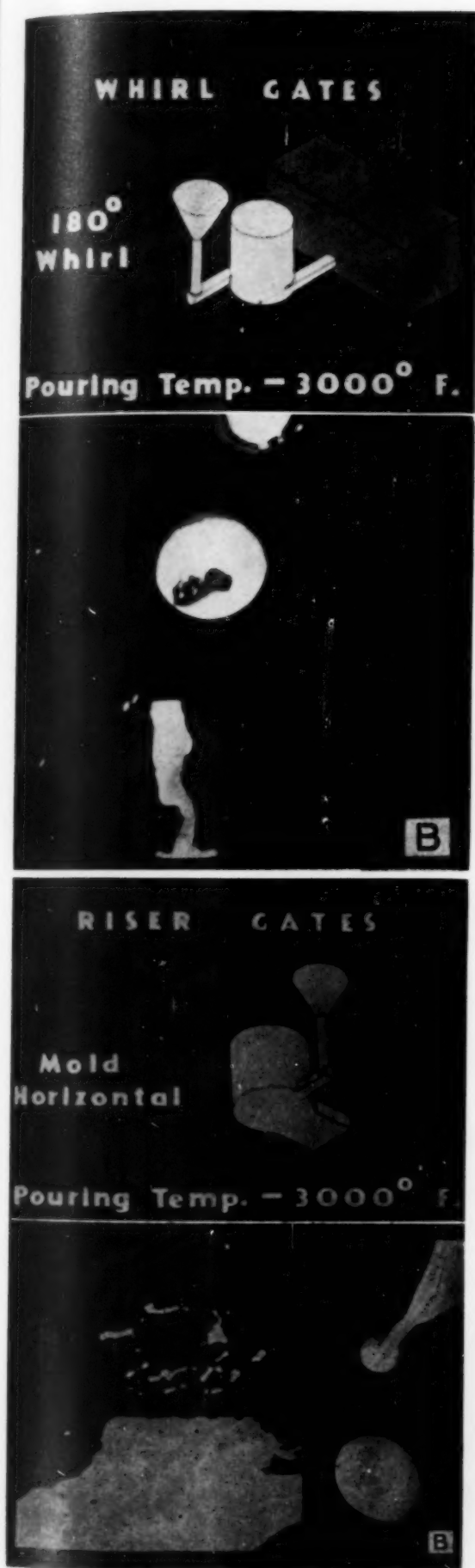


Fig. 4 (Left)—Whirl gates

tated by the entering stream. This phenomenon has also been observed by Northcott<sup>3</sup> and other investigators.

#### Whirl Gating

A whirl gate receives its name from the rotation which it imparts to the liquid metal in it. Separation of slag from metal by centrifugal force is an intended

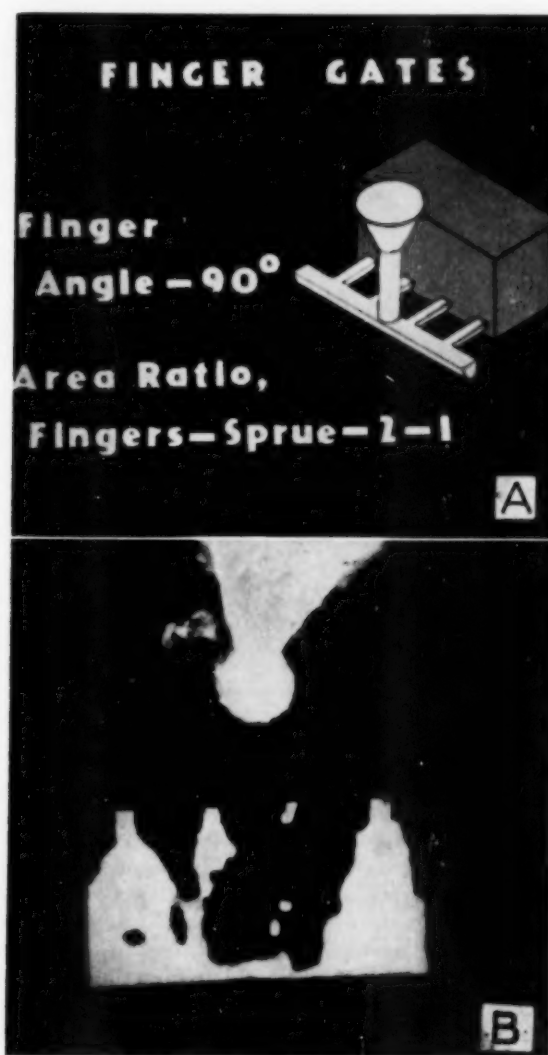


Fig. 6—Finger gates

objective. In the two variations of whirl gates studied, the runner and ingate always had the same cross-sectional area and were in the same plane, but the angle of the runner was varied with respect to that of the ingate.

In the first variation, Figs. 3A and 3B, the ingate was at an angle of 90 deg. to the runner. The metal ran

Fig. 5 (Left)—Riser gates



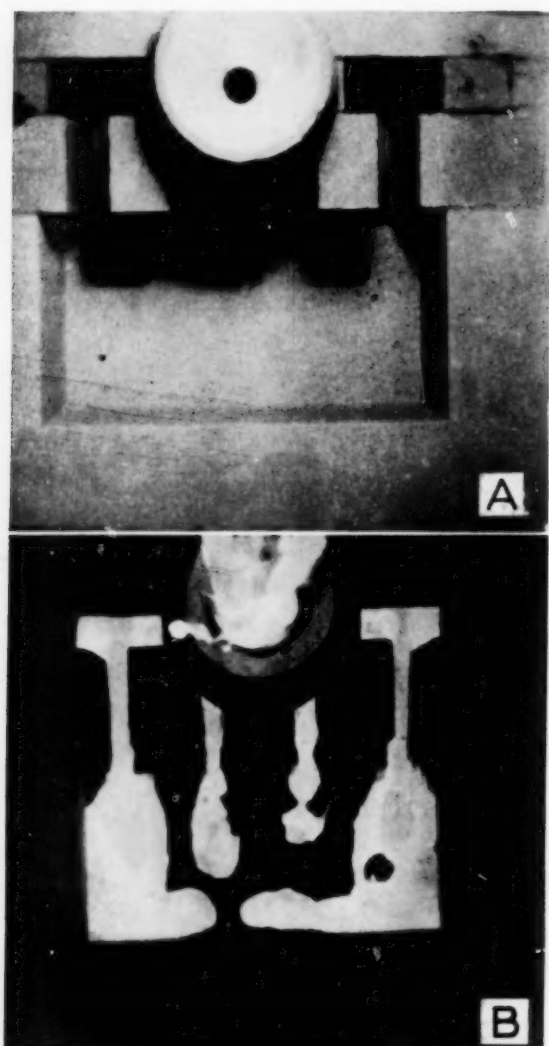


Fig. 7—Finger gates—open mold

through the gate into the casting before the bottom of the gate was completely covered, and so some slag entered the mold. Later, when the whirl became fully developed, slag appeared to be completely separated. In the second variation of this gate, Figs. 4A and 4B, the ingate was at an angle of 180 deg. to the runner. Again, metal entered the mold before the bottom of the gate was completely covered and in this stage some slag escaped into the mold. Swirls indicating turbulence were apparent on the molten steel in the mold with both designs of gate.

#### Riser Gating

A riser gate differs from a swirl gate only in having a larger ingate, allowing it to function as a riser when pouring stops. Figures 5A and 5B show a horizontal mold with a riser gate. This gate delivered metal at lower velocity than the whirl gates, but turbulence was still present in the casting. An interesting phenomena noticed with this system consisted of frequent changes in light intensity on the metal's surface during and

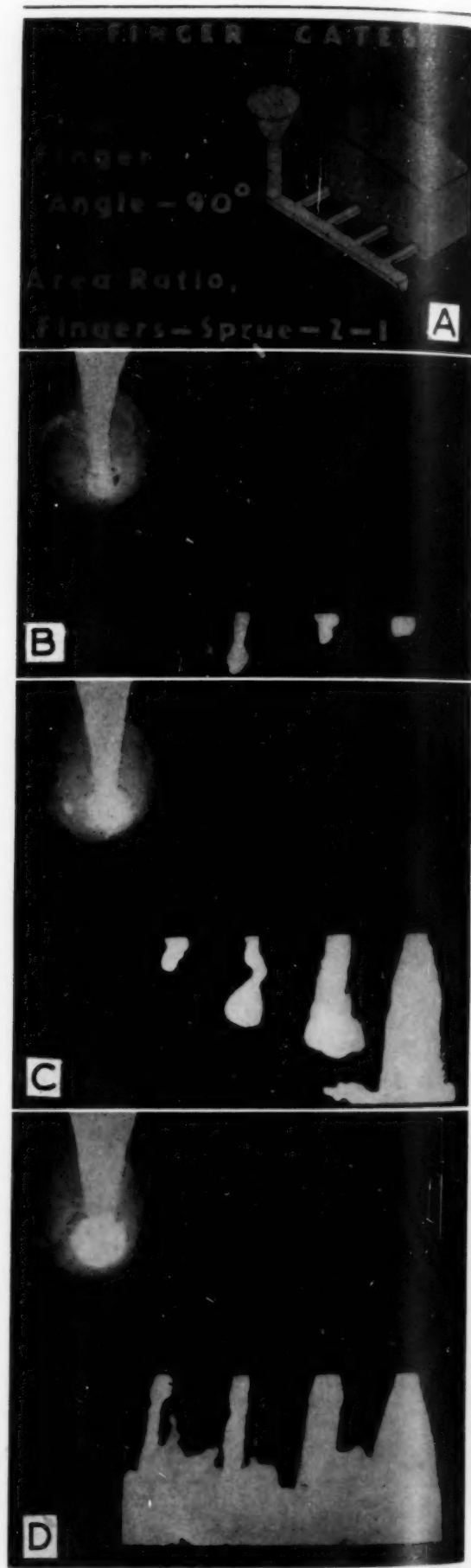


Fig. 8 (Right)—Finger gates

after pouring. The cause of these "flashes" is not known at present, but they will be referred to again in describing other gating systems.

### Finger Gating

Finger, or pencil gates, are employed to distribute metal into the mold in a horizontal plane over a large area. The number of fingers, their positions relative to the sprue and the angle entering the mold were factors studied. Arrangements of four finger gates of equal size in which the total finger area was twice that of the sprue area were studied first. Most of the molten metal, Figs. 6A and 6B, entered the casting through the fingers farthest from the sprue. This continued after the mold became full as shown by the opposing swirls on the surface of the steel.

For better observation of the movement of the steel in the fingers, the cope of the mold was removed, Figs. 7A and 7B. When pouring first started, the metal flowed past the nearest gate openings to the ends of the runner. After sufficient metal had entered to cause a build-up of metal at each end of the runner, the outer fingers began to feed. Flow continued primarily through the outer fingers as long as constant pressure was maintained in the system by continued pouring. It was observed here, and throughout this investigation, that once a flow pattern was established in a gating system, it was difficult to change it.

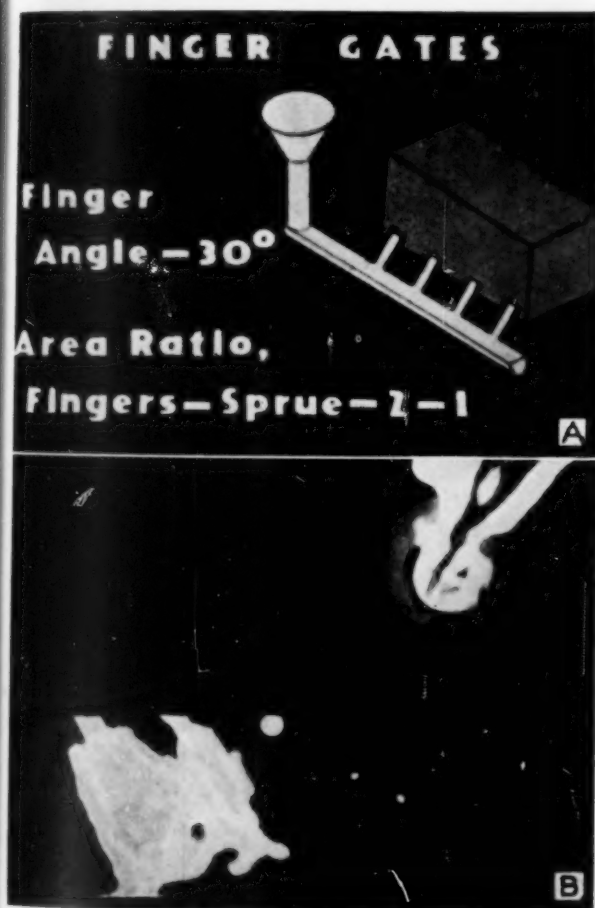


Fig. 9—Finger gates

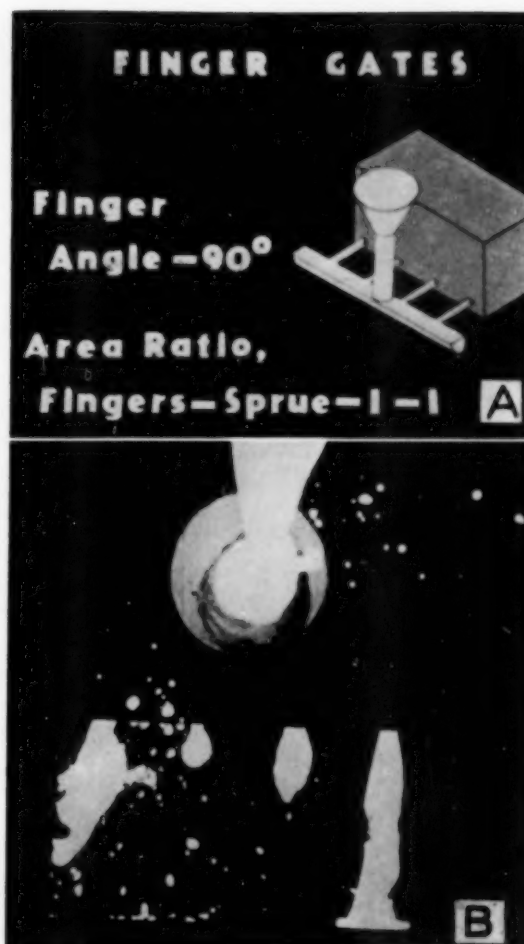


Fig. 10—Finger gates

The location of the sprue was next changed from center to end of the series of four fingers. Figures 8A through 8D, with no change in the area ratio and finger angle. The second finger from the sprue fed first, as shown in Fig. 8B, but most of the metal entered the mold through the finger farthest from the sprue, Figs. 8C and 8D. Swirling turbulence became apparent in the side of the mold opposite the sprue. This disturbance lessened as the mold filled and back flow from fingers became more uniform.

With a change in the angle, between the fingers and the casting to 30 deg., or 150 deg. to the direction of flow of metal in the runner, as in Figs. 9A and 9B, the finger farthest from the sprue fed first.

The effect of reducing the total finger area to equal that of the sprue was next investigated. Although the initial flow began from the fingers farthest from the sprue (Fig. 10B) flow from all the fingers soon became more uniform. A rolling type of turbulence replaced the swirling kind of turbulence noted previously with the larger finger to sprue ratio. When the location of the sprue was shifted from the center to the end of the runner, molten steel entered first through the fingers nearest the sprue, Fig. 11B, but shortly thereafter flow from all fingers became relatively uniform and rolling turbulence occurred. When an additional change of finger angle to mold of 30 deg. was made flow began at about the same time through all four fingers with the



Fig. 11—Finger gates

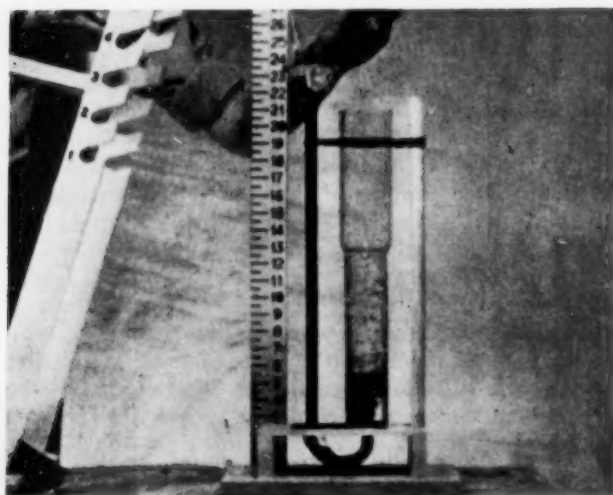


Fig. 12—Normal horn gate model

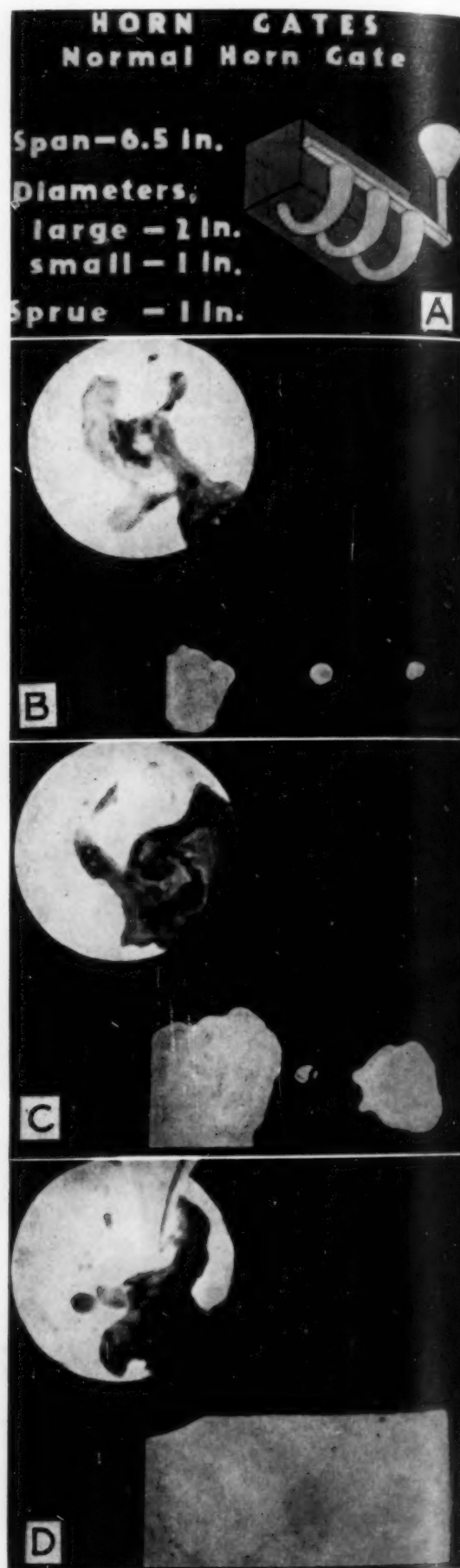


Fig. 13 (Right)—Horn gates—normal



farthest ones carrying the greater quantity of metal. The amount of flow became fairly uniform at a later stage.

This investigation of finger gating indicated that when the combined finger area was greater than that of the sprue, feeding was not uniform and swirling turbulence was the result. When, however, the combined finger area equaled that of the sprue, feeding was relatively uniform but was accompanied by a rolling turbulence.

### Horn Gating

Horn gates are designed for use on castings which cannot be gated on the edge or where it is not considered advisable to pour metal down through a gate on the top. A normal horn gate is smaller in cross section at the casting than at the down gate and a reversed horn gate is tapered in the opposite direction. A single normal horn gate produced a fountain-like jet and severe rocking of the metal in the mold. The height of this initial fountain-like jet was more apparent in a water-in-glass model. Dye introduced into the pouring stream when the model was nearly filled (Fig. 12) showed that the incoming liquid continued to rise like a jet through the water in the model.

The effect of placing three normal horn gates in a row with the sprue located on one end is shown in Figs. 13A through 13D. With this design metal en-

tered initially through the horn next to the sprue (Fig. 13B) but later most of it entered through the horn farthest from the sprue (Fig. 13C). After the mold was partially filled, the middle horn began to feed as seen in Fig. 13D.

Reversed horn gates are commonly believed to pass metal with little turbulence. With a single reversed horn a jet of molten steel again was produced but its height was slightly less than that produced by the normal horn gate. When three reversed horn gates were used in a row with the sprue on one end (Figs. 14A and 14B) the horn farthest from the sprue fed first and continued to deliver most of the metal as shown by the disturbance on the surface. Turbulence at the center of the partially filled mold indicated that the middle horn did feed to some extent at this stage. When the sprue was placed in the center of four reversed horn gates, as in Figs. 15A and 15B, the horns farthest from the sprue did all the feeding as shown by the surface.

### Step Gating

The purpose of a step gate is to produce a steady temperature gradient from the bottom of the casting to its riser without the spattering characteristic of top pouring. As the metal level in the casting reaches each succeeding higher step, this step should begin and the lower steps stop feeding.

The first step-gate system studied (Fig. 16A) was

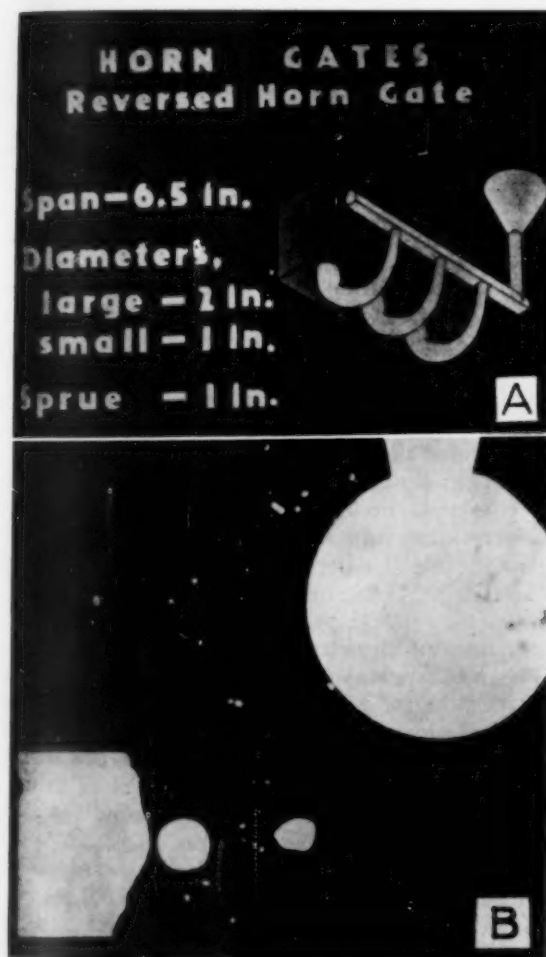


Fig. 14—Horn gates—reversed

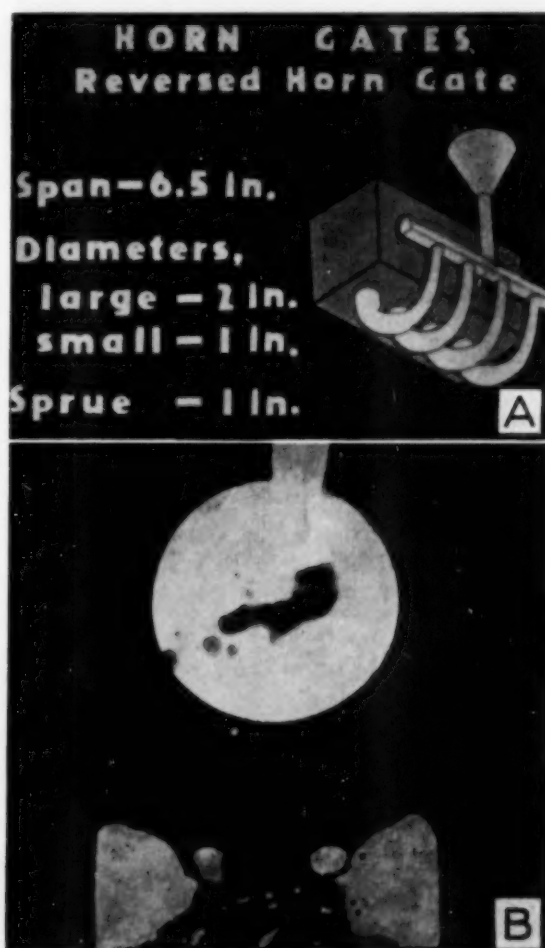


Fig. 15—Horn gates—reversed

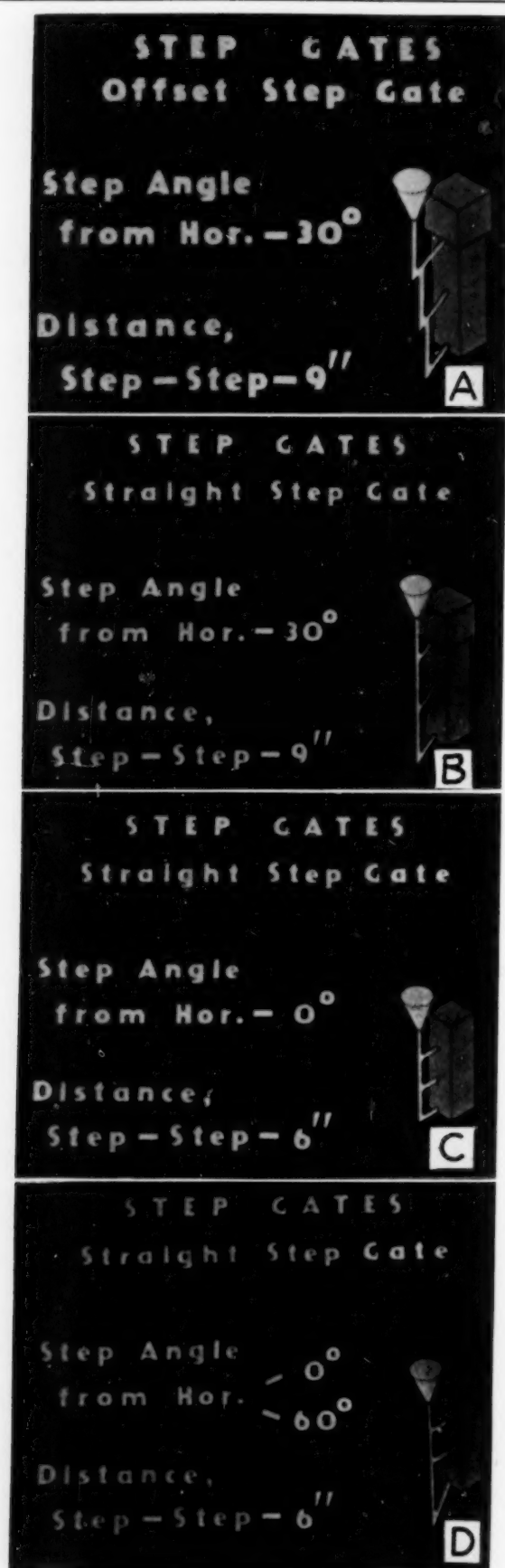


Fig. 16—Step gates

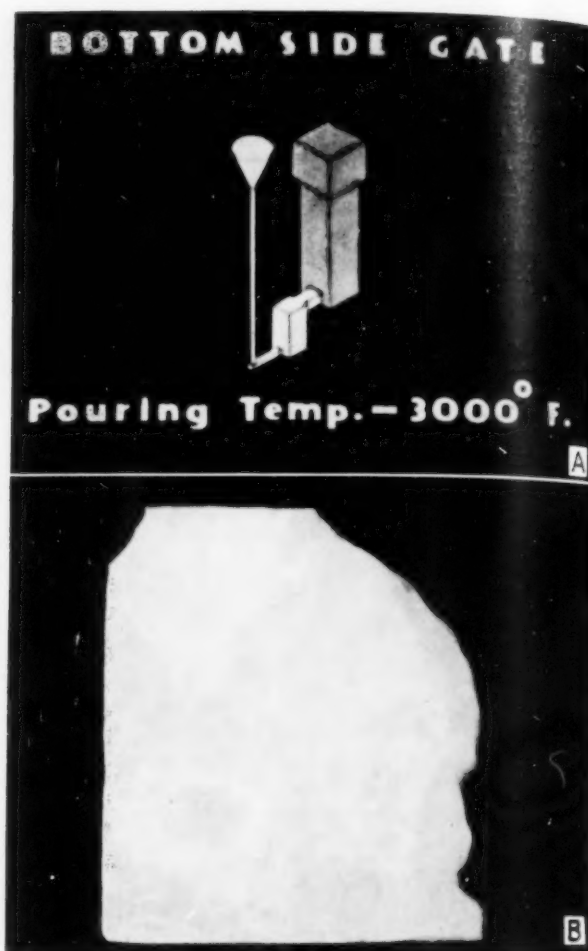


Fig. 17—Bottom side gate

an offset gate in which the vertical distance between steps was 9 in. and each sprue was offset 1 in. horizontally. The steps were inclined upward 30 deg. from the horizontal. With this gate, the bottom step fed initially but before the metal reached the middle step, the top step began to feed and the bulk of the metal entered through the top step.

The next system studied (Fig. 16B) was a straight step gate with no change in distance between steps and inclination from the previous example. With this design the metal entered the mold only through the bottom step. As the metal in the mold reached the middle and upper steps, it ran back into the gating system instead of out into the mold.

The next system (Fig. 16C) had three horizontal steps only 6 in. apart. The metal entered mainly through the lower step but a small quantity did dribble through the upper ones before the level of the metal actually reached them. When the level of metal in the mold finally reached the upper steps, metal ran back into them as in the previous system. With the lowest step, as in Fig. 16D, inclined upwards 60 deg. from the horizontal, the metal first entered the mold through this step, and as pouring continued the second step fed when the metal level reached it. This was the only step gating system that showed an upper step feeding in the desired sequence.

### Bottom Side Gate

In the bottom side gating system, shown in Figs. 17A and 17B, a closed mold cavity was placed before the mold to decrease the velocity of the metal. Metal entered the mold more evenly through this system than through any other studied. Light flashes on the surface of the metal were similar to those that were observed in the riser gate system.

### Conclusions

These experiments showed that many gating systems did not function as commonly supposed.

Whirl, riser and horn gates were ineffective in preventing turbulence in the mold. In multiple finger

gating systems, often most of the feeding was accomplished by the fingers farthest from the sprue. When all the fingers did feed uniformly, a rolling type of turbulence appeared in the mold. With step gates only one design showed feeding in the desired sequence. No gating system prevented turbulence.

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# EFFECT OF TITANIUM ON GRAIN SIZE AND TENSILE PROPERTIES OF AN ALUMINUM-4.5 PER CENT COPPER (NO. 195) CASTING ALLOY\*

By

W. E. Sicha\*\* and R. C. Boehm

## ABSTRACT

The effects of titanium additions of 0.05, 0.10, 0.20 and 0.40 per cent on the grain size and tensile properties of an aluminum-4.5 per cent copper (No. 195) casting alloy were determined.

It was found that, with closely controlled melting and pouring practices, there was an optimum titanium addition which would produce a very fine grain at any given pouring temperature. Larger titanium additions had little further effect on the grain size. The quantity of titanium required to produce this fine grain increased as the pouring temperature was raised. The control of the grain size effected by the 0.05 through 0.20 per cent titanium additions was diminished by remelting. The alloy with the highest titanium content in this range exhibited the grain refining effect of the titanium for the largest number of remelts. There was no loss in the grain refining effect from the 0.40 per cent titanium addition with remelting.

The grain refining effect of the titanium was attributed to  $TiAl_3$  particles which were reduced in size by the peritectic reaction which was not completed. These particles served as nuclei to start crystallization in the melt.

Titanium additions to the aluminum-4.5 per cent copper (No. 195) casting alloy affected the tensile properties in several ways. The grain refinement effected by the titanium addition markedly increased the tensile strength and elongation. This effect was opposed by the deleterious influence on the tensile properties of the undissolved  $TiAl_3$  particles and the coarsening and concentration of the insoluble  $\alpha-Al-Fe-Si$  constituent at the grain boundaries. Increasing the titanium progressively retarded the artificial aging which decreased the tensile strength and increased the elongation. Increasing the melting temperature decreased the tensile properties. The tensile properties obtained were the resultants of these various effects.

## Introduction

It has been recognized for a long time that titanium serves as a grain refiner in aluminum-zinc, aluminum-zinc-magnesium and aluminum-copper alloys. The binary aluminum-titanium constitutional diagram used in this investigation to explain the effects of titanium on the grain size and tensile properties of an aluminum-4.5 per cent copper (No. 195) casting alloy is shown in Figs. 1 and 2.<sup>1</sup> Investigators have reported that  $TiAl_3$  was the only intermediate

\* Abstracted from a thesis submitted to Case Institute of Technology by F. C. Boehm in partial fulfillment of the requirements for the degree of Master of Science in Physical Metallurgy.

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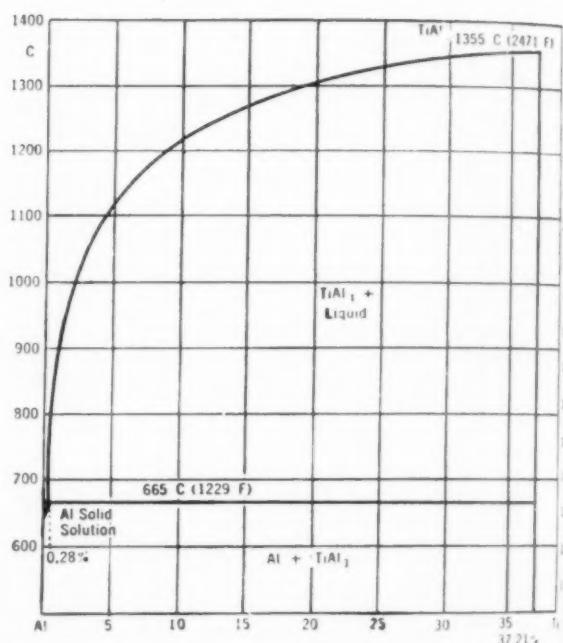


Fig. 1—System Al-Ti, aluminum end of equilibrium diagram.

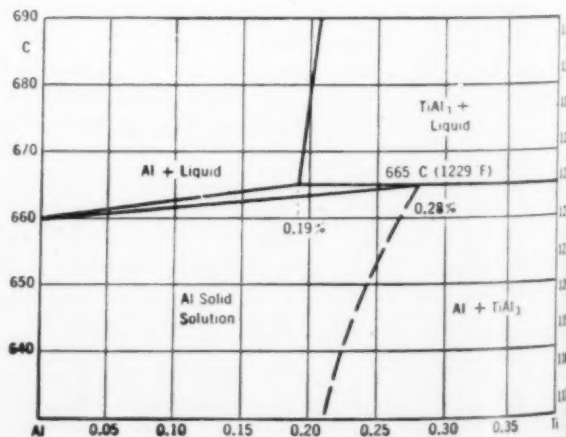


Fig. 2—System Al-Ti, detail of aluminum end of equilibrium diagram, showing the peritectic reaction.

phase containing titanium found in the Al-Cu-Ti<sup>(2)</sup> or the Al-Fe-Ti<sup>(3)</sup> ternary systems.

Previous investigations have shown that 0.05 to 0.50 per cent titanium would produce fine grains in thin and heavy sections of castings poured at relatively high temperatures.<sup>(4,5,6,7)</sup> Titanium contents above about 0.5 per cent were found to be unnecessary because there was no further refinement of the grain and the excess of titanium formed lamellar plates of TiAl<sub>3</sub> which cause lower physical properties. At the present time, a maximum of 0.20 per cent titanium is specified for most commercial alloys to which titanium is added.

Several investigators reported that the reduction in grain size produced by the titanium persisted after remelting.<sup>(6,8)</sup> Others found that the grain refining effect of the titanium was reduced in remelting but that the grain refining effect would be restored by a second 0.05 per cent addition of titanium.<sup>(7,9,10)</sup> In those instances, where the grain refining effect was decreased, there was no loss of titanium and the second addition increased the quantity of titanium in the alloy.<sup>(9,10)</sup> The best grain refinement was obtained when the titanium was added in the ladle and when the melt was cast shortly after the titanium addition.<sup>(7)</sup>

It has been suggested that TiAl<sub>3</sub> solidifying at high temperatures (Figs. 1 and 2) served as nuclei to initiate crystallization of the melt.<sup>(6)</sup> A more recent investigation has indicated that the peritectic reaction liquid Al + TiAl<sub>3</sub> → Al (solid solution) which occurs at 665 C (1229 F) may be responsible for the grain refinement but the exact mechanism has not yet been established.<sup>(8)</sup> Also, the reduction in the grain refining effect of the titanium by remelting has not been explained.

#### Material and Procedure

As a preliminary step in studying the effect of titanium, a heat of No. 195 aluminum casting alloy with a low titanium content (0.009 per cent) was prepared. Titanium was introduced into melts of the base alloy as an aluminum-5 per cent titanium rich alloy. A high frequency induction furnace was used for melting. The power input to the induction furnace was adjusted so that each test melt was poured 5 to 7 min. after a temperature of 1250 F had been reached.

The metal was melted, titanium was added at 1250 F, the temperature increased to about the pouring temperature while fluxing for 3 min. with dry nitrogen. It was skimmed, and then poured 2 to 3 min. after fluxing into green sand, two-bar standard 0.505 in. diameter, test bar molds. The charge of the remelted heats consisted of the gates and risers of the test bar castings plus the pigged metal. In all cases, the maximum melting temperature was the same as the pouring temperature.

No machining or finishing was performed on the gage section of the standard test bars used to obtain tensile properties. The tensile properties of at least four specimens were averaged for each condition investigated.

The test bars were solution heat treated for 16 hr.

TABLE 1—CHEMICAL COMPOSITIONS OF THE NO. 195 ALUMINUM CASTING ALLOYS

Number of Titanium Remelting Addition, Operations		Analyzed Composition Per Cent			
		Titanium	Copper	Iron	Silicon
Poured at 1320 F					
..	...	0.008	4.32	0.71	0.84
..	...	0.005	4.42	0.65	0.76
..	0.05	0.08	...	...	...
..	0.05	0.06	...	...	...
..	0.10	0.13	...	...	...
..	0.10	0.12	...	...	...
..	0.20	0.22	...	...	...
..	0.20	0.20	...	...	...
..	0.40	0.40	4.09	0.62	0.73

The tensile properties and grain sizes obtained from the duplicate test heats were averaged.

Poured at 1450 F					
..	...	0.009	4.60	0.68	0.78
1	...	0.006	...	...	...
2	...	0.009	4.56	0.70	0.82
3	...	0.015	4.56	0.74	0.83
4	...	0.006	...	...	...
5	...	0.006	4.42	0.66	0.77
..	0.05	0.06	4.35	0.69	0.80
1	0.05	0.06	4.48	0.74	0.82
2	0.05	0.06	...	...	...
3	0.05	0.07	4.34	0.68	0.80
4	0.05	0.06	...	...	...
..	0.10	0.10	4.56	0.70	0.82
1	0.10	0.12	4.35	0.74	0.84
2	0.10	0.12	...	...	...
3	0.10	0.14	...	...	...
4	0.10	0.11	...	...	...

Poured at 1450 F					
..	0.20	0.23	4.35	0.68	0.79
1	0.20	0.20	4.32	0.71	0.81
2	0.20	0.28	...	...	...
3	0.20	0.23	...	...	...
4	0.20	0.24	...	...	...
5	0.20	0.28	...	...	...
6	0.20	0.28	...	...	...
7	0.20	0.24	3.91	0.67	0.70
..	0.40	0.39*	4.56	0.67	0.83
		0.39**	...	...	...
		1.35***	...	...	...
1	0.40	0.41	...	...	...
2	0.40	0.44	...	...	...
3	0.40	0.39	...	...	...
4	0.40	0.40	...	...	...
5	0.40	0.44	...	...	...
6	0.40	0.44	...	...	...
7	0.40	0.45	3.87	0.63	0.72
8	0.40	0.40	3.83	0.67	0.68

#### Poured at 1500 F

..	...	0.006	4.51	0.69	0.80
..	0.05	0.06	...	...	...
..	0.10	0.13	...	...	...
..	0.20	0.20	4.25	0.73	0.82

\* Chemical sample cast from the first metal poured.

\*\* Chemical sample cast after half the metal was poured.

\*\*\* Chemical sample cast from the metal in the bottom of the crucible.

at 960 F, quenched in boiling water, and artificially aged for 5 hr. at 310 F. Artificial aging was started 3 hr. after quenching had been completed.

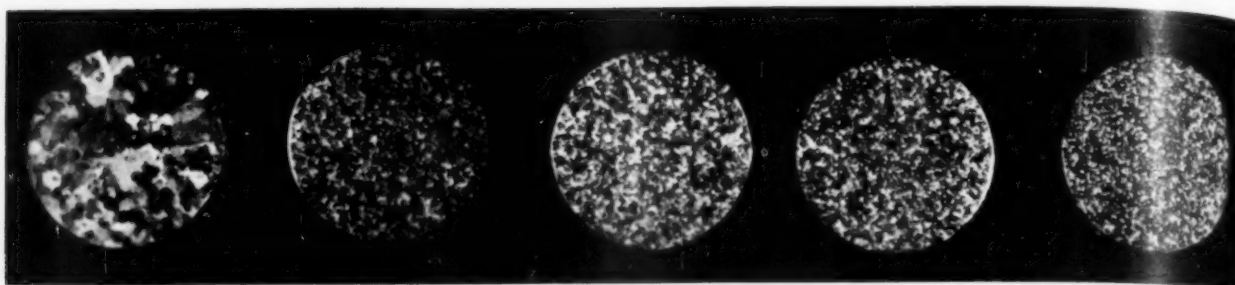


Fig. 3—Effect of titanium on the grain size of No. 195 aluminum casting alloy test bars poured at 1450 F.

From left to right—0.00, 0.05, 0.10, 0.20 and 0.40 per cent titanium added respectively. Mag. 2X.

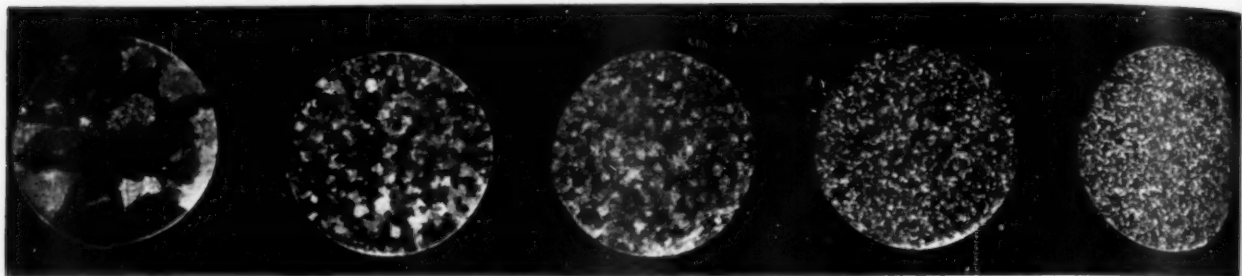


Fig. 4—Effect of titanium on the grain size of No. 195 aluminum casting alloy test bars poured at 1450 F.

From left to right—0.00, 0.05, 0.10, 0.20 and 0.40 per cent titanium added respectively. Mag. 2X.

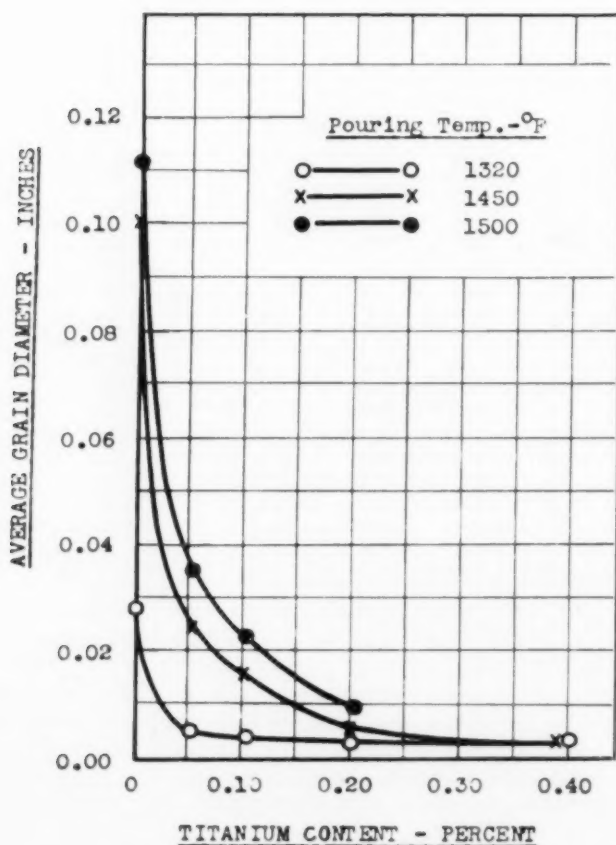


Fig. 5—Effect of titanium on the grain size of No. 195 aluminum casting alloy test bars poured at varied temperatures, solution heat treated and artificially aged.

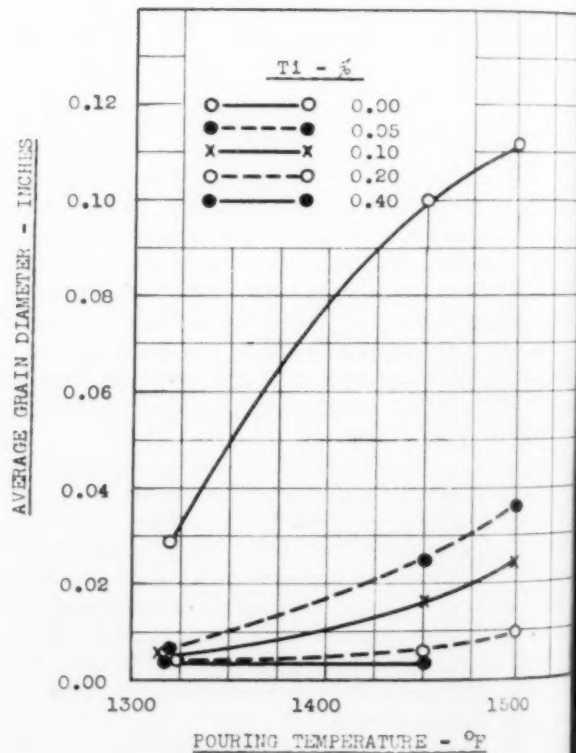


Fig. 6—Effect of pouring temperature on the grain size of No. 195 aluminum casting alloy test bars with varied titanium contents, solution heat treated and artificially aged.



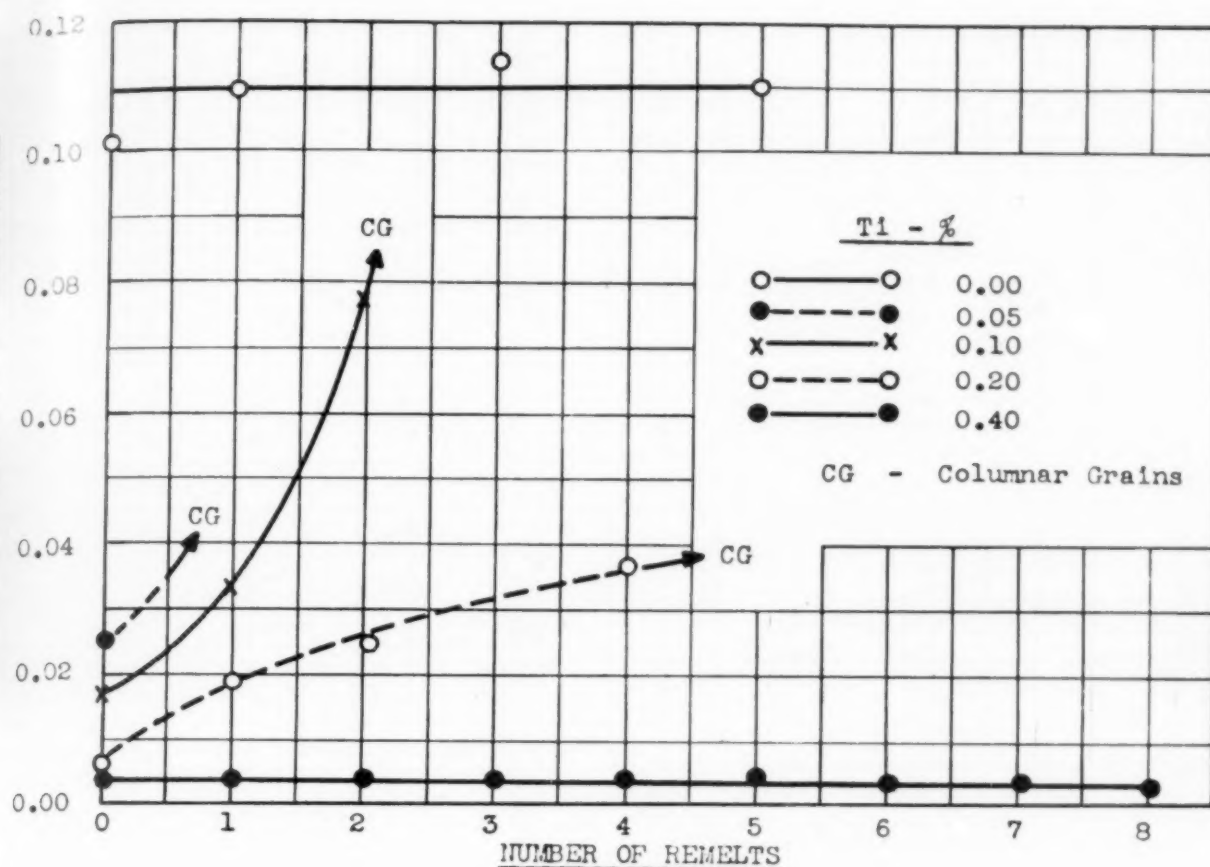


Fig. 7—Effect of remelting on the grain size of No. 195 aluminum casting alloy test bars poured at 1450 F, solution heat treated and artificially aged.

Grain size determinations were made on longitudinal sections adjacent to the fracture which were polished and etched. Four readings were averaged to obtain the grain size of each test section measured.

The samples for chemical analyses were milled from 1/4-in. thick chill cast slabs which were poured from each melt. The analyses are listed in Table 1.

### Results and Discussion

#### GRAIN SIZE.

The grain size of the No. 195 aluminum casting alloy was influenced by several variables. These factors will be discussed individually.

**Effect of Titanium on the Grain Size.** The effect of titanium on the grain size of No. 195 aluminum casting alloy is illustrated by photomicrographs in Figs. 5 and 4 and is shown graphically in Figs. 5 and 6. The test bars poured at 1320 F from a melt of the base (low titanium) alloy would be considered commercially fine grained. However, the addition of 0.05 per cent titanium effected a marked reduction in the size of the grains. Additions of 0.10 through 0.40 per cent of titanium produced grain sizes only slightly smaller than the 0.05 per cent titanium addition.

It was necessary to add 0.20 per cent titanium to obtain grain sizes of the same order in test bars poured at 1450 F as that of those produced at 1320 F with 0.05 per cent titanium. It was apparent that, with the closely controlled melting practice employed in this investigation, a definite quantity of titanium was required to produce a very fine grain at each

pouring temperature. Increasing the titanium above this amount had little further effect on grain size.

**Effect of Remelting on the Grain Size.** The addition of titanium to the base alloy melts produced fine equiaxed grains. When columnar grains were obtained by remelting, it was considered that the titanium was only partially affecting the grain size. This classification was followed even though the columnar grains were smaller than the large, equiaxed grains.

The effect of remelting on the grain refinement obtained from titanium in test bars poured at 1450 F is illustrated graphically in Fig. 7. In this graph, the sizes of only the equiaxed grains are plotted. These curves showed that the grain refining effect of 0.05 per cent titanium addition was lost after one remelting operation. While the grain size of the alloy in which 0.10 per cent titanium was introduced increased on the first remelting, the alloy would be considered commercially fine grained. The grains of the test bars poured from the second remelt were large. The fine equiaxed grains of the alloy with the added 0.20 per cent titanium increased slightly and progressively with remelting, through the fourth remelt. Columnar grains were noted in the test bars poured from remelts 5 through 7. Therefore, the 0.20 per cent titanium only partially affected the grain size of the alloy after the fourth remelt. The grain size after eight remelts

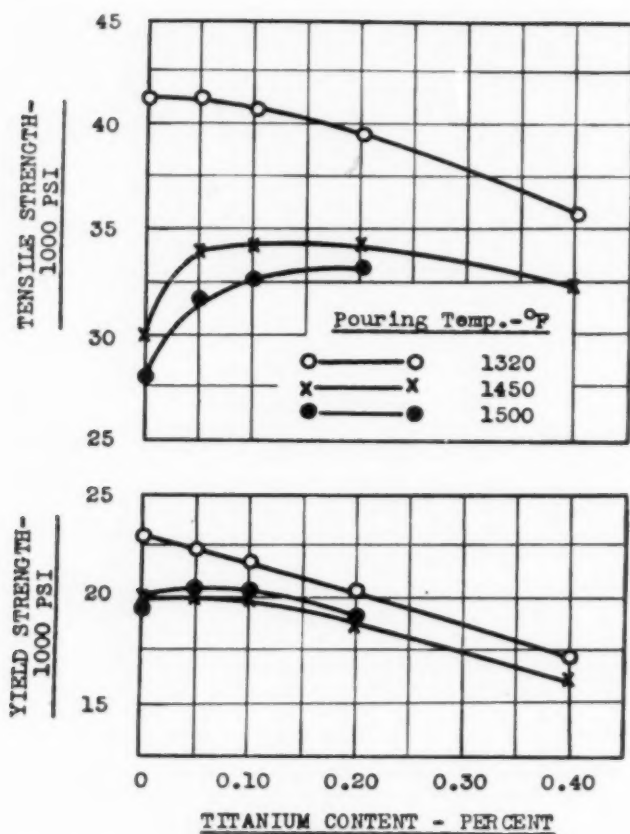


Fig. 8—Effect of titanium on the mechanical properties of No. 195 aluminum casting alloy test bars poured at varied temperatures, solution heat treated and artificially aged.

of the alloy which contained 0.40 per cent titanium was the same as that obtained from the first melt of the alloy. Consequently, remelting had no effect on the grain size of this alloy.

No noticeable decrease in the titanium of the alloys by remelting could be detected by chemical analyses. This eliminated the possibility of the loss in the grain refinement being caused by an actual reduction in the titanium content.

*The Phase  $TiAl_3$  as the Grain Refiner.* Titanium in amounts of 0.19 per cent or less in aluminum alloys would be in solution in liquid or solid aluminum under equilibrium conditions (Figs. 1 and 2). Some titanium would be in the form of  $TiAl_3$  crystals in the liquid alloys containing from 0.19 through 0.28 per cent titanium, but the  $TiAl_3$  crystals would dissolve in solid aluminum at the peritectic temperature. Crystals of  $TiAl_3$  would exist in both liquid and solid aluminum in alloys containing more than about 0.28 per cent titanium. It is apparent that under equilibrium conditions,  $TiAl_3$  particles cannot be present in liquid alloys containing about 0.19 per cent titanium or less and, therefore, could not serve as nuclei to initiate crystallization. Since the  $TiAl_3$  crystals in molten alloys containing from 0.19 through 0.28 per cent titanium under equilibrium conditions would react at the peritectic temperature to form aluminum solid solution and liquid, there would be no  $TiAl_3$

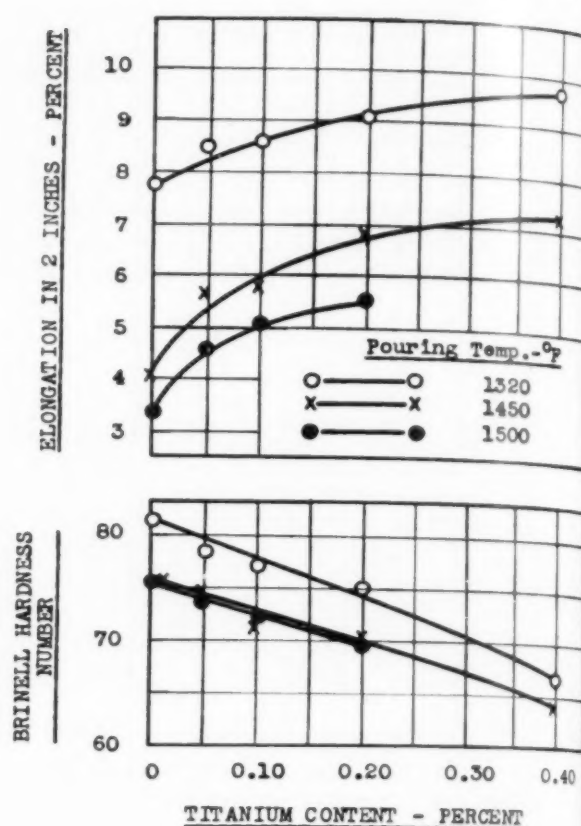


Fig. 9—Effect of titanium on the mechanical properties of No. 195 aluminum casting alloy test bars poured at varied temperatures, solution heat treated and artificially aged.

crystals present to perform the function of initiating crystallization in the remaining melt.

In practice, however, non-equilibrium conditions generally prevail. In this investigation, the titanium was introduced into the No. 195 alloys as an aluminum-5 per cent titanium rich. Most of the titanium in the rich alloy would be in the form of large  $TiAl_3$  crystals. As a result,  $TiAl_3$  crystals would be introduced into the test melts by any titanium addition. As the localized concentration of the titanium from the rich alloy addition would be reduced by distribution of the  $TiAl_3$  crystals throughout the melts, some of the crystals in the No. 195 alloys containing titanium would start to dissolve in the molten alloy. The extent of the solution would depend on the quantity of titanium added, the temperature of the melt, and the time interval between the titanium addition and the pour. Molten alloys containing more than about 0.19 per cent titanium would retain some  $TiAl_3$  crystals regardless of the time interval between the titanium addition and the pour.

If it were assumed that  $TiAl_3$  particles served as nuclei to initiate crystallization of the melt, titanium additions of 0.05 through about 0.19 per cent to the No. 195 aluminum alloy could effect grain refinement only under non-equilibrium conditions. The alloys would have to be poured before the major portion of the  $TiAl_3$  crystals from the rich had dissolved in

the liquid alloy. If the titanium were added in small amounts such as 0.05 per cent or less, the localized concentration of the  $TiAl_3$  crystals in the zone of the aluminum-5 per cent titanium alloy rich in the melt would be reduced below about 0.19 per cent titanium. However, the  $TiAl_3$  crystals would dissolve rapidly in the molten alloy because of the high melting temperature, 1355 C (2471 F), of the crystals. Increasing the quantity of the alloy rich addition would increase the localized concentration of the titanium. More time would then be required to distribute and dissolve the  $TiAl_3$  crystals in the melt. Some  $TiAl_3$  crystals could be present in No. 195 aluminum casting alloys that are poured immediately after the titanium addition regardless of the titanium content.

#### Solubility of Titanium

The solubility of the titanium in the liquid alloys increases with the melting temperature. Increasing the temperature would also cause greater movement and distribution of the  $TiAl_3$  crystals in the melt. These factors would reduce the number of  $TiAl_3$  particles remaining in the melts at the time of pour and would result in a coarsening of the grain in alloys containing less than about 0.19 per cent titanium. This is in agreement with the experimental data. Thus the grain size of the test bars of the alloys containing 0.05 through 0.20 per cent titanium increased as the pouring temperature was raised.

In melts containing from 0.19 through 0.28 per cent titanium, some of the  $TiAl_3$  crystals which existed in the liquid would dissolve. On remelting these alloys, crystals of  $TiAl_3$  would again be formed. If the formation of these crystals requires long time periods at the melting temperature of the alloy, a reduction in the number of  $TiAl_3$  crystals and therefore a loss in the grain refining effect of the titanium with remelting would be expected. However, since some  $TiAl_3$  crystals will be formed in the molten alloy, the titanium could be expected to partially affect the grain size. This was confirmed by the experimental data. There was a progressive increase in the grain size of the test bars of the first four remelts of the alloy which contained 0.20 per cent titanium. After the fourth remelt, some of the grains were columnar. Therefore, titanium had a partial effect on the grain size after the fourth remelting.

In alloys containing more than 0.28 per cent titanium,  $TiAl_3$  particles would exist in both the liquid and solid states. Therefore, there should be no loss in grain refinement by remelting. In this investigation, the grain size after the eighth remelting of the alloy containing 0.40 per cent titanium was as fine as that obtained in the initial melt.

#### $TiAl_3$ Crystal Form

The  $TiAl_3$  crystals observed in the microstructure were in the form of large lamellar plates. These were not uniformly distributed but occurred in small groups of crystals. It is difficult to understand how these large crystals could serve as nuclei to initiate crystallization throughout the melt. Further, a grain refinement was obtained in the base No. 195 alumi-

num casting alloy by the addition of 0.05 and 0.10 per cent titanium but no  $TiAl_3$  crystals were seen in the microstructure. It can be deduced, therefore, that the grain refinement was produced in these alloys by  $TiAl_3$  particles of submicroscopic dimensions. Such particles could serve as nuclei to initiate crystallization.

The grain refinement caused by titanium has been attributed to the peritectic reaction which occurs as follows: liquid Al +  $TiAl_3 \rightleftharpoons$  Al (solid solution).<sup>(2)</sup> Under non-equilibrium conditions which usually prevail,  $TiAl_3$  crystals would exist in the melts regardless of the titanium content. A portion of the  $TiAl_3$  crystals will undergo the peritectic reaction. However, the cooling rate of alloys under normal casting conditions would be such that the peritectic reaction would not be completed. Therefore, these  $TiAl_3$  crystals would no longer be in the form of the large lamellar plates but reduced to microscopic or sub-microscopic dimensions. In this reduced size, the  $TiAl_3$  particles would produce grain refinement in the alloy. It would appear that the grain refinement effected by titanium additions to No. 195 aluminum casting alloys can be attributed to  $TiAl_3$  particles that are appreciably reduced in size by the peritectic reaction.

#### Tensile Properties

The tensile properties were influenced by several factors which will be discussed individually.

*Effect of Titanium on the Tensile Properties.* The effects of titanium on the tensile properties of No. 195 aluminum casting alloy are presented in Figs. 8 and 9. In test specimens cast at 1320 F, increasing the titanium content reduced the tensile strength, yield strength and hardness but increased the elongation. The base No. 195 aluminum alloy was inherently fine grained at the pouring temperature of 1320 F. The decreased tensile properties with increasing titanium obtained at this pouring temperature can be attributed to more undissolved  $TiAl_3$  particles in the melts, a concentration of the insoluble  $\alpha$ Al-Fe-Si constituent in the grain boundaries and a slower rate of artificial aging. The slower rate of artificial aging with increasing titanium would also explain the reduced yield strength and increased elongation.

The grain sizes of the base No. 195 alloys poured at 1450 F and 1500 F were large. The 0.05 per cent titanium addition effected a marked grain refinement which was reflected by substantial increases in tensile strength and elongation. Additions of 0.10 and 0.20 per cent titanium produced a further smaller reduction in the grain size and small additional increases in tensile strength and elongation. The improvement in tensile properties caused by the grain refinement obtained by increasing titanium to 0.20 per cent, at the pouring temperatures of 1450 F and 1500 F, were opposed or supplemented by the other factors influencing the properties of the alloys poured at 1320 F. In the range of 0.20 to 0.40 per cent titanium, there was no apparent change in the grain size of the No. 195 alloys poured at 1450 F. The tensile property curves of the alloys containing 0.20 to 0.40 per cent titanium which were poured at 1450 F were approximately parallel to the curves obtained from alloys of similar titanium contents that were poured at 1320 F.



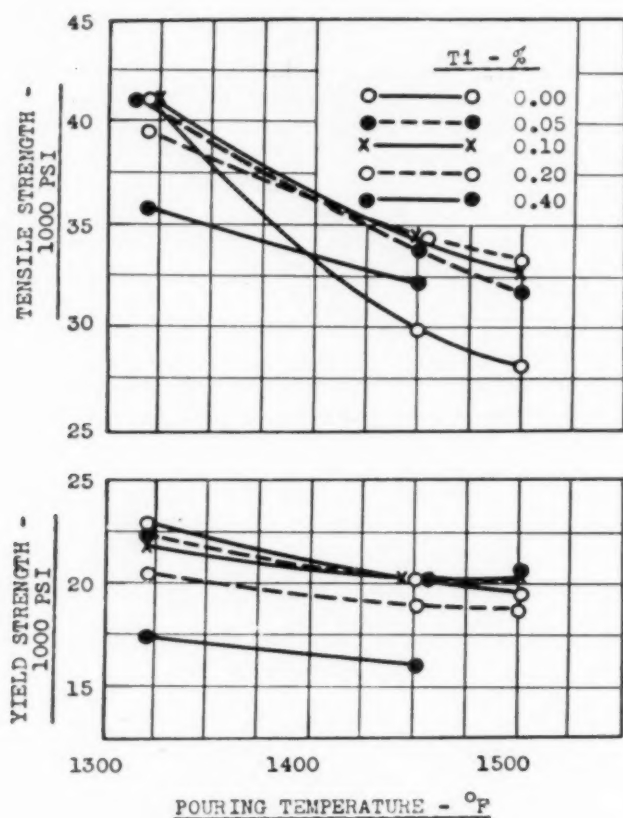


Fig. 10—Effect of pouring temperature on the mechanical properties of No. 195 aluminum casting alloy test bars with varied titanium contents, solution heat treated and artificially aged.

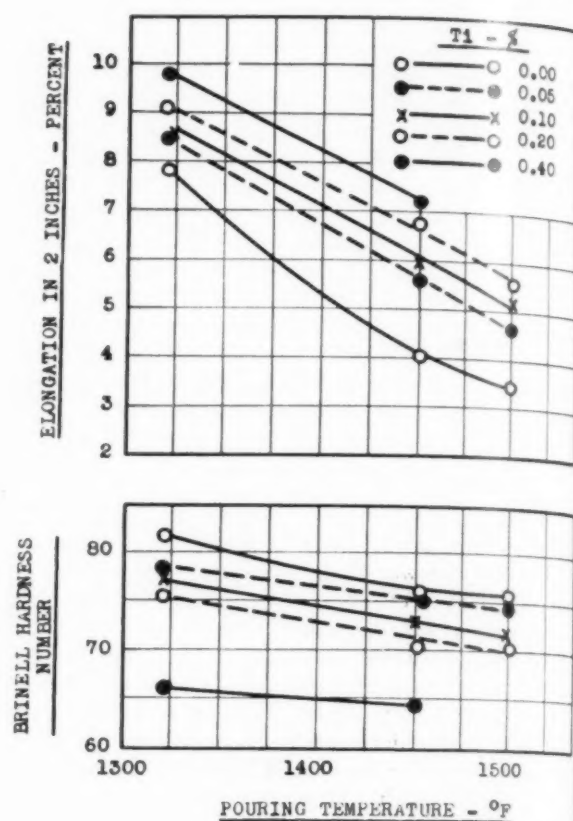


Fig. 11—Effect of pouring temperature on the mechanical properties of No. 195 aluminum casting alloy test bars with varied titanium contents, solution heat treated and artificially aged.

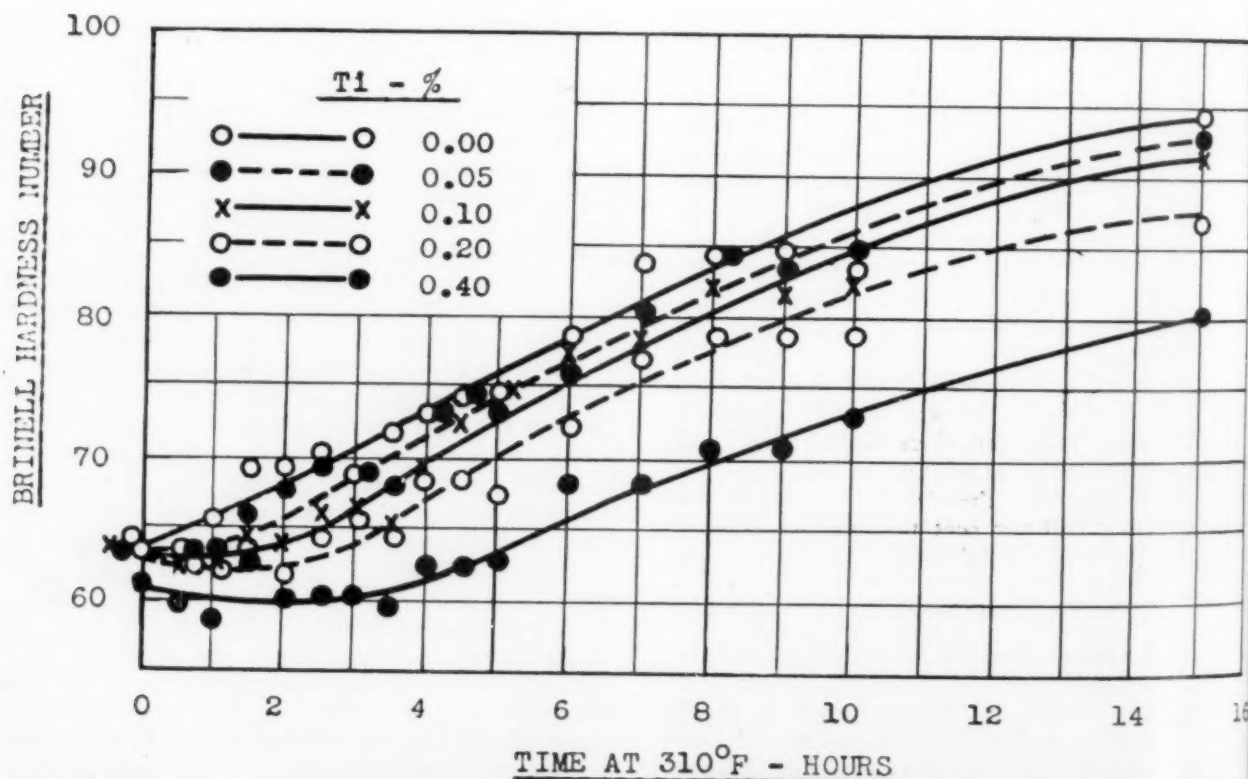


Fig. 12—Precipitation hardening curves for No. 195 aluminum casting alloy test bars with varied titanium

contents, poured at 1450 F, solution heat treated and artificially aged.

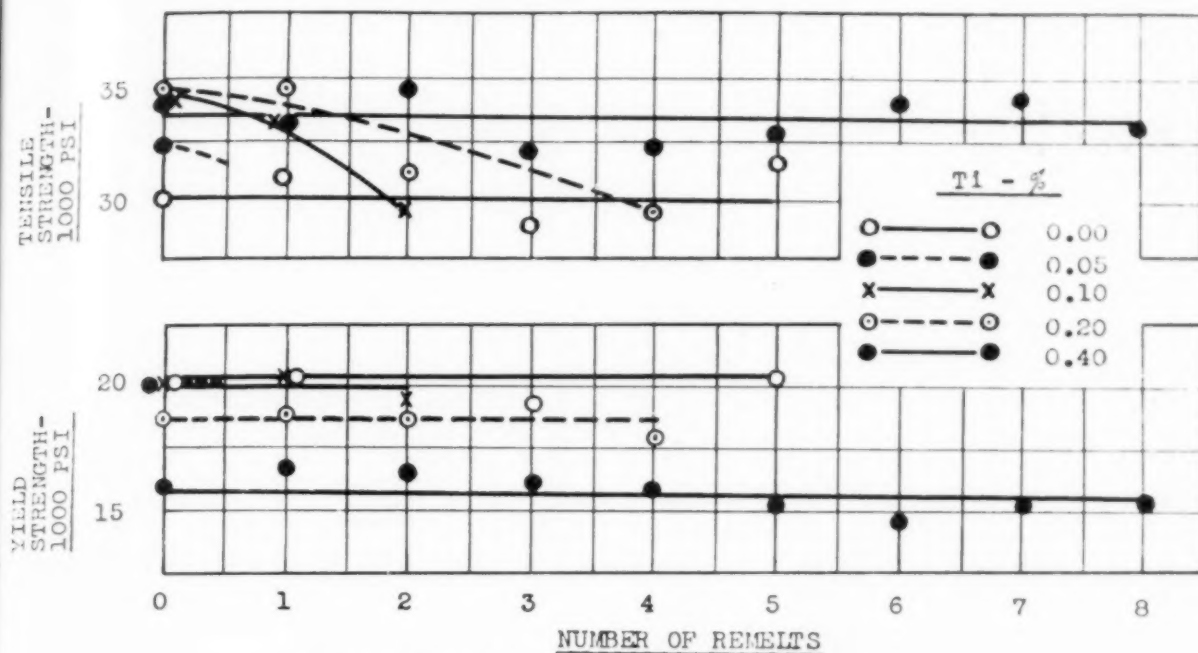


Fig. 13—Effect of remelting on the mechanical properties of No. 195 aluminum casting alloy test bars with varied titanium contents, poured at 1450 F, solution heat treated and artificially aged.

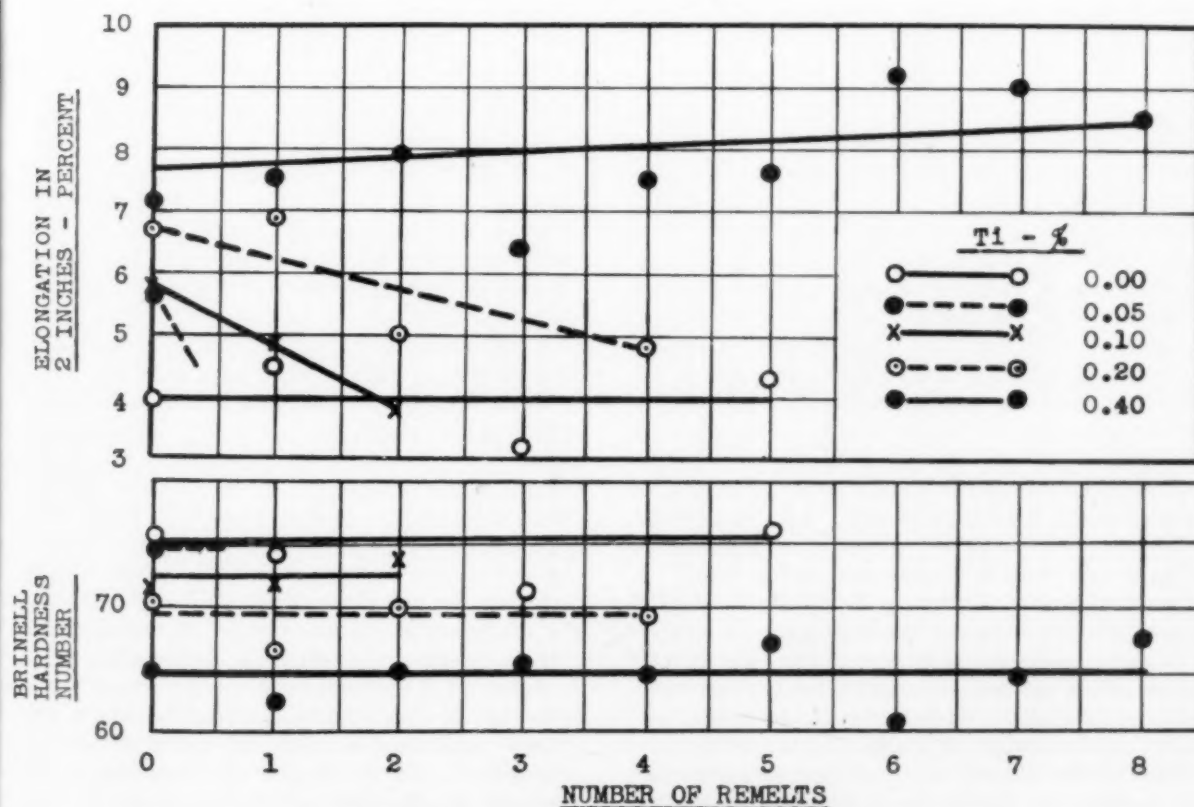


Fig. 14—Effect of remelting on the mechanical properties of No. 195 aluminum casting alloy test bars, with varied titanium contents, poured at 1450 F, solution heat treated and artificially aged.

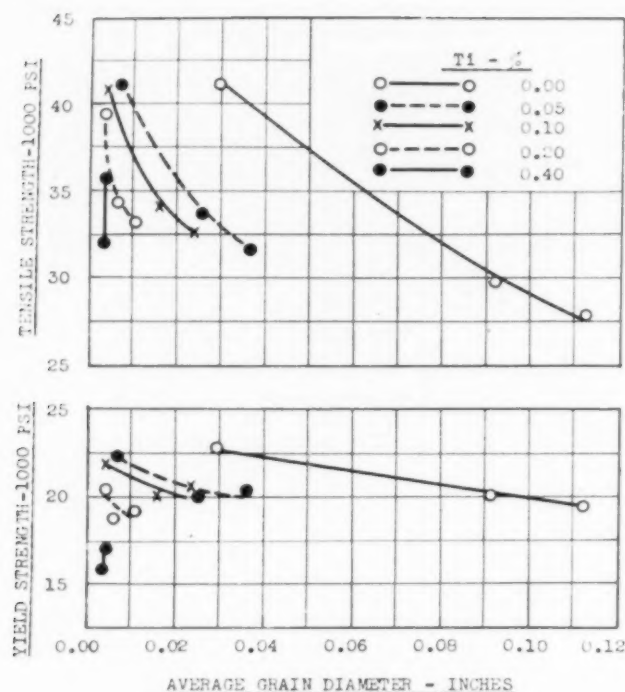


Fig. 15—Effect of grain size on the mechanical properties of No. 195 aluminum casting alloy test bars with varied titanium contents, solution heat treated and artificially aged. Coarser grain sizes were obtained by increasing the pouring temperature.

*Effect of Pouring Temperature on the Tensile Properties.* Figures 10 and 11 show the effect of the pouring temperature on the tensile properties of the No. 195 alloys. Increasing the pouring temperature substantially reduced the tensile properties even though there was no detectable change in grain size. Generally, coarse dendritic structures are associated with high pouring temperatures. In an individual cast grain the dendrite arms are outlined by undissolved or insoluble constituents and the cross section of the arms varies in size depending on the rate of solidification. Reduced tensile properties are obtained in fine grained alloys if the dendrite arm size is large and, therefore, this is the more important factor in determining the tensile properties. The marked reduction in tensile properties obtained with increased pouring temperature was attributed to this factor.

*Effect of Artificial Aging on the Hardness.* Figure 12 shows the precipitation hardening curves for No. 195 alloys with varied titanium contents. The Brinell hardness values revealed that increased titanium progressively retarded the artificial aging. Increasing the titanium in the range investigated may reduce the solubility of the copper in aluminum and therefore would reduce the hardness. A second explanation may be that the titanium retards the rate of precipitation of the  $\text{CuAl}_2$  particles.

*Effect of Remelting on the Tensile Properties.* Following the same practice employed with the grain size curves, only the results of the test bars that had equiaxed grains are plotted in Figs. 13 and 14 which

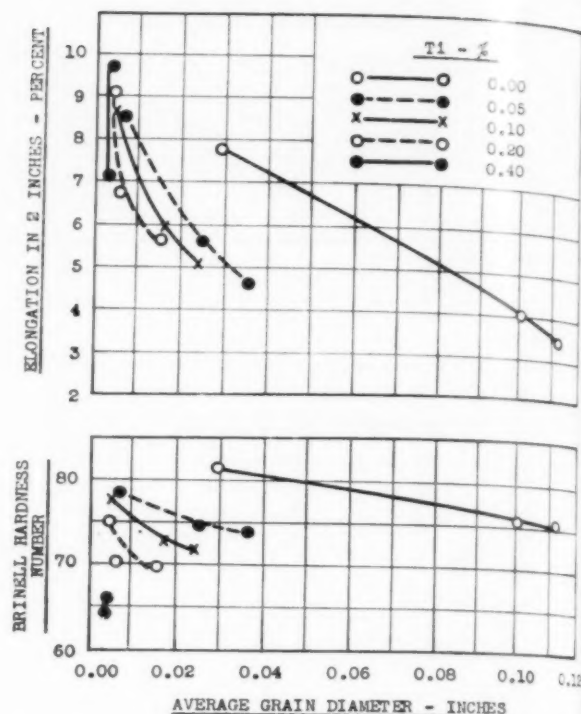


Fig. 16—Effect of grain size on the mechanical properties of No. 195 aluminum casting alloy test bars with varied titanium contents, solution heat treated and artificially aged. Coarser grain sizes were obtained by increasing the pouring temperature.

show the effect of remelting on the tensile properties of the No. 195 alloy poured at 1450 F. The tensile values are related to the grain size variations. There was no appreciable change in the grain size with remelting of the base No. 195 alloy or of the alloy containing 0.40 per cent titanium, and remelting had little effect on the tensile properties of these alloys. The tensile properties of the alloys which contained 0.05, 0.10 and 0.20 per cent titanium decreased with remelting and the grain size increased.

*Relationship of Grain Size and Tensile Properties.* Plotting the grain size against the tensile properties as in Figs. 15 and 16 differentiates the deleterious effects of the undissolved  $\text{TiAl}_3$  particles and the concentration of the insoluble  $\alpha\text{Al-Fe-Si}$  constituents in the grain boundaries from the beneficial grain refinement. It was evident from these curves that by increasing the titanium content a progressive decrease in the grain size was necessary so that the resultant tensile properties would be of the same order.

*Effect of Titanium on Microstructure.* Figures 17 through 20 show the microstructures of the No. 195 alloys with varied titanium contents which were poured at 1450 F. As the titanium content was increased, the insoluble  $\alpha\text{Al-Fe-Si}$  constituent coarsened slightly and concentrated at the grain boundaries. Metallographic examination revealed that the fracture usually followed the insoluble constituents. Hence, it is logical to conclude that the coarser and more continuous this network, other factors the same, the lower the tensile properties.



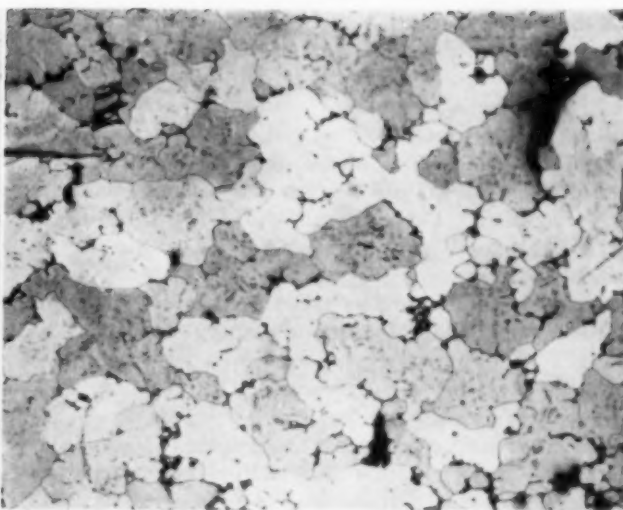
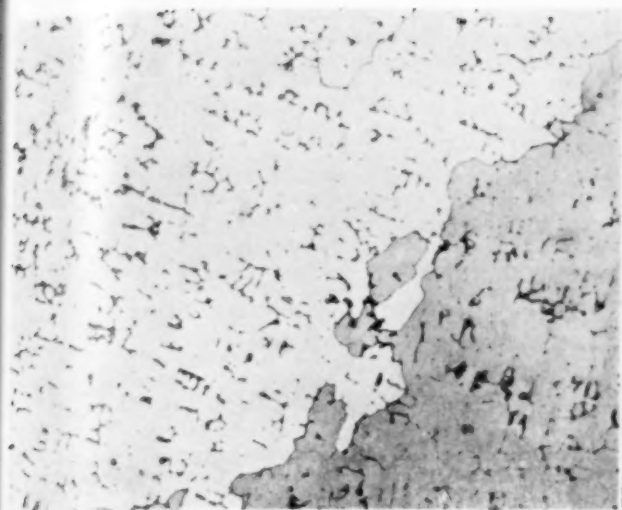


Fig. 17—Representative microstructure of No. 195 aluminum alloy containing 0.006% titanium and poured at 1450 F. Particles of  $\alpha$ Al-Fe-Si distributed throughout the grains. Keller's etch. Mag. 50X.

Fig. 18—Representative microstructure of No. 195

aluminum alloy containing 0.23% titanium and poured at 1450 F. Particles of  $\alpha$ Al-Fe-Si predominantly concentrated at the grain boundaries. A  $\text{TiAl}_3$  crystal is shown at the left above center. Keller's etch. Mag. 50X.

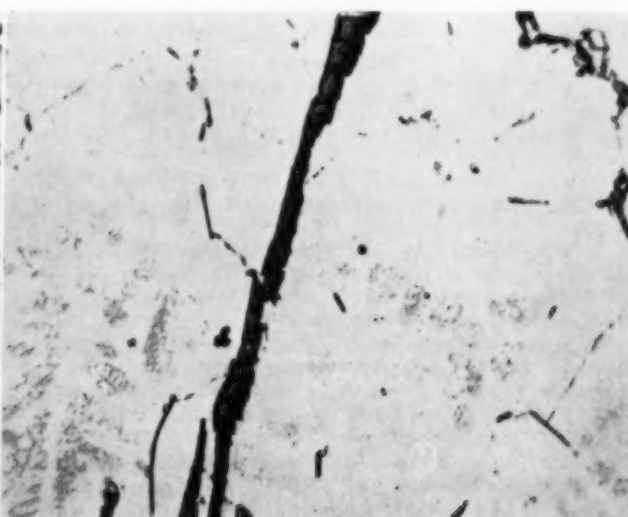
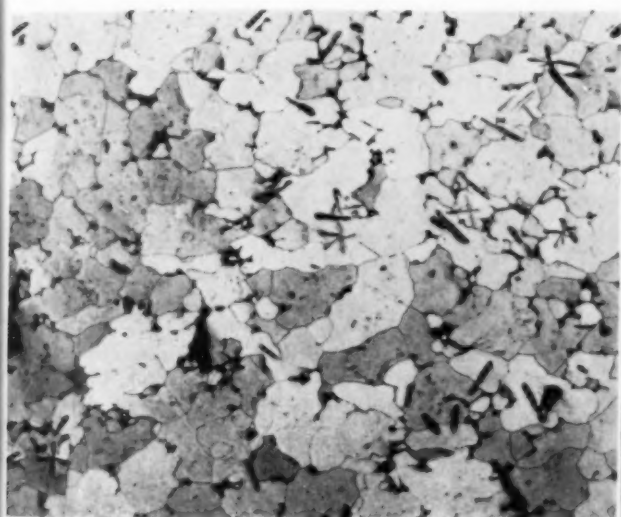


Fig. 19—Representative microstructure of No. 195 aluminum alloy containing 0.39% titanium and poured at 1450 F. Several groups of  $\text{TiAl}_3$  crystals are evident. Keller's etch. Mag. 50X.

Fig. 20—Representative microstructure of No. 195

aluminum alloy containing 0.23% titanium and poured at 1450 F. Coarse  $\text{TiAl}_3$  crystals appear in the center and groups of fine  $\text{TiAl}_3$  particles are evident in matrix. Electrolytically polished. No etch. Mag. 250X.

Crystals of  $\text{TiAl}_3$  could not be detected in the microstructure of the base alloy or in the alloys containing 0.05 or 0.10 per cent titanium. It is assumed, therefore, that the  $\text{TiAl}_3$  particles in these alloys were of submicroscopic size. A few large isolated  $\text{TiAl}_3$  crystals were observed in the alloy which contained 0.20 per cent titanium. Adjacent to the large  $\text{TiAl}_3$  crystals were groups of very small particles which occurred in a dendrite-like pattern. It was believed that these were particles of  $\text{TiAl}_3$  which in passing through the peritectic range were reduced in size but because of the rapid cooling rate did not completely dissolve in the alloy. Since these extremely

small particles of  $\text{TiAl}_3$  were observed in this alloy, it was logical to conclude that submicroscopic particles of  $\text{TiAl}_3$  would also be present. Therefore, these minute particles are probably present in the alloys which had lower titanium contents. In the alloy which contained 0.40 per cent titanium, the large  $\text{TiAl}_3$  crystals were not evenly distributed but occurred in groups.

#### Conclusions

1. This investigation has indicated that the grain refinement obtained by the addition of titanium to No. 195 aluminum alloy was effected by particles of

TiAl<sub>3</sub> which were reduced in size in passing through the peritectic range. These minute particles served as nuclei to initiate crystallization throughout the melt.

2. The grain refinement produced by the 0.05, 0.10 and 0.20 per cent titanium additions was dependent on the pouring temperature, holding time after the titanium addition, and the number of times the alloy was remelted. These factors did not influence the grain refinement produced by the 0.40 per cent titanium addition.

3. There was no detectable loss in the titanium contents of the alloys with remelting.

4. The tensile properties obtained with variations in titanium were the resultant of four factors: (a) the beneficial effect of the grain refinement, (b) the deleterious influence of the undissolved TiAl<sub>3</sub> particles, (c) the undesirable slight coarsening and concentration of the insoluble  $\alpha$ -Al-Fe-Si constituent at the grain boundaries, and (d) the effect of the retarded artificial aging which was proportional to the quantity of titanium added.

#### Acknowledgment

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#### DISCUSSION

Chairman: HIRAM BROWN, Solar Aircraft Co., Des Moines, Iowa

Co-Chairman: H. R. YOUNGRANTZ, Apex Smelting Co., Chicago

WALTER BONSAK (Written Discussion):<sup>1</sup> Congratulations to the authors for a fine piece of research work. The effect of grain refiners is a very interesting subject and the authors have systematically confirmed previous experiences. It is interesting to note that our observation of the loss of grain

refining by repeated remelting has been confirmed although no loss of titanium could be found by analysis. Also that with increasing titanium additions reduction in tensile strength occurs and that slower rate of artificial aging, reduced yield strength, and increased elongation is obtained. Furthermore, the lower the titanium addition the sooner the grain refining effect is lost on remelting.

So far the authors stated facts; from here on however, they try to prove a theory of grain refining which I would like very much to be correct because it is such a beautiful simple explanation. However from some of the facts stated and from our own observations and studies the explanation of grain refinement is not quite as simple. Now I am not the first to argue to the pro or con of the mechanism of grain refining, but since the authors opened up an old controversy and since I am forced by our studies to be on the other side, I would like to try and show my analysis.

At first, it may be quite correct to apply the equilibrium conditions of the binary Aluminum-Titanium System in a quaternary system of Aluminum-Copper-Iron-Silicon. Furthermore, in casting, as pointed out by the authors, equilibrium conditions are seldom obtained and especially peritectic reactions can be very slow and more so when the reaction temperature range is only 5 C. In the micrographs presented, peritectic reaction can not be seen on the Aluminum-Titanium needles. We have never observed it in a casting containing titanium. The fine TiAl<sub>3</sub> particles found in the matrix in Fig. 20 are still of relatively large size since the magnification is only 250 times, and for nucleation the size should be of nuclear or at least of molecular dimension, in other words, invisible at 250 diameters.

The other theory to which I am more and more forced to lean is that the effect of grain refining is caused by a non-metallic matter which is carried in by the titanium hardener or grain refiner and is present in the form of a mist or cloud. This may be a reaction product between aluminum and TiO<sub>2</sub> and aluminate or a titanate. This mist or cloud floats highly dispersed throughout the melt. This cloud prevents the formation of crystals during the solidification range. Gwyer and Phillips\* in 1926 referred to such crystallization interference as similar to the action of a colloid in manufacture of ice cream. Without the addition of such colloids, ice cream will be coarse grained and watery; with a colloid, fine grained, creamy and smooth.

Now then, for arguments sake let us suppose that in aluminum this colloid is the cloud of titanate or aluminate. The cloud matter is not soluble in the aluminum and is pushed ahead of the growing crystal until a skin is formed heavy enough to prevent further grain growth. It will prevent the growth of large grain of the main constituent by forming envelopes around the grain. The more cloud matter available the sooner envelopes are formed and grain growth stopped. These envelopes will in turn prevent constituents of grain boundaries to go into solid solution during solidification and heat treatment. The more cloud matter the greater the interference in solid solution. Therefore the more titanium the lower the tensile strength, hardness (Brinell) and yield strength due to lack of dissolved copper, etc., the higher the elongation due to the annealing effect of an unsatisfied solid solution. During aging less copper is available for precipitation.

To provide further proof of the plausibility for the above theory: A fine grained alloy containing 4 per cent copper, 3 per cent silicon and 0.12 per cent titanium was remelted four times. The metal was heated to 1500 F and agitated for 10 minutes with a gentle stream of N<sub>2</sub>. The skim formed during remelting was drawn off and the metal poured into ingots. The ingots were sectioned and the grain size was determined. The grain size increased with each remelt considerably and finally became exceedingly coarse. The skim from each melt was analyzed by x-ray diffraction and it was found that the first skim contained among others K<sub>2</sub>TiO<sub>3</sub> and TiO<sub>2</sub> and as the skim became freer of these compounds the alloy became coarser. These tests were repeated several times with the same results. In all these tests the basic titanium analysis did only change within the possible analytical error or was practically constant.

\* "The Constitution and Structure of the Commercial Aluminum-Silicon Alloys", A. G. C. Gwyer and H. W. L. Phillips, *The Journal of the Institute of Metals* (British, Vol. 36, No. 2, pp. 291-3 (1926).

<sup>1</sup> Apex Smelting Co., Cleveland

The facts found by these tests and the other effects shown by the authors and explained above give good reasons to lean more toward the cloud interference theory than to the nucleation theory.

MR. BOEHM: The applicability of the binary Al-Ti constitutional diagram in connection with this investigation may be slightly open to criticism. However, the phase field limits of this diagram are substantially the same as in the ternary Al-Cu-Ti and Al-Fe-Ti systems at the aluminum end with respect to phases containing titanium. This observation supports the opinion that an equilibrium diagram for 195 alloy would be similar to the Al-Ti diagram as far as titanium containing constituents are concerned.

We do not believe that the particles shown in the photomicrograph at 250X are the ones that initiate grain refinement. The important point is that since some  $TiAl_3$  particles have been reduced in size, considerably smaller particles of nucleating size probably are also present.

I am not prepared at this time to discuss in detail the grain refinement theory advocated by Mr. Bonsack. However, the explanation given in the paper is adequately supported by the data obtained.

HAROLD BERNSTEIN:<sup>2</sup> We ran some class 195 melts recently and added the grain refiner as copper-titanium master alloy instead of a titanium-aluminum alloy, and we obtained substantially the same refinement as with the aluminum master alloy. That would indicate that if it is the  $TiAl_3$  and subsequent modification that is governing the grain refinement, then  $TiAl_3$  must be formed when the copper master alloy is added to the melt, which indicates a possible differentiation in the mechanism from what the authors have proposed in this paper.

R. A. QUADT:<sup>3</sup> We have been intrigued by this transitory effect of titanium with regard to grain refinement. Some years ago we ran a series of investigations, attempting to prove what the authors have done here. Our method of attack was a little different. We made up an alloy of 4 per cent copper with about 0.15 per cent titanium, and held it for a long period of time at about 1400 F, pouring small castings under controlled conditions at various intervals, on which we measured the decrease in grain refinement. Finally we assumed we had none because we got the same coarse grain as we had before we added the titanium. Then we oxidized the aluminum to the point where we had increased the titanium content to about 0.25 per cent. That, of course, also increased the copper content, and then we made a similar series of casts and were not able to get coarse grains any more, no matter how long we held the material at 1400 F.

We assumed, of course, that we had moved into a new field.  $TiAl_3$  plus liquid, at 1400 F. With this new alloy we continued to get fine-grained castings no matter how many times we remelted or how long we held it at 1400 F. We then added pure aluminum to the melt, diluted the titanium back to 0.17 per cent again, and by reproducing the melting and holding techniques we were able to eliminate effectively our grain refinement again. What we were attempting to prove, of course, was that we could make the grain refinement come and go without adding more titanium but just by using the same titanium and changing its form in the melt, and we came to the conclusion that  $TiAl_3$  was a very important factor.

W. E. SICHA: This is a very interesting experiment because it

definitely supports the views expressed and the conclusions reached in this paper. These test results are a valuable contribution.

W. E. SICHA and R. C. BOEHM (*Authors' closure*): With reference to Mr. Bonsack's written discussion, available data sustain the suitability of using the binary Al-Ti equilibrium diagram for illustrative purposes as was done in presenting the paper. References 2 and 3 of the paper cite the development of ternary diagrams for the system Al-Cu-Ti and Al-Fe-Ti which were reproduced by Mondolfo.<sup>11</sup> These constitutional diagrams disclose practically no deviation in the former and only a small variation in the latter from that of the Al-Ti system with respect to the titanium content represented by the left terminus of the peritectic horizontal. In view of this evidence, it appears that the binary diagram probably is reasonably representative of the more complex commercial alloy 195 with titanium additions.

It was implied by Mr. Bonsack that a peritectic reaction does not occur in the alloy 195-titanium system. We believe there is ample evidence in the careful work of several investigators<sup>1,2,3,11</sup> on related titanium containing alloys to substantiate our contention that a peritectic transformation takes place in this commercial alloy system. The fact that the solid solubility of titanium is higher than the solubility in liquid aluminum<sup>1</sup> is very good evidence.

The object in presenting Fig. 20 was to show fine as well as coarse  $TiAl_3$  particles which indicates the excellent probability of the presence of considerably finer particles, unresolved at 250X, necessary for nucleation. Grain refinement imparted by 0.05 and 0.10 per cent Ti additions, which did not provide  $TiAl_3$  particles visible in the microstructure, also serves to demonstrate that the particles initiating crystallization are of submicroscopic size.

One fundamental concept of the cloud interference theory used by Mr. Bonsack to explain reported variations in tensile properties with different titanium contents is wholly compatible with this theory but in direct disagreement with experimental observations. If cloud envelopes exist they certainly should interfere with solution of copper during heat treatment. However, there was no undissolved  $CuAl_2$  in the microstructures subsequent to solution heat treatment regardless of titanium content.

The experiments described by Mr. Bonsack demonstrate grain coarsening produced by remelting an Al-Cu-Si alloy containing 0.12 per cent Ti. This effect of remelting is wholly consistent with the nucleation theory. Also, the potassium titanate and titanium oxide detected in the skimmings probably were introduced with the titanium-bearing material employed in making the titanium addition. A finite quantity of this non-metallic material probably was introduced originally. The amount available to separate from the melt and accumulate in the skimmings in subsequent remelting operations would become progressively smaller. This change of content of these oxides in the skimmings could be completely independent of association with grain size alteration resulting from remelting.

Perhaps the best support for the nucleation theory as an explanation for the effect of titanium as a grain refiner for aluminum alloys is the fact that all experimental data, with which the authors are familiar, in the paper as well as that published by other investigators, are in accord with this theory.

<sup>1</sup> U. S. Naval Gun Factory, Washington, D. C.

<sup>2</sup> Federated Metals Div., American Smelting & Refining Co., Perth Amboy, N. J.



# PEARLITIC MALLEABLE IRONS, PLAIN AND ALLOYED

By

Richard Schneidewind\* and D. J. Reese\*\*

## ABSTRACT

The authors have investigated the mechanical properties of ferritic and pearlitic malleable irons, plain and alloyed. Particular emphasis was placed upon the relationship between design strength and toughness as measured by the yield point and the elongation respectively. It was found that this relationship might vary widely but that with proper heat treatment the yield points bore a fixed relationship with elongation. This can be expressed mathematically by the equation

$$Y + 50 \log E = b$$

where  $Y$  is the yield point in 1000 psi,  $E$  is the per cent elongation, and  $b$  is a constant.

This constant equals 95 when the properties meet the A.S.T.M. or Army Ordnance specifications. The average value for the constant obtained in this investigation was 102 which means that when properly heat treated a pearlitic iron should have 7000 psi greater yield point than demanded by specification for any given elongation.

Best results were obtained when the tempering temperature was kept low, about 1200 F., in order that the greatest amount of stress relief and spheroidization might take place with the

minimum amount of second stage graphitization. The temperatures from which the iron was originally either air- or oil-quenched determined the maximum strength obtainable by the heat treatment.

The use of alloys coupled with proper heat treatment made possible the securing of 5000 to 10,000 psi greater yield point for a given elongation than was possible with unalloyed iron. The constant in the above equation averaged 108 for alloyed irons with some values as high as 115. This improvement was also noted in the case of fully annealed alloyed irons.

Although oil-quenched irons showed higher maximum yield points than air-quenched irons, after proper heat treatment the relationships between yield point and per cent elongation were the same. Oil-quenched irons of a given yield strength usually have lower tensile strengths than air-quenched; in other words oil-quenched irons have a higher yield to tensile strength ratio.

Where irons are alloyed to permit the production of pearlitic malleable in a cycle designed to produce annealed malleable with plain irons, care must be exercised to use the proper alloys in the proper amounts to the particular iron and cycle. There is danger in stabilizing massive carbides with some alloys and of producing primary graphite by inoculation with others.

## Influence of Microstructure on Mechanical Properties

THE MECHANICAL PROPERTIES of pearlitic malleable irons are dependent upon the following variables:

- (a) The amount, shape, and distribution of the temper carbon particles.
- (b) The strength, ductility, and toughness of the ferrite, which in this case is alloyed with silicon and other elements.
- (c) The amount, shape, and distribution of the carbides.

These factors in turn are affected by the nature of the base metal and the manner in which the iron is heat treated. For example, when white iron is malleabilized, the temper carbon pattern is established as a result of composition, soaking temperature, rate of heating, and other similar conditions. This pattern will not change materially upon any subsequent heat treatment

so that variable, (a), dealing with temper carbon may be considered fixed, once first stage graphitization is completed. Primary graphite is, of course, harmful and should not be present.

The nature of the ferrite phase, whether free or part of the pearlite, is fixed by the composition except for possible dispersion hardening effects. Hence, item (b) dealing with ferrite may be considered constant for any given iron.

Item (c), dealing with carbides, is controllable by heat treatment. In pearlitic malleable irons, free massive or eutectic carbides have a harmful effect upon ductility and toughness. If present in large quantities carbides of this type can also result in low strength as well. The degree of spheroidization of eutectoid carbides is a controlling factor in the properties of pearlitic irons.

## Heat Treatments

Pearlitic types of malleable iron can be produced by various heat treatments; treatments which have been used commercially can be grouped as follows:

1. Arrested malleabilization; the iron may be air

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quenched from the soaking temperature after first stage graphitization is completed or from some lower temperature above the critical and then tempered.

2. Arrested malleabilization as above except that the iron is quenched in oil usually from a temperature around 1500 F and then is tempered.

3. Re-absorption treatments where completely annealed iron is reheated to some temperature above the critical temperature in order to dissolve carbon into the austenite followed by air or oil quenching and tempering.

4. Certain alloying elements exhibit little or no tendency to produce primary graphite during solidification and also little or no decelerating action on the rate of first stage graphitization. They do have, however, a markedly retarding effect on the rate of second stage graphitization. If the quantity of the alloying element is controlled, a given annealing cycle can produce a desired amount of spheroidized carbides in the matrix.

All of these methods have in common holding the metal at a temperature below the critical. This part of the treatment may accomplish the following ends:

1. Relief of quenching stresses.
2. Decomposition of martensite if quenched previously.
3. Spheroidization of eutectoid carbides.
4. Decomposition of carbides to ferrite and temper carbon.
5. Surface decarburization.

These methods differ in significant respects. The amount of combined carbon to be acted upon in the subsequent tempering treatment is controlled by the temperature from which the iron is cooled. The higher the temperature of quenching or air cooling, the greater will be the amount of combined carbon; also the austenite grain size will be greater at higher temperatures.

The rate of cooling determines the matrix; air cooling generally produces a pearlitic matrix with more or less free ferrite in the form of bulls-eyes.

If completely annealed iron is reheated to above the critical and air cooled, the distribution of the ferrite and pearlite is distinctly different than if air cooled after an arrested malleabilization. All these differences in microstructure are accompanied by differences in mechanical properties.

#### Review of Published Data

The results of arrested malleabilization have been

well discussed by Joseph.<sup>1</sup> The composition of the base irons is given as follows:

Element	Per Cent
Carbon	2.35-2.75
Silicon	1.20-1.70
Manganese	0.35-0.50
Sulphur	0.17 max.
Phosphorus	0.10 max.

It is believed that the largest share of the irons described contained carbon at the high end and silicon at the low end of the ranges above.

The white iron was soaked at 1700 F to complete first stage annealing and then air quenched. By tempering for a suitable length of time at 1250 F a wide variety of mechanical properties were obtained. Table 1 presents the properties of irons so treated computed from the data. Figure 1 shows the relationship

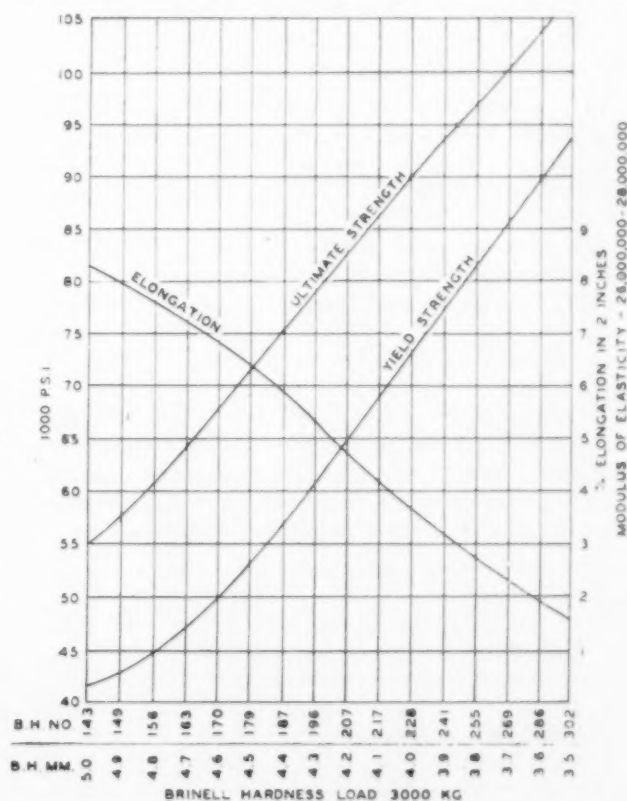


Fig. 1—Properties of pearlitic malleable iron (C. F. Joseph)

TABLE 1—PROPERTIES OF PEARLITIC IRON (C. F. Joseph)

Time of Tempering at 1250 F, hr	Yield Point, psi	Tensile Strength, psi	Elongation, Per Cent	Hardness Brinell	T.S. B.H.	Y.P. B.H.	Y.P. T.S.	b
0	90,000	104,000	1.9	286	364	314	0.865	103.95
2	83,000	97,000	2.7	255	380	325	0.856	104.60
4	71,000	88,000	4.0	220	400	322	0.808	101.10
5	67,000	84,000	4.4	212	396	316	0.800	99.20
6	61,000	79,000	5.3	196	402	311	0.774	97.20
8	53,000	72,000	6.3	179	402	296	0.736	93.00
10	47,500	64,500	7.2	163	396	291	0.736	90.50
12	41,000	54,000	8.4	136	397	302	0.760	87.20

between the various physical properties as compared with the Brinell Hardness. The microstructures of these irons consisted of temper carbon in a pearlitic matrix. Where tempering had been carried out for 4 hr or more at 1250 F, an increasing amount of ferrite was produced in bulls-eye arrangement. After 12 hr the matrix contained about 75 per cent ferrite.

Schneidewind and White<sup>2</sup> studied irons produced at three different foundries. The specified heat treatments were carried on at a commercial heat treating establishment. The compositions were as follows:

Element	Content, Per Cent		
	I	II	III
Carbon .....	2.35	2.33	2.30
Silicon .....	1.42	1.50	1.70
Manganese .....	0.23	0.28	0.50
Sulphur .....	0.06	0.03	0.05
Phosphorus .....	0.05	0.032	0.05

The low sulphurs were a result of electric furnace melting of selected scrap. Although the properties of all three sets of iron would pass the present A.S.T.M. specifications, the irons from Foundry I were not as good as those of the other two.

A fourth set was prepared by Foundry III and contained 0.5 per cent molybdenum.

The heat treatment consisted in heating at 1700 F for 15 hr, dropping to 1525 F in 2 hr, and oil quenching. Tempering was done at 1325 F for 2 and 4 hr respectively.

The properties are presented in Table 2. Each value presented is the average of six determinations.

Forbes<sup>3</sup> studied the effect of manganese, an alloy which appreciably retards second stage graphitization. The composition of his irons were as follows:

Element	Per Cent
Carbon .....	2.43
Silicon .....	1.03
Manganese .....	0.34-0.71-1.02
Sulphur .....	0.08
Phosphorus .....	0.15

The heat treatments, somewhat lengthy, were used to bring out the optimum properties. They were as follows:

Treatment A	1. Heat to 1720 F in 20 hr
	2. Heat at 1720 F for 24 hr
	3. Air cool to room temperature in 1.5 hr
	4. Reheat at 1280 F for 30 hr
Treatment B	1, 2, and 3 as in A
	4. Reheat at 1600 F for 20 hr
	5. Air cool
	6. Reheat at 1280 F for 30 hr
Treatment C	1, 2, and 3 as in A
	4. Reheat at 1500 F for 20 hr
	5. Air cool
	6. Reheat at 1280 F for 30 hr

Treatment D	1, 2, and 3 as in A
	4. Reheat at 1400 F for 20 hr
	5. Air cool
	6. Reheat at 1280 F for 30 hr

Properties which resulted are presented in Table 2.

For any one iron treated by treatments, B, C, and D, the properties differ from that of treatment A in the amount of combined carbon present prior to tempering at 1280 F and in austenitic grain size. Examination of the data will show that these factors were relatively unimportant to the final results so that in later discussions the averaged values for each manganese classification will be used. The manganese altered the stability of the carbides and hence the quantity remaining after each heat treatment.

### Experimental Work

One of the present writers made three preliminary tests for the purpose of studying the effect of alloys on irons which were given annealing treatments normally used to produce fully annealed malleable iron with unalloyed material. Three separate heats were prepared at one foundry where melting conditions could be carefully controlled. Each heat was then annealed by three different foundries; four foundries collaborated in the three tests.

### Annealing Cycles

The annealing cycles used at three of the four foundries were as follows. No data were submitted by Foundry D regarding the heat treatment cycle.

#### Foundry A

From room temperature to 1750 F in 6.3 to 7.2 hr, average rate of heating, 4.5 to 3.9 F per min.

Held at 1750 F for 10.5 to 12.0 hr.

Cooled to 1325 F in 2.1 to 2.4 hr, average rate of cooling was 3.4 to 3.0 F per min.

TABLE 2—PROPERTIES OF IRONS (Schneidewind and White)

Foundry	Y.P. psi	T.S. psi	Elongation, percent	Y.P. T.S.	b
Plain Irons					
1700 F — 1525 F — oil quench — 1325 F (2 hr)					
III	60,300	76,510	6.5	0.793	100.5
I	66,620	82,660	5.6	0.805	104.2
II <sup>1</sup>	60,180	79,230	5.9	0.760	98.7
1700 F — 1525 F — oil quench — 1325 F (4 hr)					
III	50,042	64,442	8.4	0.778	96.1
I	52,730	68,842	6.7	0.767	94.1
II <sup>1</sup>	47,225	63,150	6.5	0.749	87.4
1700 F — 1525 F — oil quench — 1325 F (1 hr)					
I	63,917	85,550	3.75	0.747	92.6
1800 F — 1525 F — oil quench — 1325 F (8½ hr)					
I	45,350	58,900	10.5	0.770	96.5
0.5 percent Molybdenum Irons					
1700 F — 1525 F — oil quench — 1325 F (2 hr)					
III	70,567	77,783	3.5	0.909	97.8
1700 F — 1525 F — oil quench — 1325 F (4 hr)					
III	64,962	73,962	7.9	0.878	109.9

<sup>1</sup>Irons from this foundry were not good; took longer to anneal, had pinholes, rough surface, etc.



Cooled from 1325 to 1290 F in 8.4 to 9.6 hr—average rate of cooling was 0.07 to 0.06 F per min.  
Total time from room temperature to removal from furnace at 1290 F was 27.3 to 31.2 hr.

TABLE 3—PROPERTIES OF IRONS (Forbes)

Heat Treatment	Y.P. psi	T.S. psi	Elongation, per cent	Y.P. T.S.	b
0.34 Per Cent Manganese					
A	39,590	56,670	16.5	0.697	99.5
B	36,940	55,900	16.0	0.661	97.1
C	38,900	56,710	21.3	0.686	105.4
D	40,410	58,930	20.0	0.686	105.5
Aver.	38,960	57,052	18.45	0.683	101.9
0.71 Per Cent Manganese					
A	48,470	69,290	16.5	0.698	109.4
B	47,120	80,880	15.5	0.666	106.6
C	46,040	67,130	14.7	0.685	104.4
D	47,390	67,800	16.0	0.698	107.6
Aver.	47,550	68,788	15.65	0.694	107.0
1.02 Per Cent Manganese					
A	57,290	85,990	10.5	0.668	108.4
B	60,310	87,150	13.0	0.693	116.0
C	58,090	83,090	11.5	0.699	111.1
D	57,330	83,520	12.3	0.685	112.2
Aver.	58,255	85,063	11.82	0.685	111.9

## Foundry B

From room temperature to 1550 F in 36 hr, average rate of heating was 0.69 F per min.

Held at 1550 F—55 hr.

Cooled from 1550 to 1325 F in 23 hr, average rate of cooling was 0.16 F per min.

Cooled from 1325 to 1270 F in 20 hr, average rate of cooling was 0.046 per min.

Cooled from 1270 to 1050 F in 41 hr, average rate of cooling was 0.09 F per hr.

Total time from room temperature to removal from furnace at 1050 F is 175 hr.

## Foundry C

From room temperature to 1650 to 1700 F in 30 hr, average rate of heating was 0.89 F per min.

Held at 1650 to 1700 F for 30 hr.

Cooled from 1650—1700 F to 1320 F in 18 hr, average rate of cooling was 0.328 F, per min.

Cooled from 1320 F to 1250 F in 30 hr, average rate of cooling was 0.039 F per hr.

Total time from room temperature to removal from furnace to 1250 F is 108 hr.

## Properties

The properties of the irons and their compositions are presented in Tables 4, 5, and 6. Foundry condi-

TABLE 4—PROPERTIES OF MALLEABLE IRONS

Bar No.									Yield Strength, psi			Tensile Strength, psi			Per Cent Elongation		
	Cu	Ni	Mo	C	Si	Mn	S	P	C	D	A	C	D	A	C	D	A
1	Base iron			2.36	1.34	0.41	0.120	0.084	36260	37100	35050	53740	52400	53100	20.0	20.0	14.84
2	...	...	0.15	...	...	...	...	...	37220	38100	39000	56150	54800	56600	21.0	19.5	17.97
3	...	...	0.30	...	...	...	...	...	38500	39100	38700	57440	56100	57200	23.0	21.5	21.09
4	...	0.5	...	...	...	...	...	...	38860	39800	34200	54400	54700	54800	17.0	16.5	16.41
5	...	1.0	...	...	...	...	...	...	39790	40000	40000	50520	55300	56000	...	16.0	11.72
6	...	1.5	...	...	...	...	...	...	34350	33600	...	...	33600	36900	...	...	...
7	...	2.0	...	...	...	...	...	...	13990	24040	...	...	24050	16500	...	...	...
8	...	0.5	0.15	...	...	...	...	...	39480	38100	41600	56440	56300	57400	20.0	15.0	20.31
9	...	1.0	0.15	...	...	...	...	...	39790	43400	42100	57480	58000	58250	19.0	16.0	14.06
10	...	1.5	0.15	...	...	...	...	...	41210	46200	52000	49740	57400	69250	...	11.5	...
11	...	2.0	0.15	...	...	...	...	...	47370	29500	56300	58420	29500	72000	9.5	0.0 <sup>a</sup>	...
12	...	0.5	0.3	...	...	...	...	...	40740	43250	36600	57480	59000	60400	17.0	22.0	17.19
13	...	1.0	0.3	...	...	...	...	...	45160	43000	39400	60950	48900	66800	16.5	10.5	...
14	...	1.5	0.3	...	...	...	...	...	37270	38500	39800	...	38500	44700	...	...	...
15	...	2.0	0.3	...	...	...	...	...	...	17900	58200	50530	17900	59800	...	...	...
16	0.5	...	...	...	...	...	...	...	43700	39500	34100	57400	55600	56200	19.0	20.0	19.53
17	1.0	...	...	...	...	...	...	...	43900	43500	41500	57630	59000	64000	16.5	16.5	14.06
18	1.5	...	...	...	...	...	...	...	44200	43250	47500	56840	56800	61250	13.5	14.0	12.50
19	0.5	...	0.15	...	...	...	...	...	45000	40400	37800	58580	56750	56200	19.5	17.5	14.06
20	1.0	...	0.15	...	...	...	...	...	43900	47200	49150	60630	61000	69100	17.0	12.5	7.81
21	1.5	...	0.15	...	...	...	...	...	43900	43900	47200	58260	60400	68200	15.0	15.0	7.81
22	0.5	...	0.3	...	...	...	...	...	44050	42300	38800	59840	59500	61500	18.0	17.0	14.84
23	1.0	...	0.3	...	...	...	...	...	44530	47700	45800	60320	63950	69900	15.5	16.5	9.37
24	1.5	...	0.3	...	...	...	...	...	44220	47900	42400	62060	62600	65500	18.0	15.0	10.16
25	0.5	0.5	...	...	...	...	...	...	43900	44950	39000	53680	58000	59700	12.0	18.0	15.63
26	1.0	1.0	...	...	...	...	...	...	45800	47500	50600	53050	55050	61450	5.0	7.0	2.34
27	0.5	0.5	0.15	...	...	...	...	...	44530	44150	45650	57800	59400	62500	18.5	17.5	13.28
28	1.0	1.0	0.15	...	...	...	...	...	47370	52100	48200	59680	61700	62600	...	...	...
29	0.5	0.5	0.3	...	...	...	...	...	47050	44800	48750	60000	61200	67350	15.0	15.5	14.06
30	1.0	1.0	0.3	...	...	...	...	...	50520	52500	46000	...	...	62100	...	...	...
31	Base iron			2.27	1.36	0.42	0.121	0.084	37900	39100	36900	53370	54400	54700	18.5 <sup>aa</sup>	17.5	21.09

<sup>a</sup>Broken out of gauge.

<sup>b</sup>Surface check.

TABLE 5—PROPERTIES OF MALLEABLE IRONS

Bar No.	Cu	Ni	Mo	C	Si	Mn	S	P	Yield Strength, psi			Tensile Strength, psi			Per Cent Elongation		
									C	B	A	C	B	A	C	B	A
1	Base iron			2.31	1.29	0.37	....	....	38600	39000	37400	58740	56000	56600	17.0	22.0	24.2
4	...	0.5	...	....	....	....	....	....	39800	42500	40200	59700	59700	57500	17.5	22.5	17.4
5	...	1.0	...	....	....	....	....	....	45480	44700	44600	63300	60000	60700	21.5	17.5	14.8
6	...	1.5	...	....	....	....	....	....	48000	45400	47800	63480	62000	64300	20.0	17.5	7.8
7	...	2.0	...	....	....	....	....	....	52100 <sup>a</sup>	49800	67600	62700 <sup>a</sup>	64700	90700	7.5 <sup>a</sup>	13.5	4.7
13	...	1.0	0.3	....	....	....	....	....	47370	47500	47000	64740	63700	71100	15.0	21.0	14.8
14	...	1.5	0.3	....	....	....	....	....	52100	51600	52700	69500	67700	64400	14.0	18.5	2.9
15	...	2.0	0.3	....	....	....	....	....	55260 <sup>a</sup>	52800	63600	69500 <sup>a</sup>	67900	74500	11.5 <sup>a</sup>	16.5	3.9
26	1.0	1.0	...	....	....	....	....	....	54950	51900	75300	67400	66000	94200	12.0	13.5	2.9
30	1.0	1.0	0.3	....	....	....	....	....	60000	57000	54000	75800	71600	69900	12.5	12.5	3.1

<sup>a</sup>Defective Bar.

<sup>a</sup>Defective Bar.

TABLE 6—PROPERTIES OF MALLEABLE IRONS

Bar No.	Cu	Ni	Mo	C	Si	Mn	S	P	Yield Strength, psi			Tensile Strength, psi			Per Cent Elongation		
									Set No. 1		Set No. 2	Set No. 1		Set No. 2	Set No. 1		Set No. 2
									A	B	A	A	B	A	A	B	A
1	Base iron			2.25	1.30	0.41	0.105	0.11	37400	35000	36000	54000	53000	56000	21.1	17.5	24.2
4	....	0.67	....	....	....	....	....	....	37100	40200	37100	60000	55000	59000	15.6	12.0	20.3
5	....	1.32	....	....	....	....	....	....	42500	43800	49750	65850	58500	63500	10.2	16.5	18.0
6	....	1.95	....	....	....	....	....	....	40300	46100	51500	77900	60500	74000	6.3	19.0	9.4
12	....	0.69	0.37	....	....	....	....	....	45500	42600	45000	67000	61000	64750	10.9	17.5	15.6
13	....	1.35	0.40	....	....	....	....	....	46300	46300	44100	84500	64900	82000	7.0	14.5	10.2
14	....	2.10	0.40	2.20	1.27	....	....	....	56750	51000	51000	110400	67500	110000	7.0	21.0	3.1
16	0.79	....	....	....	....	....	....	....	44750	38000	41700	61500	56500	61150	18.0	19.0	17.2
17	1.35	....	....	....	....	....	....	....	49250	42000	44300	76000	58600	67500	7.0	20.0	11.7
18	1.97	....	....	....	....	....	....	....	45000	42700	47600	65750	59000	66800	12.5	20.0	9.4
22	0.73	....	0.34	....	....	....	....	....	46000	43600	46250	66500	60200	64000	13.3	19.0	13.3
23	1.30	....	0.35	....	....	....	....	....	52500	44400	53300	94750	61500	73350	7.0	19.0	7.8
24	1.90	....	0.38	2.19	1.28	....	....	....	52500	45100	52500	78000	62800	75000	5.5	20.0	6.3
25	0.74	0.68	....	....	....	....	....	....	56000	48600	54900	68000	61000	80900	10.9	21.5	8.6
26	1.52	1.30	....	....	....	....	....	....	53000	50200	55300	75000	63000	87300	9.4	18.0	3.9
29	0.64	0.61	0.33	....	....	....	....	....	44350	42300	74700	71500	62600	66000	9.4	20.5	....
30	1.24	1.21	0.35	2.20	1.28	....	....	....	56500	51000	62600	100900	67500	105000	3.7	20.5	1.6

TABLE 7—ALLOY ADDITIONS

Pour No.	Ni	Cu	Mo
1	...	...	...
2	0.5	...	...
3	1.0	...	...
4	1.5	...	...
5	0.5	0.5	...
6	1.0	1.0	...
7	0.5	...	0.3
8	1.0	...	0.3
9	1.5	...	0.3
10	0.5	0.5	0.3
11	1.0	1.0	0.3
12	...	0.5	...
13	...	1.0	...
14	...	1.5	...
15	...	0.5	0.3
16	...	1.0	0.3
17	...	1.5	0.3

tions during the pouring of irons of the first test (Table 4) were not so well controlled as in the other two heats and the results naturally are of less value in this study.

It was felt that more systematic study should be made to determine the potentialities and the behavior of these irons. For this reason a fourth heat was poured and foundry data were recorded.

The metal issued from the air furnace at between 2920 and 2960 F, the first reading having been made with a "Pyro" and the second with an L. and N. optical pyrometer. The molten metal was poured from the ladles into the molds at temperatures ranging from 2810 to 2740 F. The entire tests involving the 18 different pours took 45 min. It was felt that this speed would insure against differences in superheat, composition and other foundry conditions.

The basic composition of this metal was taken as the average of the analyses at 1:00 and 1:30 P.M., as follows:

C	.....2.3	per cent
Si	.....1.38	per cent
Mn	.....0.46	per cent
S	.....0.136	per cent

A fifth heat was made several months later and standard bars were poured. Foundry conditions were the same as above and the analysis for carbon and silicon are included.

Time	Carbon, Per Cent	Silicon, Per Cent	Manganese, Per Cent
3:00 P.M.	2.32	1.39	...
3:30 P.M.	2.27	1.38	0.47

A rather high manganese content was used at this time because of the deterioration of the coke quality and the resultant higher sulphur content of the iron. The amount of manganese was used simply to counteract sulphur.

The compositions with respect to alloy additions are shown in Table 7.

Some specimens of these irons were left in the white condition while the remainder were given an annealing treatment by Foundry A.

TABLE 8—PROPERTIES OF PLAIN IRON NO. 1

Hr at Temp.	Y.P.	T.S.	Elong.	B.H.	T.S. B.H.	Y.P. B.H.	Y.P. T.S.	b
Annealed, reheated 2 hr at 1500 F, air cooled, tempered at 1200 F								
0	56800	84000	5.67	195	431	291	0.678	94.5
1	56250	77500	6.75	192	404	292	0.725	97.8
2	53600	80500	9.13	183	440	293	0.666	101.6
4	52100	78350	10.75	176	445	296	0.665	103.7
8	52425	76475	12.50	170	443	308	0.687	107.3
16	46100	68600	14.50	156	440	296	0.674	104.2
24	44133	60000	17.08	152	395	291	0.735	105.7
35	43600	58900	18.00	145	406	300	0.740	106.3
Annealed, reheated 2 hr at 1600 F, air cooled, tempered at 1200 F								
2	60500	89150	6.0	202	441	300	0.678	99.4
4	58760	89625	7.75	202	444	291	0.655	102.3
8	57250	88500	8.75	205	432	280	0.648	104.3
16	58350	83900	8.5	204	412	286	0.695	104.9
24	54300	79200	11.0	179	442	304	0.689	106.4
32	45550	66575	15.75	154	433	296	0.685	105.4
Annealed, reheated 2 hr at 1500 F, oil quenched, tempered at 1200 F								
4	73700	81600	4.25	213	383	346	0.904	102.4
8	68575	79450	5.0	195	407	352	0.865	103.5
16	54475	69150	9.5	178	388	308	0.787	103.4
White, heated 17 hr at 1700 F, furnace cooled to 1500 F, held 2 hr, furnace cooled to 1200 F and held at 1200 F.								
8	47400	61800	7.5	147	422	322	0.765	91.1
16	45800	62700	12.0	145	432	316	0.732	99.8
White, heated 7 hr at 1700 F, furnace cooled to 1500 F, held 2 hr, air cooled, tempered at 1200 F.								
8	64000	83200	4.0	190	438	337	0.770	94.0
White, heated 17 hr at 1700 F, furnace cooled to 1500 F, held 2 hr, air cooled, tempered at 1200 F.								
4	68000	93000	5.0	216	431	317	0.731	103.5
8	62800	85950	6.75	197	436	319	0.733	104.3
16	47800	65200	13.0	157	415	305	0.733	103.5
White, heated at 1700 F for 6 hr, furnace cooled to 1355 F and held.								
2	52000	58000	3.0	192	302	271	0.895	75.9

TABLE 9—PROPERTIES OF NO. 2 IRON CONTAINING 0.5% Ni

Hr at Temp.	Y.P.	T.S.	Elong.	B.H.	T.S. B.H.	Y.P. B.H.	Y.P. T.S.	b
Annealed, reheated 2 hr at 1500 F, air cooled, tempered at 1200 F								
0	66700	84900	4.0	200	425	334	0.785	96.8
4	54400	74200	5.0	165	449	330	0.734	89.3
8	50200	61050	4.75	150	406	335	0.823	84.0
16	39200	52000	9.0	130	400	302	0.753	86.9
White, heated 17 hr at 1700 F, furnace cooled to 1500 F, held 2 hr, air cooled, tempered at 1200 F								
4	56500	68750	4.5	175	393	327	0.822	89.2
8	52600	68300	6.5	169	404	312	0.770	92.8



TABLE 10—PROPERTIES OF NO. 3 IRON CONTAINING 1.0% NI

Hr at Temp.	Y.P.	T.S.	Elong.	B.H.	T.S.	Y.P.	Y.P.
					B.H.	B.H.	T.S.
Annealed, reheated 2 hr at 1500 F, air cooled, tempered at 1200 F							
0	71100	89000	3.0	216	412	329	0.800
4	54750	68400	4.75	170	403	323	0.803
8	45550	60825	7.25	144	422	317	0.750
16	48600	63750	8.5	145	439	335	0.764
24	36000	42000 <sup>a</sup>	5.5 <sup>a</sup>	130	...	277	.....
White, heated 17 hr at 1700 F, furnace cooled to 1500 F, held 2 hr, air cooled, tempered at 1200 F							
4	57700	69900	4.5	175	400	329	0.829

TABLE 11—PROPERTIES OF NO. 5 IRON CONTAINING 0.5% NI AND 0.5% Cu

Hr at Temp.	Y.P.	T.S.	Elong.	B.H.	T.S. B.H.	Y.P. B.H.	Y.P. T.S.
Annealed, reheated 2 hr at 1500 F, air cooled, tempered at 1200 F							
0	68533	88700	2.83	212	418	324	0.774
2	65550	86800	4.25	200	434	328	0.744
4	60633	79450	5.0	180	441	336	0.764
8	56700	74767	6.33	169	442	336	0.759
16	47850	65125	9.13	148	440	324	0.735
24	42000	54000	12.5	132	410	318	0.778
Annealed, reheated 2 hr at 1445 F, air cooled, tempered at 1200 F							
4	46900	64000	5.0	160	379	294	0.734
8	46500	63800	9.5	155	412	300	0.730
16	44300	58200	12.0	137	425	324	0.760
Annealed, reheated 2 hr at 1500 F, air cooled, tempered at 1100 F							
4	81000	95100	3.0	228	417	355	0.852
8	flaw	.....	.....	...	...	...	.....
16	70800	89000	4.0	216	412	327	0.795
Annealed, reheated 2 hr at 1500 F, air cooled, held 4 hr and reheated at temperatures indicated							
70 F	68533	88700	2.83	212	418	324	0.774
600 F	.....	.....	.....	410	...	...	.....
800 F	.....	.....	.....	263	...	...	.....
1000 F	70700	93200	3.0	229	406	308	0.759
1100 F	69000	94400	3.5	217	435	318	0.732
1200 F	60633	79450	5.0	180	441	336	0.764
1300 F	52700	70500	7.0	170	415	310	0.748
White, heated 17 hr at 1700 F, furnace cooled to 1500 F, held 2 hr, air cooled, tempered at 1200 F							
4	57500	77275	4.75	185	417	310	0.745
8	53300	65275	5.4	151	432	352	0.828
16	46517	57550	7.0	141	408	330	0.808
24	45800	56150	8.0	140	401	327	0.815
Annealed, reheated 2 hr at 1445 F, oil quenched, tempered at 1200 F							
4	67450	75500	5.25	207	364	326	0.894
8	55100	66700	7.5	175	381	315	0.825
16	47500	57400	9.5	146	393	325	0.829
Annealed, reheated 2 hr at 1500 F, oil quenched, tempered at 1100 F							
4	92600	96100	2.0	269	357	344	0.964
8	74600	100500	3.0	228	442	327	0.743
16	flaw	.....	.....	...	...	...	.....
Annealed, reheated 2 hr at 1500 F, oil quenched, and reheated at temperatures indicated							
600 F	.....	.....	.....	230	...	...	.....
800 F	.....	.....	.....	225	...	...	.....
1000 F	79500	96200	2.0	229	420	347	0.826
1100 F	75000	88000	3.0	212	415	363	0.853
1200 F	70400	79500	4.5	208	382	338	0.885
1300 F	51500	60000	8.5	160	375	322	0.858

These irons were then heat treated experimentally in various manners and the specimens were tested for yield point, tensile strength, elongation, and hardness. The results are presented in Tables 8 through 16. The values under plain iron are all averages of from two to four specimens; the alloyed iron results are in general individual tests.

### Interpretation of Results

The results presented in the foregoing tables can be interpreted in order to yield considerable information which is not obvious by simple inspection of the data.

### Influence of Time of Tempering

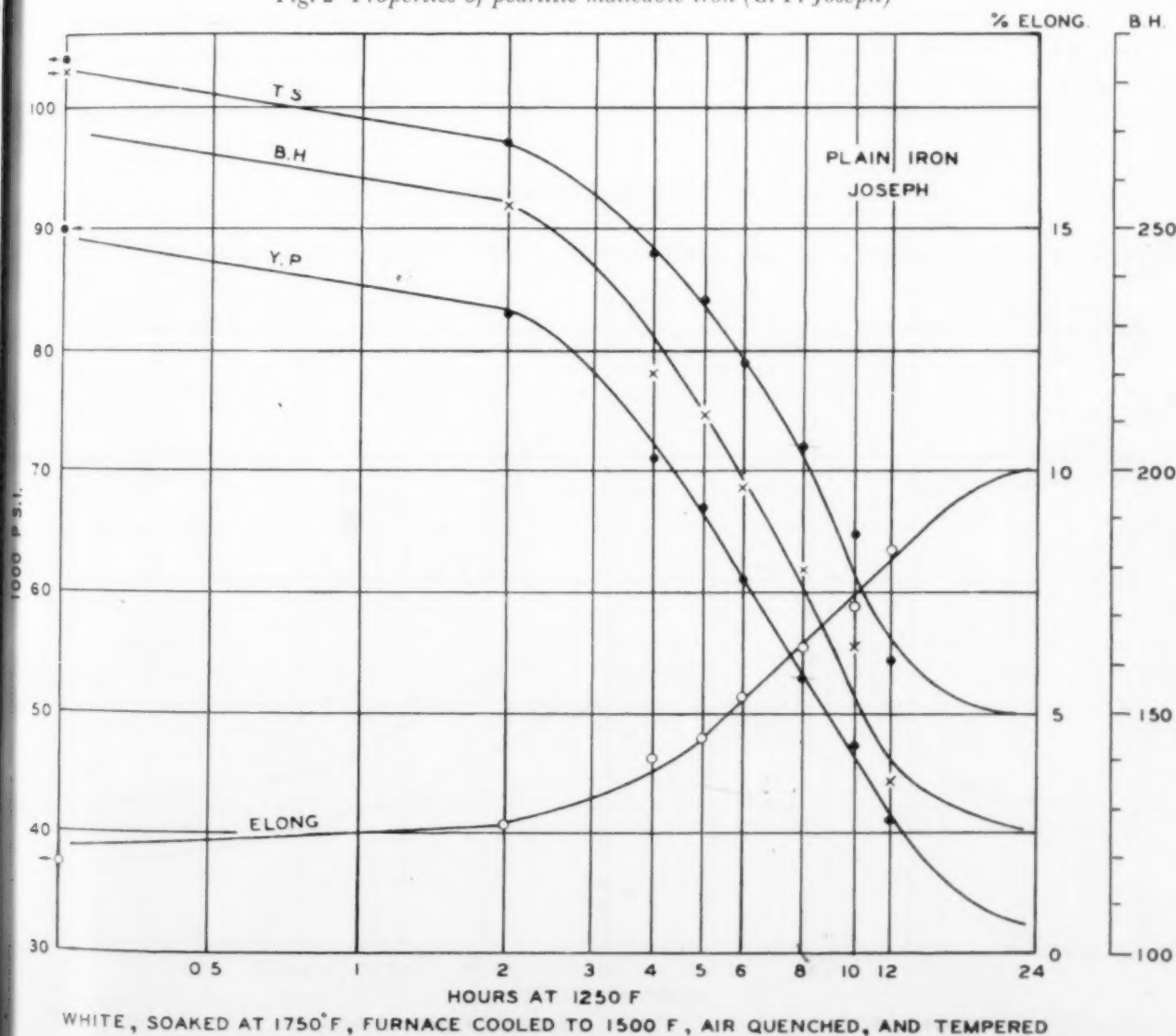
It may be noted from, for example, Table 8, that the period of time at tempering temperature influences the properties. In this case a plain, fully annealed iron was reheated at 1500 F for 2 hr to saturate the austenite; the samples were air quenched and then tempered at 1200 F. It can be noted that tempering

up to 8 hr had little effect upon the yield point although the elongation rose as soon as tempering was started. After 8 hr, the yield point dropped appreciably and the elongation rose. Similar behavior is shown in Table 1 dealing with Joseph's<sup>1</sup> iron although it is less obvious from the table of data alone.

Figures 2 and 3 show the influence of time of tempering at 1250 F of Joseph's iron and of plain iron at 1200 F of this research respectively. Time is plotted as abscissa on a logarithmic scale while yield point, tensile strength, Brinell hardness, and elongation are plotted as ordinate. With Joseph's iron, air-quenched from 1750 F, tempering at 1250 F for 2 hr had little effect on the properties. When the time of tempering exceeded 2 hr there was a rapid change in all properties. No data are given beyond 12 hr and the lines in the graph have been extrapolated to come to equilibrium at 32,500 psi yield point and over 10 per cent elongation.

Figure 3, dealing with the data of Table 8, shows the behavior of a lower carbon iron, air quenched from 1500 F and tempered at 1200 F. At this lower

Fig. 2—Properties of pearlitic malleable iron (C. F. Joseph)



temperature, the yield point was not appreciably affected until after 8 hr of heating. The elongation, however, rose after 1 hr of tempering, thus producing a more favorable yield point to elongation relationship. The longest period for tempering in this investigation was 35 hr. The curves have been extrapolated to the average values for fully annealed iron of this class.

The foregoing two examples indicate that a higher tempering temperature tends to accelerate the decrease in yield point and increase in elongation. Since, however, two different irons with two different pre-treatments were used some doubt may be raised at the validity of this observation. It is felt that the first stage of tempering where yield is almost constant and elongation increases is due to relief of cooling stresses and to spheroidization of the carbides. The second stage where yield drops and elongation rises is brought about by reduction in amount of carbides due to graphitization. This view is in keeping with the results of Massari<sup>4</sup> who heat treated car-wheel irons at various temperatures both above and below the critical. He found an increasing time lag in start of graphitization of combined carbon as lower annealing temperatures were used.

#### Influence of Tempering Temperatures on Properties

In order to study the influence of tempering temperatures, annealed specimens of Iron No. 5 which

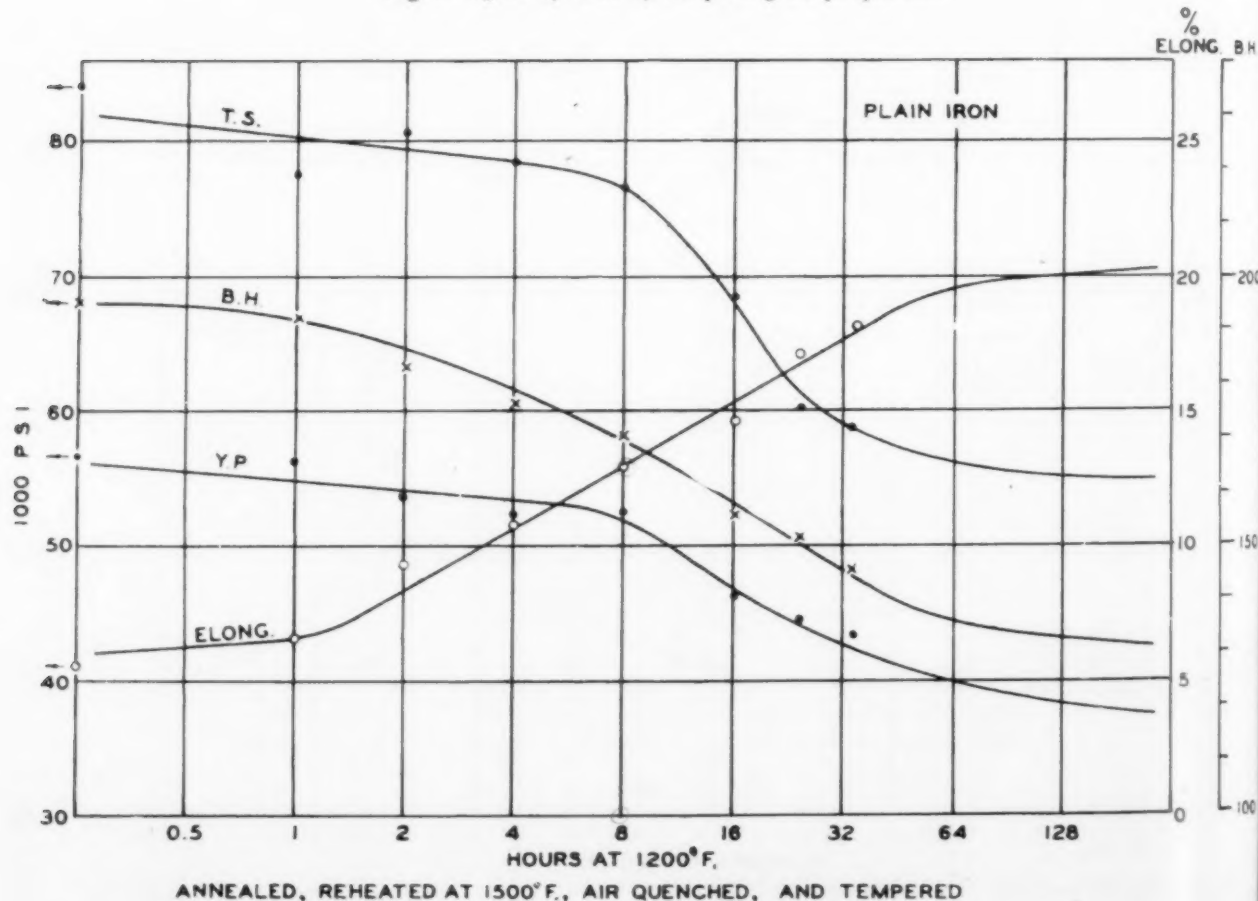
contained 0.5 per cent nickel and 0.5 per cent copper were reheated to 1500 F for 2 hr, air quenched, and then tempered for 4 hr at various temperatures up to 1300 F. Tensile bars were tempered at 1000, 1100, 1200, and 1300 F and smaller specimens were tempered at 600 and 800 F. The results are given in Table 1 and graphically in Fig. 4. The curves show definitely that higher tempering temperatures operate faster than low ones. The Brinell hardness of the specimen tempered at 600 F was over 400 whereas the specimen not tempered at all after air quenching was 212. Since the structures upon quenching were martensitic in appearance, this hardness behavior may be due to retained austenite. It must be borne in mind that the matrix is alloyed not only with 0.5 per cent copper and 0.5 nickel but also with over 1.3 per cent silicon and a small amount of manganese.

It is felt, however, that if plain iron is used, tempering temperatures in excess of 1200 F will provide poorer control of properties since graphitization will set in perhaps sooner than spheroidization. The values for plain irons in Table 2 bear this out since the drop in yield point is very sharp at the 1325 F<sup>2</sup> temperature of tempering used in that test.

#### Influence of Quenching Temperature

When fully annealed irons are reheated above the critical, the highest temperature used will determine the austenite grain size and the amount of combined

Fig. 3—Effect of time of tempering on properties.





carbon. To study the effect of quenching temperature, a group of annealed plain irons (No. 1) were reheated for 2 hr at 1600 F, air quenched, and tempered for various periods at 1200 F. The results are shown in Table 8 and also in Fig. 5. Although the behavior is the same in general as when a reheating temperature of 1500 F was used (see Fig. 3), it is obvious that the yield strength starts at a higher level due to the increased combined carbon.

Due to the lack of annealed specimens, white bars were heated at 1700 F for 17 hr, furnace cooled to 1500 F, air quenched, and tempered at 1200 F. The austenite grain size as determined by the 1700 F annealing temperature and the combined carbon by microscopic examination were higher than that of either of the other two treatments. The results of the physical tests are given in Table 8 and in Fig. 6. The starting level of yield point is higher than where the heating temperature prior to air quenching was lower.

#### Influence of Quenching Medium

The behavior of annealed iron, reheated 2 hr at 1500 F, air quenched, and tempered at 1200 F was shown in the first part of Table 8 and in Fig. 3.

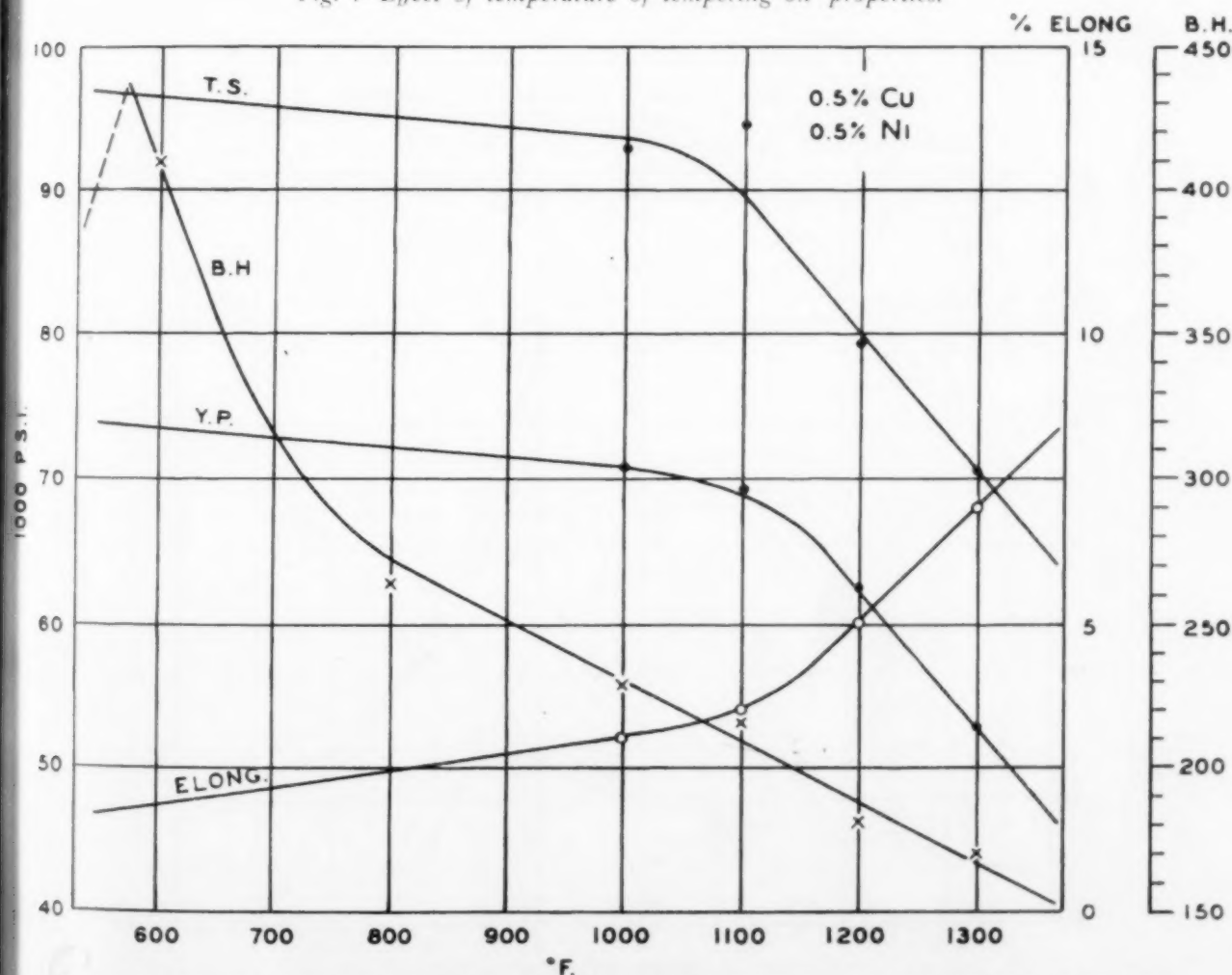
The same treatment was used on another group of specimens except that the irons were oil quenched from 1500 F. The results are given in Table 8 and in Fig. 7. Comparison of Figs. 3 and 7 show that oil quenching leaves the iron in a harder and stronger condition than air quenching but on tempering, these differences are equalized. Comparison of the tabular values indicate that oil quenching produces a higher ratio of yield point to tensile strength; a given yield point is accompanied by a lower tensile strength in oil quenched irons than in air quenched.

#### Relationship Between Yield Point and Elongation

The engineer is always confronted by the fact that he cannot have maximum strength and maximum toughness simultaneously. It is the desire of the foundryman, however, to provide the best relationship of strength and toughness possible. In malleable irons these properties are best measured by the yield point and by the per cent elongation.

The properties of irons from Table 8 were therefore replotted in Fig. 8 where yield point in 1000 psi is plotted as ordinate and per cent elongation as abscissa. It is noted for example for irons air quenched from

Fig. 4—Effect of temperature of tempering on properties.



\* 5 IRON, ANNEALED, REHEATED AT 1500°F, AIR QUENCHED, AND TEMPERED 4 HOURS

1500 F that the first values fall on a line which is almost horizontal; then the values fall on a curved line with a steeper slope. The points for irons air quenched from 1600 F follow a similar course except that the initial points form a branch higher up than the irons treated at 1500 F. Those irons heated at 1700 F start still higher up. When the values for oil quenched irons are included, the graph is composed of several branches intersecting an envelope curve which represents the best relationship of yield point and elongation obtainable with this iron.

The significance of Fig. 8 is that pearlitic malleable irons may possess almost any set of physical properties; only by proper heat treatment can the optimum properties be achieved.

It was felt that some confirmation might be necessary to establish the existence of the envelope curve and its shape. In Fig. 9, the envelope part of the curve is reproduced along with the data given by Joseph in Table 1. The A.S.T.M. specifications<sup>5</sup> for the various grades of pearlitic and annealed irons were also plotted. It can be seen that Joseph's results plot along a smooth curve. The A.S.T.M. specifications not only plot along a smooth curve but this curve is parallel to the envelope curve determined in this investigation.

The shape of the envelope curve indicated that a logarithmic relationship might exist between yield point and elongation. Figure 10 is a repetition of Fig.

8 except that the abscissa is plotted on a logarithmic scale. Now the graph consists of branches intersecting a straight line which is the envelope curve. The equation of this line is

$$Y = -50 \log E + b$$

where  $Y$  equals yield point in 1000 psi,  $E$  equals per cent elongation, and  $b$  is a constant. If the values for annealed plain iron from Tables 4, 5, and 6 are added, the points fall along a line where

$$Y + 50 \log E = 102$$

If the A.S.T.M. and Army Ordnance specifications are plotted in this manner, these specifications may be expressed by the equation

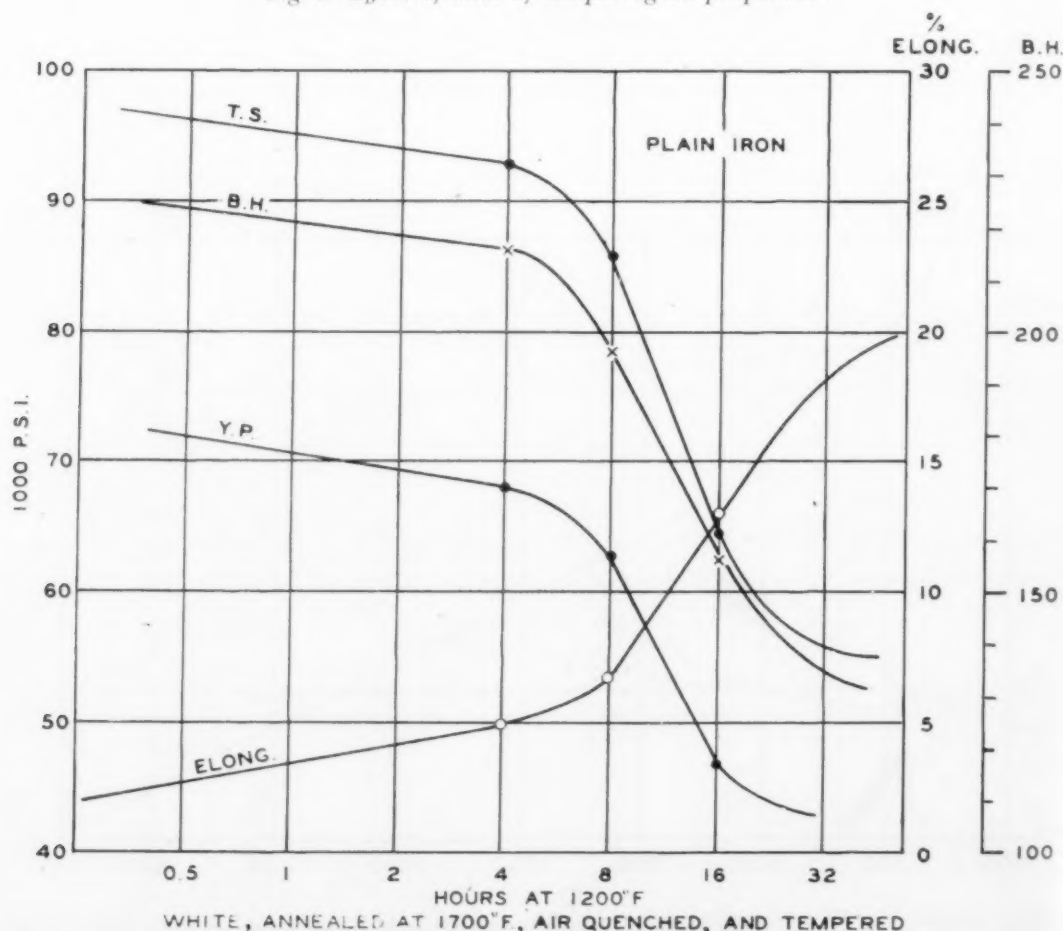
$$Y + 50 \log E = 95$$

This means that for any given elongation the expected yield point for plain irons should be 7000 psi higher than specification or 1000 (102-95) psi. In order to facilitate computations Table 17 is included in which the value,  $50 \log E$ , has been computed for elongations up to 20 per cent.

The constant,  $b$ , is, therefore, a sort of quality index; if the yield point in 1000 psi is added to  $50 \log E$ , the sum must be at least 95 to pass specifications. The higher the value of  $b$ , the better the combination of strength and toughness which has been achieved. This constant has been computed for all data presented in all the tables.

If the values for the quality index,  $b$ , from Table 8

Fig. 5—Effect of time of tempering on properties.



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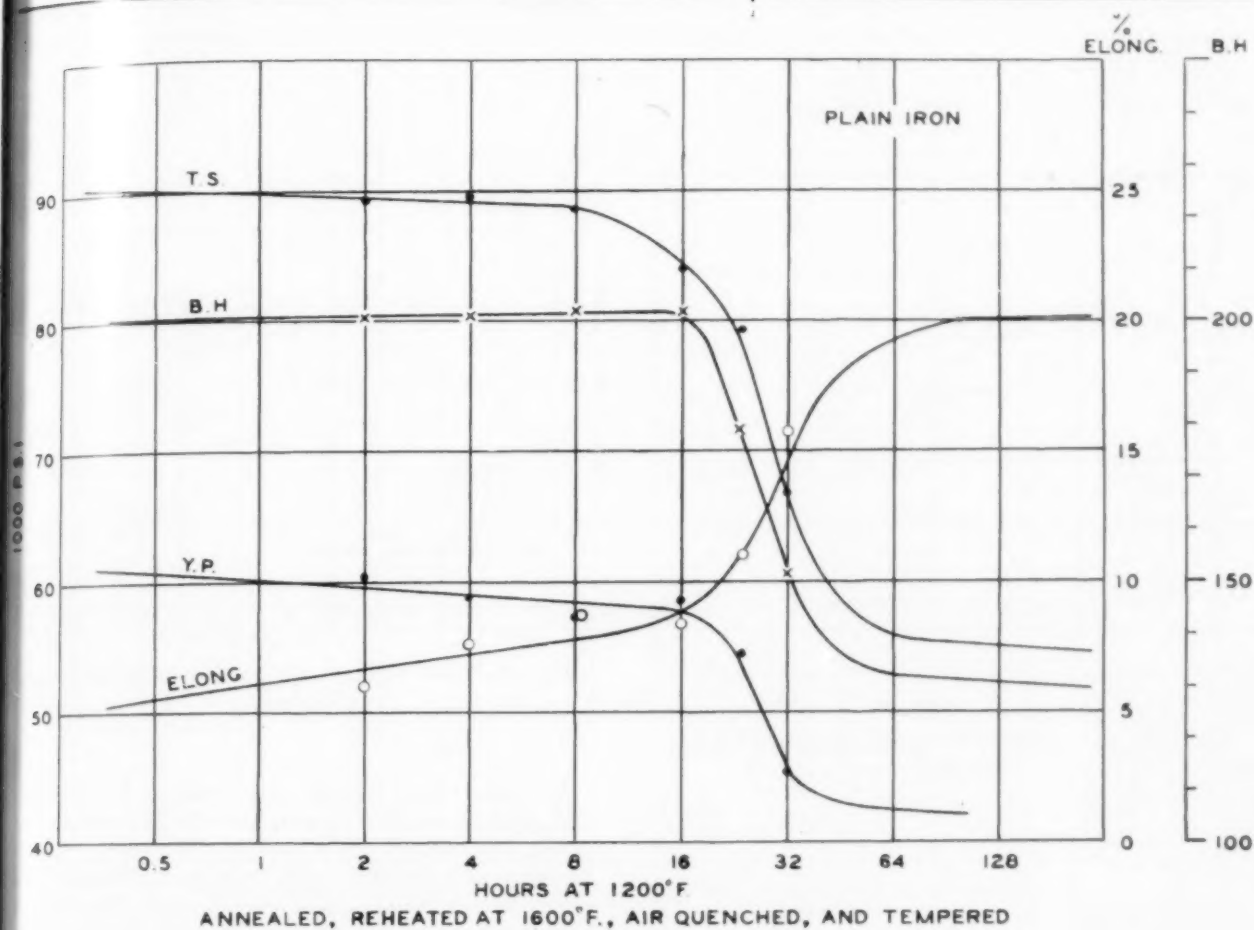


Fig. 6—Effect of time of tempering on properties.

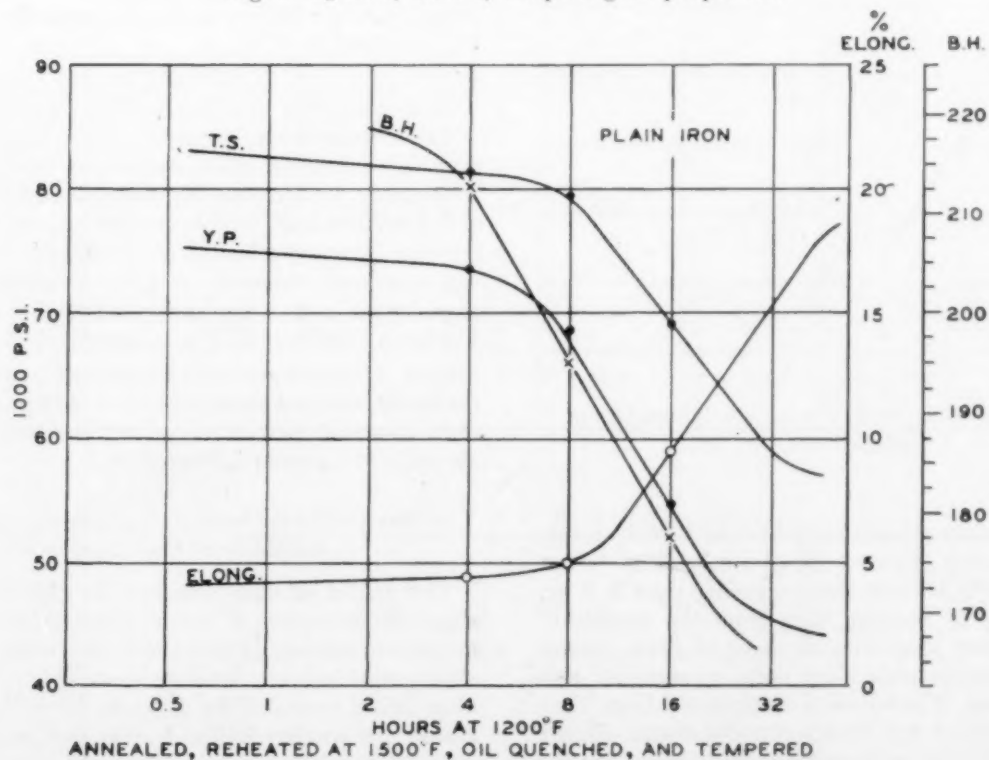


Fig. 7—Effect of time of tempering on properties.



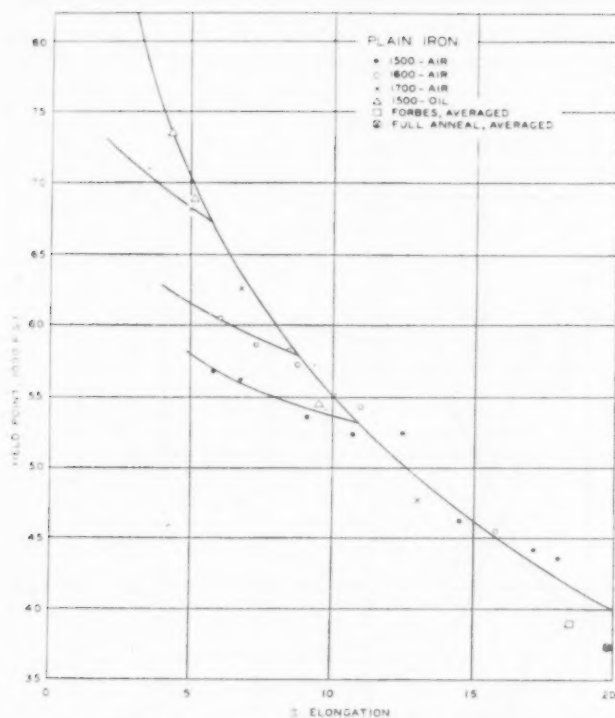


Fig. 8—Yield strength—Elongation relationship of plain irons.

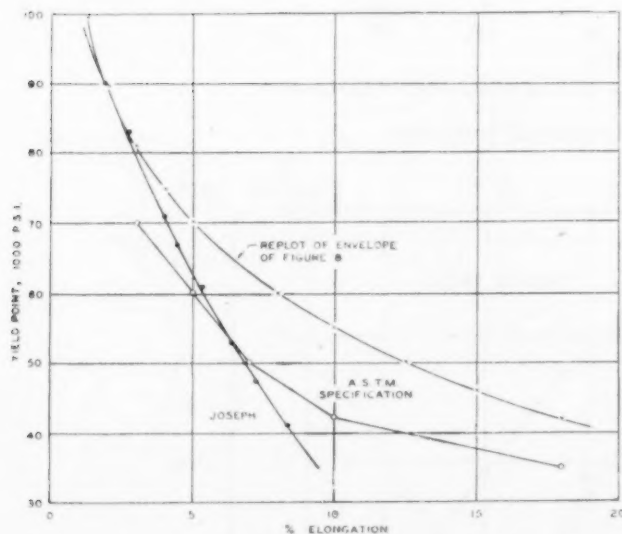


Fig. 9—Yield strength—Elongation relationship of plain irons.

are plotted against time of tempering at 1200 F, it may be seen from Fig. 11 that maximum quality is not attained at 1200 F until the tempering time is 8 hr.

As a matter of interest, data from the literature<sup>6</sup> dealing with the properties of wrought plain carbon steels were plotted with yield point as ordinate and  $\log E$  as abscissa. The values were obtained from Sisco and from various S.A.E. heat treatment charts. Figure 12 shows that there is an unmistakable trend for the values to fall on a straight line whose equation is

$$Y + 150 \log E = 275.$$

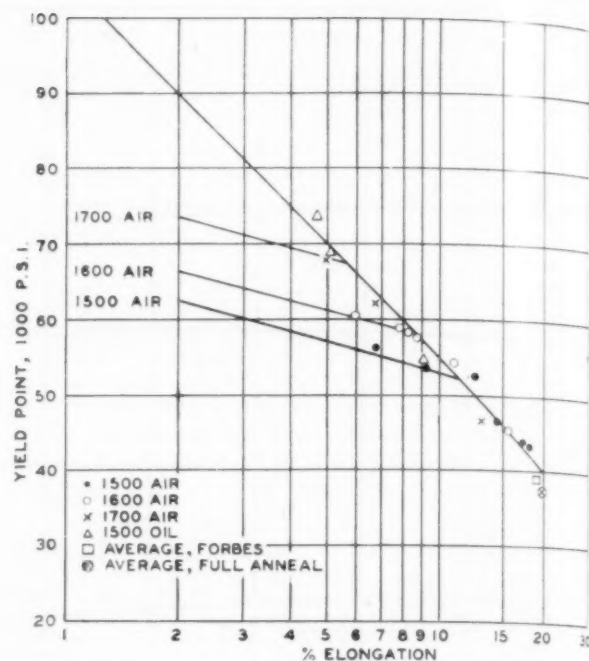


Fig. 10—Yield Strength—Elongation relationship of plain irons.

Wrought alloyed steels were similarly investigated and Fig. 13 shows a similar trend. The line on the graph was constructed according to the mean data of Janitzky and Baeyertz who published averaged data regarding the properties of alloyed steels. Other points have been added for 12.5 per cent Cr turbine blade steel, nitralloy, and other alloyed steels. The scatter of points is not excessive and the equation expressing the relationship is

$$Y + 350 \log E = 565.$$

This relationship does not hold for austenitic steels.

These three charts simply illustrate trends and care must be exercised when comparing malleable irons with steels. In determining elongation of steels, standard American and British practice uses a relationship between gage length and cross sectional area of the test specimen where  $L = 4.5 \sqrt{A}$ , where  $L$  equals gage length and  $A$  equals cross sectional area of the specimen. Malleable iron standard bars use  $L = 3.6 \sqrt{A}$ . Hence a malleable iron whose elongation is normally reported as 10 per cent would probably show only about 14 per cent if a test bar were machined according to standard dimensions.

#### Use of Quality Index, $b$ , as a Measure of the Suitability of Heat Treatment

The tables of data obtained in this investigation show a wide variety of values which vary considerably among themselves. Particularly in Tables 4, 5, and 6 there are values for the same irons which differ widely after being annealed by different foundries. For this reason the quality index,  $b$ , was computed for these irons and is presented in Table 18. It may be seen that the index values are consistently better in irons given heat treatments at Foundry B.

TABLE 12—PROPERTIES OF NO. 8 IRON CONTAINING 1% NI AND 3% MO

Hr at Temp.	Y.P.	T.S.	Elong.	B.H.	T.S. B.H.	Y.P. B.H.	Y.P. T.S.	b
Annealed, reheated 2 hr at 1500 F, air cooled, tempered at 1200 F								
4	79500	92100	4.0	210	438	379	0.863	109.6
8	80500	92200	5.0	210	438	383	0.875	115.4
16	67000	77000	8.0	175	440	383	0.870	112.1
Annealed, reheated 2 hr at 1500 F, air cooled, tempered at 1300 F								
8	58600	74500	9.0	160	466	367	0.790	160.3
Annealed, reheated 2 hr at 1500 F, oil quenched, tempered at 1200 F								
4	88000	91300	2.0	222	411	396	0.965	103.1
8	78500	79000 <sup>1</sup>	2.0 <sup>1</sup>	210	...	373	.....	.....
16	70000	70600 <sup>1</sup>	2.0 <sup>1</sup>	190	...	369	.....	.....
Annealed, reheated 2 hr at 1500 F, oil quenched, tempered at 1300 F								
8	62500	68500	7.0	180	381	347	0.913	104.8

TABLE 13—PROPERTIES OF NO. 10 IRON CONTAINING 0.5% NI, 0.5% CU, AND 0.5% MO

Hr at Temp.	Y.P.	T.S.	Elong.	B.H.	T.S. B.H.	Y.P. B.H.	Y.P. T.S.	b
Annealed, reheated 2 hr at 1500 F, air cooled, tempered at 1200 F								
4	74000	87000	4.5	222	392	333	0.850	106.7
8	72000	85000	6.0	216	393	333	0.848	110.9
16	66000	78267	7.0	193	401	338	0.844	108.3
24	59400	71600	9.5	182	393	326	0.830	108.3
White, heated 17 hr at 1700 F, furnace cooled to 1500 F, held 2 hr, air cooled, tempered at 1200 F								
8	73400	87500	4.5	216	405	340	0.839	106.5
24	54900	66750	9.0	167	400	329	0.824	102.6

TABLE 14—PROPERTIES OF NO. 12 IRON CONTAINING 0.5% CU

Hr at Temp.	Y.P.	T.S.	Elong.	B.H.	T.S. B.H.	Y.P. B.H.	Y.P. T.S.	b
Annealed, reheated 2 hr at 1500 F, air cooled, tempered at 1200 F								
0	67400	83800	3.5	200	419	337	0.804	94.6
4	51500	67800	4.75	165	412	312	0.760	85.3
8	45850	62800	7.0	144	436	318	0.731	88.1
16	40000	52000	8.5	136	383	295	0.770	86.5
24	40700	50500	8.0	124	407	328	0.808	86.0
White, heated 17 hr at 1700 F, furnace cooled, to 1500 F, held 2 hr, air cooled, tempered at 1200 F								
4	57050	70450	4.5	165	426	346	0.810	89.7

TABLE 15—PROPERTIES OF NO. 13 IRON CONTAINING 1.0% CU

Hr at Temp.	Y.P.	T.S.	Elong.	B.H.	T.S. B.H.	Y.P. B.H.	Y.P. T.S.	b
Annealed, reheated 2 hr at 1500 F, air cooled, tempered at 1200 F								
0	69250	89900	3.5	205	438	338	0.781	96.5
4	56225	71300	3.75	180	397	313	0.789	85.0
8	44000	56750	9.0	132	430	334	0.775	91.7
16	44300	55700	8.0	134	416	330	0.795	89.5
White, heated 17 hr at 1700 F, furnace cooled, to 1500 F, held 2 hr, air cooled, tempered at 1200 F								
6	52900	67000	6.5	156	429	339	0.789	93.1

TABLE 16—PROPERTIES OF NO. 16 IRON CONTAINING 1.0% CU AND 0.3% MO

Hr at Temp.	Y.P.	T.S.	Elong.	B.H.	T.S. B.H.	Y.P. B.H.	Y.P. T.S.	
Annealed, reheated 2 hr at 1445 F, air cooled, tempered at 1200 F								
4	70800	81000	5.5	172	471	411	0.875	107.8
Annealed, reheated 2 hr at 1500 F, air cooled, tempered at 1200 F								
8	75000	89100	5.0	176	507	425	0.843	109.3
16	74500	83600	4.5	169	495	441	0.890	107.2
Annealed, reheated 2 hr at 1500 F, air cooled, tempered at 1300 F								
8	60000	75400	8.75	152	495	395	0.800	107.0
Annealed, reheated 2 hr at 1500 F, oil quenched, tempered at 1200 F								
4	92000	94000	2.0	210	447	436	0.980	107.0
8	86500	91500	3.0	200	458	435	0.945	109.3
16	74500	82000	5.5	195	420	383	0.910	111.5
Annealed, reheated 2 hr at 1500 F, oil quenched, tempered at 1300 F								
8	68900	79500	6.0	192	414	359	0.865	107.3
Annealed, reheated 2 hr at 1445 F, oil quenched, tempered at 1200 F								
4	86500	92500	3.5	200	462	432	0.936	113.5
White, heated 17 hr at 1700 F, furnace cooled to 1500 F, held 2 hr, air cooled, tempered at 1200 F								
8	73400	90800	6.5	185	491	397	0.810	114.1
24	60200	72550	9.5	159	466	379	0.529	109.1

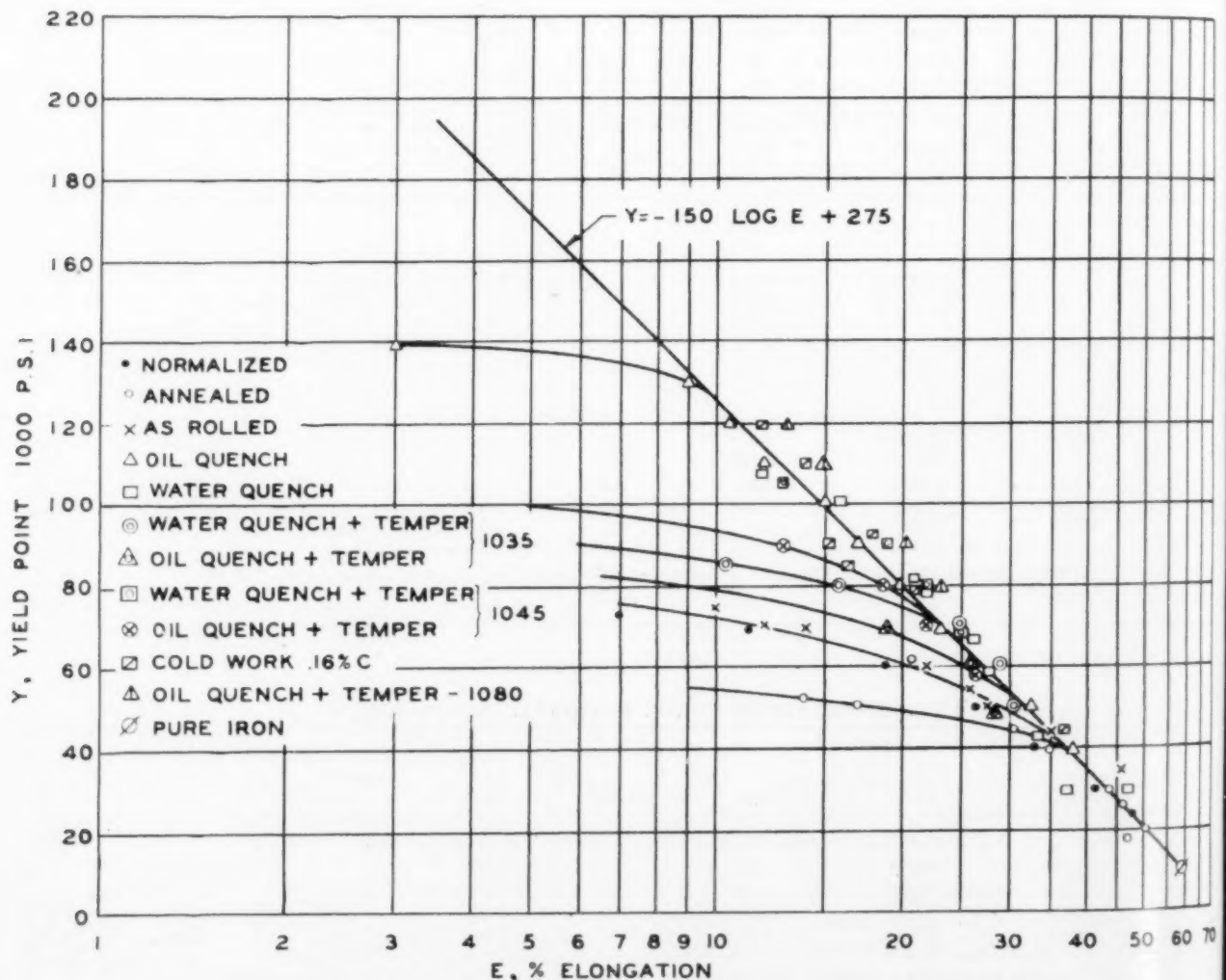


Fig. 11—Yield strength-Elongation relationship of carbon steels.



TABLE 17—VALUE OF 50 LOG E FOR VARIOUS ELONGATIONS

Elongation per cent	log E	50 log E
1.0	0	0
1.5	0.177	8.85
2.0	0.301	15.05
2.5	0.398	19.90
3.0	0.477	23.85
3.5	0.544	27.20
4.0	0.602	30.10
4.5	0.653	32.65
5.0	0.698	34.90
5.5	0.740	37.00
6.0	0.778	38.90
6.5	0.803	40.15
7.0	0.845	42.25
7.5	0.875	43.75
8.0	0.903	45.15
8.5	0.930	46.50
9.0	0.954	47.70
9.5	0.978	48.90
10.0	1.000	50.00
10.5	1.022	51.10
11.0	1.042	52.10
11.5	1.061	53.05
12.0	1.080	54.00
12.5	1.098	54.90
13.0	1.114	55.70
13.5	1.130	56.50
14.0	1.147	57.35
14.5	1.162	58.10
15.0	1.176	58.80
15.5	1.191	59.50
16.0	1.204	60.20
16.5	1.218	60.90
17.0	1.231	61.55
17.5	1.243	62.15
18.0	1.255	62.75
18.5	1.267	63.35
19.0	1.279	63.95
19.5	1.290	64.50
20.0	1.301	65.05

The values for plain irons were averaged for each test and the averaged values for alloyed irons were likewise computed and presented in Table 19. Examination of these averages show that the heat treatment at Foundry B brought out the best properties from the irons; although the treatments at all the foundries were commercially satisfactory for plain irons.

In further discussion of these irons only the values obtained from Foundry B's annealing will be used except where otherwise stated.

Table 19 also points out that it is dangerous to alloy irons indiscriminately for heat treatment in a standard annealing cycle designed for producing fully annealed plain iron. If such a treatment is to be used the amount and kind of alloy will have to be determined for each foundry's particular annealing cycle.

#### Influence of Alloys

Alloys have been employed in pearlitic malleable irons in order to increase the mechanical properties. Table 3 which presents the data obtained by Forbes<sup>3</sup> shows that manganese in excess of that normally included to counteract the effects of sulphur progressively increases the properties and is reflected in an increase in quality index with increasing amounts of alloy.

TABLE 18—QUALITY INDEX b FOR VALUES SHOWN IN TABLES 4, 5, AND 6

Iron	Foundry	C	D	A
From Table 4				
2		103.2	102.6	101.7
3		106.6	105.6	104.8
4		100.4	100.7	95.0
5		.....	100.2	94.0
6		.....	.....	.....
7		.....	.....	.....
8		104.5	96.9	107.1
9		103.7	103.6	99.6
10		.....	99.3	.....
11		.....	.....	.....
12		102.3	111.5	99.2
13		106.6	94.1	.....
14		.....	.....	.....
15		.....	.....	.....
16		107.7	104.6	98.6
17		104.8	104.4	99.0
18		100.7	100.6	102.4
19		109.5	102.6	95.3
20		104.8	104.4	86.5
21		102.7	102.7	92.2
22		106.8	103.9	97.4
23		104.0	108.6	93.8
24		107.0	106.4	92.7
25		97.4	107.7	98.6
26		80.7	89.8	70.4
27		107.9	106.3	101.7
28		.....	.....	.....
29		105.9	104.3	106.3
30		.....	.....	.....
Base		101.3	102.2	93.8
Base		101.3	101.3	102.1
From Table 5				
4		102.0	110.0	102.2
5		112.0	106.9	103.3
6		113.1	107.6	92.8
7		95.9 <sup>1</sup>	106.3	101.2
13		105.2	113.7	105.7
14		109.5	114.0	82.7
15		108.3 <sup>1</sup>	113.7	93.6
26		109.0	108.4	105.3
30		114.9	111.9	78.0
Base iron		100.2	106.3	106.5
From Table 6				
4		96.7	96.6	102.3
5		92.7	104.7	113.1
6		80.3	110.1	100.3
12		97.5	104.8	104.6
13		88.6	104.4	94.5
14		99.0	117.0	76.0
16		107.5	102.1	103.5
17		91.5	107.1	97.8
18		99.9	107.8	96.4
22		102.2	107.6	102.5
23		94.8	108.4	98.2
24		89.5	110.2	91.1
25		108.0	115.1	101.6
26		101.7	113.0	85.3
29		93.0	107.9	.....
30		89.5	111.6	.....
Base iron		101.3	102.6	93.8
Base iron		101.3	101.3	103.1

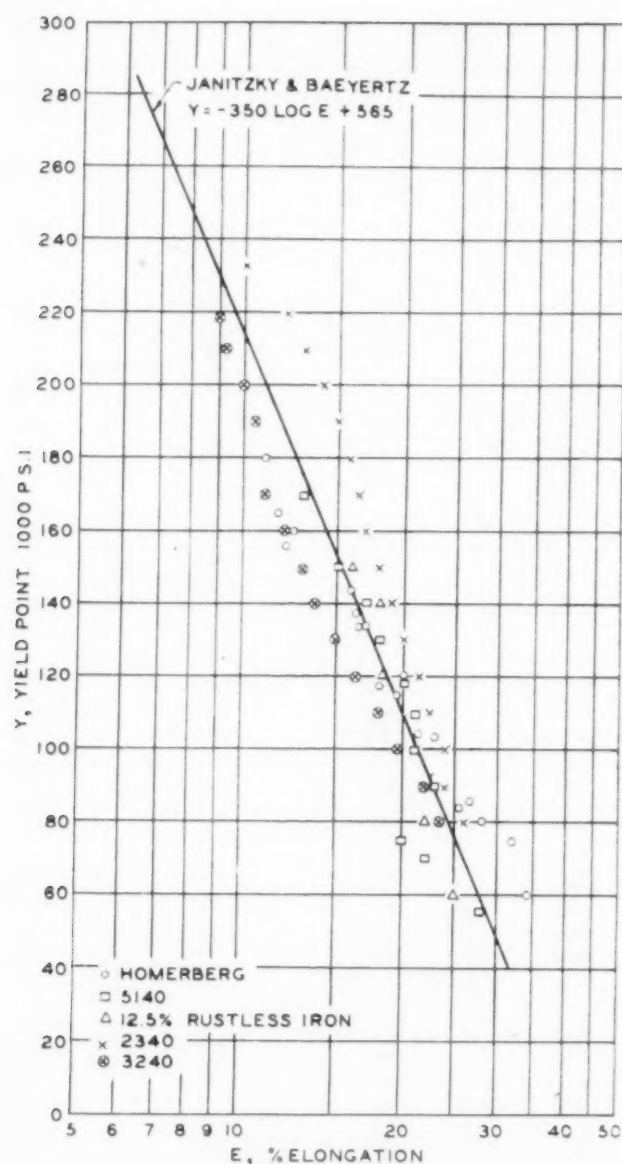


Fig. 12—Yield strength-Elongation relationship of alloyed steels.

TABLE 19—PERFORMANCE OF DIFFERENT FOUNDRIES IN HEAT TREATING THE SAME IRON

Foundry	A	B	C	D
Quality Index <i>b</i> for Plain Iron				
Test I	97.95	.....	101.3	101.8
Test II	106.50	106.3	100.2	.....
Test III	103.3	102.0	.....	.....
	98.45			
Average	99.86	103.4	101.08	101.8
Quality Index <i>b</i> for Alloyed Irons				
Test I	96.87	.....	103.59	103.1
Test II	97.13	110.26	109.25	.....
Test III	97.50	107.90	.....	.....
	97.70			
Average	96.76	109.00	105.10	103.1
Number of bars	60	25	28	22

Advantages in the use of alloys theoretically would result either from their going into solid solution or

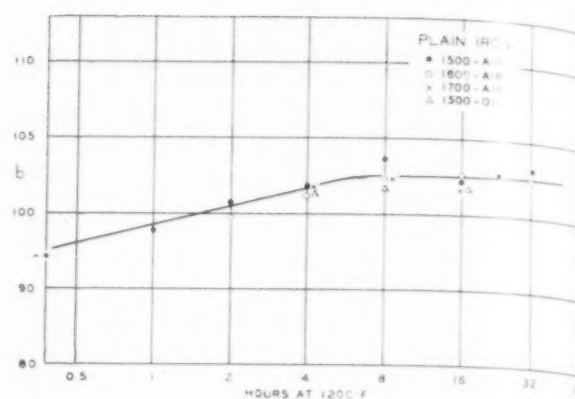


Fig. 13—Effect of tempering on quality index, *b*.

TABLE 20—EFFECT OF ALLOYS ON YIELD POINT OF FULLY ANNEALED IRON

Per Cent Alloy	Per Cent Increase in Yield Point	Per Cent Elongation			
Ni					
0.5	9.93	22.5	110.0		
0.67	11.27	12.0	96.8		
1.0	15.60	17.5	106.5		
1.32	20.10	16.5	104.7		
1.5	17.40	17.5	107.8		
1.95	27.60	19.0	110.1		
2.0	28.80	13.5	106.2		
Cu					
0.50	12.6	19.5	103.6		
0.79	14.70	19.5	104.4		
1.00	21.00	16.5	102.4		
1.35	19.40	15.9	102.3		
1.50	23.05	13.3	101.2		
1.97	20.00	16.3	103.9		
Mo					
0.15	2.65	21.0	103.2		
	2.69	19.5	102.6		
Mo					
0.30	3.43	23.0	106.6		
	5.40	21.5	105.6		
Ni+ Mo					
0.69	0.37	17.89	17.5	104.3	
1.0	0.3	22.85	21.0	113.7	
1.35	0.4	27.15	14.5	104.4	
1.5	0.3	33.45	18.5	114.0	
2.0	0.3	36.60	16.5	113.7	
2.1	0.4	41.10	21.0	117.0	
Cu+ Mo					
0.73	0.34	20.70	19.0	107.6	
1.3	0.35	27.15	19.0	108.4	
1.9	0.38	41.10	20.0	110.2	
Cu+ Ni					
0.74	0.68	34.50	21.5	115.1	
1.0	1.0	34.20	13.5	108.4	
1.52	1.3	38.80	18.0	113.0	
Cu+ Ni+ Mo					
0.64	0.61	0.33	17.08	20.5	107.3
1.0	1.0	0.3	47.45	12.5	111.3
1.24	1.21	0.35	41.20	20.5	111.6

from their influence on critical cooling rate. On the other hand, alloys may inoculate the castings to produce some primary graphite, they may stabilize the massive carbides, or stabilize pearlitic carbides.

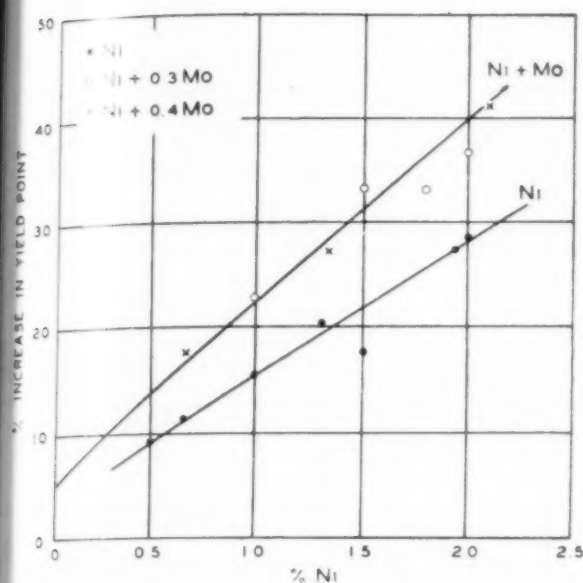


Fig. 14—Influence of Ni and Mo.

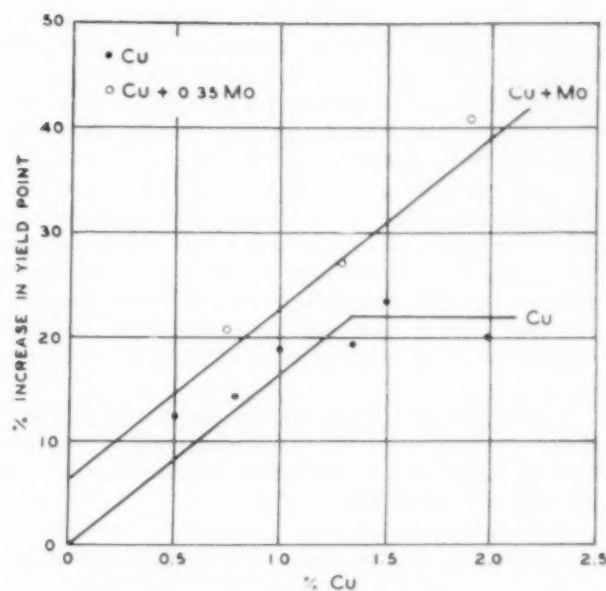


Fig. 15—Influence of Cu and Mo.

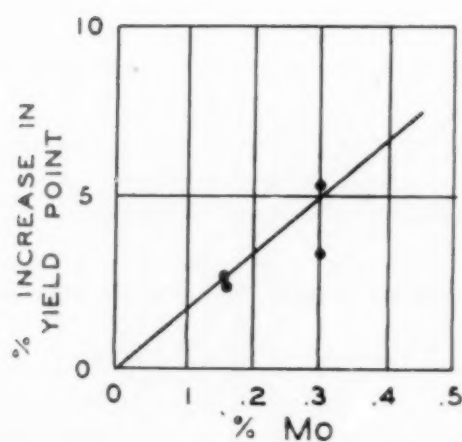


Fig. 16—Influence of Mo.

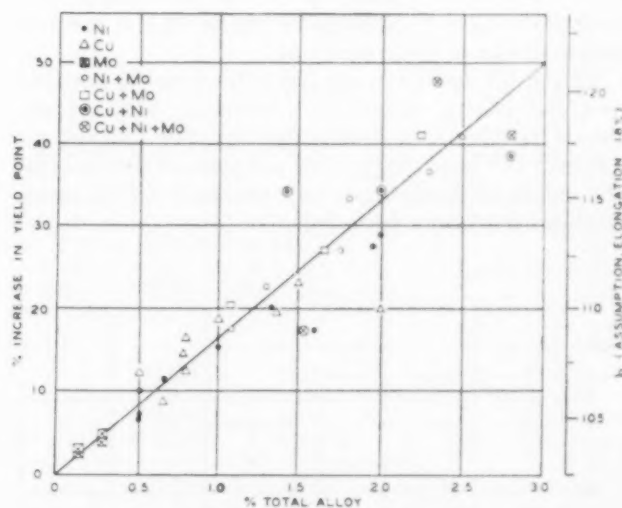


Fig. 17—Influence of alloys on annealed malleable iron.

The influence of alloys by virtue of solid solution effects was studied by using data from Tables 4, 5, and 6 where heat treatment was done by Foundry B. Other values were averaged in if the elongations were such as to indicate fully annealed, ferritic iron. First, the yield points of plain iron bars were averaged. This number was subtracted from the yield points of the individual alloyed irons. The difference equals the increase in yield strength and when divided by the value for plain iron gives the per cent increase due to alloys. These computations are summarized in Table 20. The elongations have been added to indicate the degree of completion of annealing as well as the quality index,  $b$ .

The results for various alloys and combinations of alloys are given in Figs. 14, 15, and 16. Since all these irons were fully or nearly ferritic the alloys must have influenced the properties solely by virtue of their solid

solution effect. In Fig. 15 the line for copper seems to break at around 1.5 per cent which is its limit of solubility in iron. In Fig. 17 the increase in yield point is plotted against amount of total alloy. The trend is unmistakable that the quantity of alloy in solid solution is the determining factor in establishing yield point. If all the irons are assumed to possess 18 per cent elongation, then the value for quality index,  $b$ , may be read from the scale at the right.

There may be a difference in the effect of different alloys, but in the limited number of test runs no differences were detected.

When the values for quality index,  $b$ , are computed for all irons, whether ferritic or pearlitic, from Tables 4, 5, and 6 and heat treated at Foundry B and if divided for statistical examination, Figs. 18 and 19 show the frequency. Plain irons have a statistical average quality index of around 102 and alloyed irons



of this test average 108. From the discussion previously made, the index should be controllable by alloy additions. An alloyed iron of index 110 should at a given elongation have 15,000 psi greater yield point than specification and 8000 psi greater than a plain iron, 1000 (110-95) and 1000 (110-102) psi respectively.

The results from the last heat of irons tested were disappointing. The irons alloyed with nickel or copper or mixture exhibited microscopic patches of primary graphite. This was probably due to inoculation effects. Those irons which contained molybdenum in addition were satisfactory since this alloy probably counteracted the inoculating tendencies of nickel and copper.

The results of Ni-Mo, Cu-Mo, and Cu-Ni-Mo irons shown in Tables 12, 13, and 16 are satisfactory. The quality index factors check those obtained previously on fully annealed irons.

#### Malleable Irons in Comparison with Competing Materials

The foregoing results indicate that malleable and pearlitic malleable irons have very satisfactory strength to toughness relationships provided that proper heat treatments have been employed to bring out the full potentialities of these materials.

In Fig. 20 the values of alloyed irons in the pearlitic and the ferritic states are presented. Most of the points fall between lines whose quality indexes are 105 and 115 respectively. For purposes of comparison, the values for plain irons and for the A.S.T.M. specifications have been included.

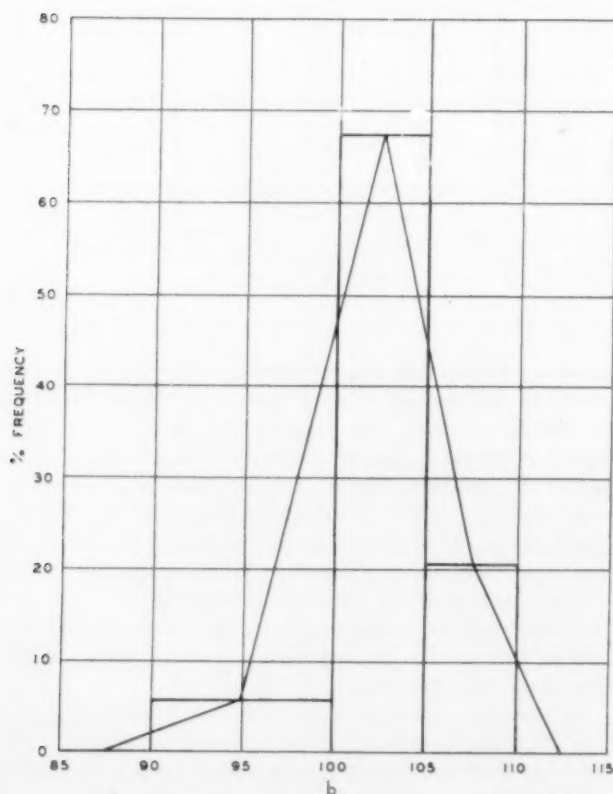


Fig. 18—Distribution of quality index, *b*, of plain irons.

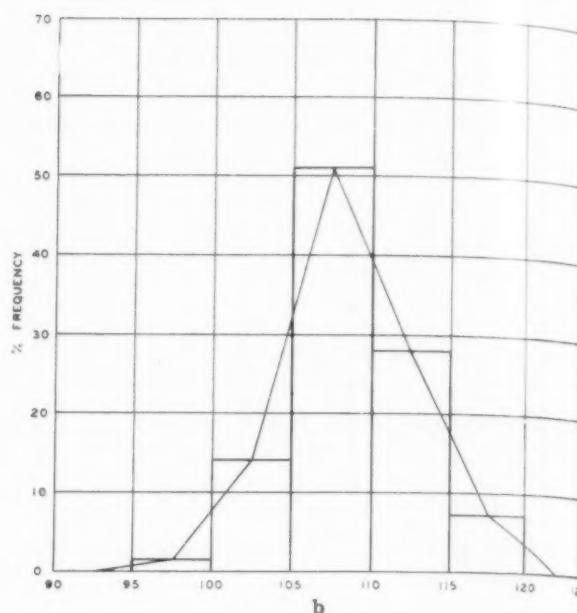


Fig. 19—Distribution of quality index, *b*, of alloyed irons.

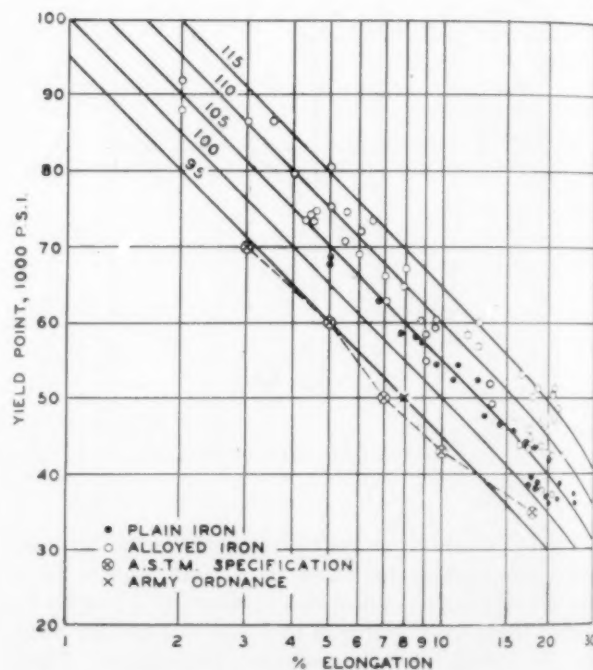


Fig. 20—Yield strength-Elongation relationships of plain and alloyed malleable irons.

It must be remembered, however, that in Fig. 20 the elongation values for malleable irons are based on a shorter gage length than is standard for steel. The elongation would appear less if the standard gage length had been used. The results shown for malleable irons should not be difficult to obtain in commercial practice since equivalent properties have been published by a variety of foundries.<sup>6</sup>

### Summary of Conclusions

On the basis of the work described in this paper, the following conclusions have been drawn:

1. The properties of pearlitic malleable iron are dependent upon the quantity, size, shape, and distribution of the three phases: carbon, ferrite, and carbide.

2. Tempering a quenched iron results in, first, increase in elongation with no or little diminution of yield strength and second, an increase in elongation with a proportional decrease in yield strength.

3. When the second part of the tempering action is reached, the relationship between yield and elongation (or between strength and toughness) becomes fixed. Mathematically, the relationship for annealed and pearlitic malleable irons may be expressed as

$$Y + 50 \log E = b$$

where  $Y$  = yield point in 1000 psi,  $E$  = per cent elongation, and  $b$  = a constant.

4. The values of the A.S.T.M. specifications for annealed and pearlitic irons may be expressed by the above equation with  $b = 95$ .

5. Statistical averages for plain irons with many types of heat treatment give  $b = 102$ ; alloyed irons give an average  $b = 108$  to 110 with some values over 115. This means that an alloyed iron ( $b = 108$ ) has a yield of 6000 psi (108-102) higher than a plain iron ( $b = 102$ ) when both are heat treated to give the same elongation.

6. Adding alloys indiscriminately to irons which are to be heat treated along with unalloyed irons and with the same cycle is dangerous. For good results the quantity and combination of alloys must be evaluated for each annealing cycle because the presence of massive carbides resulting from stabilization or of primary graphite due to inoculation will detract from the properties.

7. The method of computation using the equation

$$Y + 50 \log E = b$$

is valuable in determining the suitability of a given annealing cycle for a given iron.

8. The constant,  $b$ , may be considered a quality index factor for fully annealed or pearlitic malleable irons.

9. In fully annealed irons, improvement in properties by alloys is solely by virtue of their being in solid solution, although some precipitation hardening may be induced in copper-bearing irons. The data show that the increase in properties is proportional to the quantity of alloy regardless of whether it be nickel, copper, molybdenum, or combinations of these. In the amounts used all seem equally effective although copper alone seems to have little effect above 1.05%.

10. On air- or oil-quenched irons, alloys modify the critical quenching rate. Some alloyed irons are air hardening and some exhibit retained austenite upon quenching in oil.

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### DISCUSSION

Chairman: G. VERNERHOLM, Ford Motor Co., Dearborn, Mich.  
Co-Chairman: W. B. McFERRIN, Electro Metallurgical Co., Detroit

C. A. NAGLER (Written Discussion):<sup>1</sup> The authors are to be complimented on their fine paper dealing with physical properties of heat treated, plain and alloyed malleable irons. The paper contains a wealth of information on the various tensile strength, yield strength, elongation, and Brinell hardness of the heat treated malleable iron.

In analyzing the data in Tables 4, 5, and 6 there appears to be some differences in physical properties for the same approximate analysis and heat treatment.

	Table 4 C Yield Strength	Table 5 C Yield Strength	Table 4 A Yield Strength	Table 5 A Yield Strength
Bar No.	Strength	Strength	Strength	Strength
4	38,860	39,800	34,200	40,200
5	39,790	45,480	40,000	44,600
6	34,350	48,000	_____	47,800
7	13,990	52,100	_____	67,600
		C Si	Mn S	P
Table 4—Base Iron	2.36	1.34	0.41 0.120	0.084
Table 5—Base Iron	2.31	1.29	0.37 _____	_____

Can you account for the wide difference in yield strength for Bar No. 7 with 2 per cent Ni as drawn in Tables 4 and 5, 13,990 and 52,100 psi respectively?

The assumption is made that the alloy additions shown in Table 4 were calculated.

The paper could have included some photomicrographs showing typical microstructures produced by the use of heat treatment A, B, C, and D.

PROF. SCHNEIDWIND (Written Reply to Prof. Nagler): As pointed out in the paper, conditions were not well controlled in the first test which was of an exploratory nature. Further, it is significant that for the group of irons cited by Mr. Nagler, seven out of twelve elongation figures are missing which indicates that the specimens must have been defective internally although not apparent visually. No metallographic study was made of these irons since the specimens were not returned after testing.

J. E. REIDER:<sup>2</sup> This paper is a very valuable contribution to the study of high mechanical properties in annealed and pearlitic

<sup>1</sup> Wayne University, Detroit

<sup>2</sup> Bureau of Mines, Ottawa, Canada

malleable irons, plain and alloyed. I have no argument whatsoever with any of the details of the paper. They are so complete that it will take hours to assimilate. In his paper Dr. Schneidewind gives a chart showing  $50 \log E$  for various elongations to save effort in calculating out. In dealing with forgings, when you get to the top and the bottom end of the curve, peculiar things happen around 1 and 2 per cent elongation. Factor  $50 \log E$ , for a change from 1 to 2 per cent elongation, corresponds to a large figure, 15,000 or 16,000 psi change in yield point. On the other hand, around 20 per cent elongation, a 1 per cent change corresponds to about 1,000 psi change in yield point.

Does Dr. Schneidewind feel that in the relatively low elongation irons, this formula would hold? I have a feeling that below 5 per cent elongation, it may not be valuable. It seems to indicate a big difference in the usefulness of iron whether it is 1 or 2 per cent elongation when getting into the very high strength ranges. If you were to go into impact strengths, would that reflect on the other mechanical properties?

Dr. Schneidewind: I agree that when we get down to the low elongations, the formula is not too significant. In fact, I do not think any mathematical formula will hold for properties of metals but may serve simply as a guide. At 3 per cent elonga-

tion, Mr. Joseph's iron would fall somewhere within the range although the iron he reported on, I think, has higher carbon content than the one we used. At 3 per cent elongation the chart shows a yield strength of 78,000 psi for a 2.6 per cent carbon iron. For a 2.3 per cent carbon iron, the yield point calculated is just a shade over 80,000 psi. At those high yield points, I do not think that our method of measuring elongation is very accurate.

MR. REHDER: I agree with Dr. Schneidewind. I was thinking particularly of impact strengths. In gray cast irons, the difference between the smallest measurable elongation and maybe 1 or 2 per cent makes an important difference in the impact strength.

DR. SCHNEIDEWIND: I wish we had a standard test for impact that we could follow, because I think elongation is only a part of the answer. In some work I have done previously, I plotted the per cent elongation against the angles of twist in torsion and there was excellent correlation down to about 6 per cent elongation. That held for fully annealed and for pearlitic malleable when plotting the per cent elongation against the degrees twist in a torsion test.

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# IMPROVEMENT OF MACHINABILITY IN HIGH PHOSPHORUS GRAY CAST IRON

By

William W. Austin, Jr.\*

## ABSTRACT

This paper summarizes the results of an investigation carried out at the Southern Research Institute under the sponsorship of the Woodward Iron Company of Birmingham, Alabama. Following a careful review of the literature on the metallurgy of cast iron and its relation to machinability, a program of experimental work was undertaken with the objective of finding commercially feasible means of overcoming the harmful effects of phosphorus upon the machinability of high phosphorus gray iron castings.

Three fundamental methods of approach have been followed in investigating the problem of improving machinability of high phosphorus gray irons. They may be summarized as follows:

First, modification of the mode of occurrence of the phosphide constituent in the microstructure. This approach has as its object overcoming the abrasiveness of iron phosphide by changing its structure, properties, and composition through alloy additions.

Second, control of microstructure exclusive of the phosphide constituent. The object of this method of attack is to produce a soft ferritic matrix structure instead of the relatively hard pearlitic matrix normally characteristic of high phosphorus gray irons. A structure of this type would exert less abrasive or dulling action on cutting tools during machining operations.

A third approach to the problem deals with outright removal of constituents known to have adverse effects on machinability. Desulphurization and dephosphorization treatments of the molten iron, and combinations of these treatments with alloy additions are included under this heading.

The results obtained from these studies indicate that significant improvement in machinability of a selected base iron may be attained by a combined treatment involving sodium carbonate desulphurization and zirconium alloy addition. The cost of this treatment based on laboratory scale operations is estimated at less than \$2.50 per ton of treated iron.

## Introduction

THE ALMOST UNIVERSAL occurrences of sulphur, oxygen, and phosphorus as detrimental components of ferrous alloys have presented technical problems of major importance since the earliest beginnings of metallurgical science. In the production of gray cast iron, extensive research led to the establishment of commercially feasible desulphurization and deoxidation treatments now in widespread use throughout the industry. Unfortunately the elimination of the harmful effects of phosphorus on gray cast iron is a goal that has not yet been fully attained. Although the

problem of phosphorus in cast iron has received a great deal of attention since the classic work of J. E. Stead in 1900,<sup>1</sup> much still remains to be done before an entirely successful and commercially practicable dephosphorization procedure can be established. This is particularly true of cast irons produced from southern ores where the phosphorus content of pig iron ranges between 0.60 and 0.90 per cent.

It is well known that this amount of phosphorus in gray cast iron has pronounced effects upon the fluidity of the molten metal and upon the abrasion resistance and upon the machinability of the finished castings. The effect on fluidity is beneficial, giving rise to ease of pouring and excellent casting qualities for thin sections of intricate design. The effect of phosphorus on abrasion resistance and machinability, however, is not always to be desired.

This is particularly true whenever high phosphorus castings are machined at high cutting speeds. Here the hard abrasive phosphide causes a drastic reduction in tool life, and such conditions as segregation and nonuniformity of structure give rise to relatively poor machining qualities. The poor machinability rating of high phosphorus gray iron has hindered to a great extent the universal acceptance by the automotive industry and other large volume consumers of castings produced wholly from southern pig iron.

## Purpose of This Study

The purpose of this investigation, therefore, is to seek commercially feasible means of eliminating the harmful effects of phosphorus on the machinability of high phosphorus gray iron castings. From the metallographic standpoint it is well known that the adverse effects of phosphorus on machinability of gray iron are due largely to the formation of a cellular network of the extremely hard phosphide eutectic, steadite, in a matrix of moderately hard pearlite. In machining, the cutting edge of the tool is forced into contact with this network of steadite which cannot readily be pushed aside, being firmly imbedded in the rigid pearlite matrix. This results in a pronounced abrasive or dulling action on the cutting edge thereby shortening tool life.

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Consequently, if the microstructure of high phosphorus cast iron could be altered so as to change radically the mode of occurrence of the phosphide eutectic, it is seen that machinability might be markedly improved. Hence, it is desirable to produce a structure consisting of a more intimate distribution of steadite in a softer matrix structure of ferrite and fine graphite instead of the usual hard pearlitic matrix. In the latter type of microstructure the cutting tool on making contact with the more finely distributed phosphide particles would not be subjected to as severe abrasive action since the soft ferritic matrix would yield and allow the steadite particles to be pushed aside. The achievement of this control of microstructure in high phosphorus gray irons would obviously result in greatly improved tool life and reduced machining costs.

#### Literature Cited

In many cases the cost of machining operations employed to obtain the desired dimensions and surface finish on castings amounts to many times the total cost of the rough casting. Because of these high machining costs numerous technical investigations have been directed toward the evaluation and improvement of machinability. A few of the more pertinent references to machinability are cited because of their relation to the problem at hand.

Machinability has been defined by Boulger<sup>2</sup> as a "complex property of a material which controls the facility with which it can be cut to the size, shape, and surface finish required commercially." Evaluations of machinability are usually based on determinations of tool life, rate of metal removal, or power consumption. Machinability tests are customarily accomplished by standardizing a given machining operation so as to eliminate the effects of as many variables as is possible, and then proceeding to measure some other important variable believed to have a close relationship to the cutting quality or machinability. For example, the quantity of metal removed for a constant feed load or tool life is often used as a criterion for machinability. Measurements of the amount of heat generated during the cutting operations or the power requirement for the operation of a given machine under constant conditions have also been used as bases for comparative tests of machinability.<sup>3</sup>

#### Types of Machinability Tests

Two general types of machinability tests are recognized. First, short tests based on the resistance of the material being machined, and second, long tests based on the life of the cutting tool. Of the short tests, the drill penetration test developed by Keep and Loren in 1897<sup>4</sup> may be described as a typical example of the short term machinability test. The number of revolutions of a standard test drill is plotted against the depth of penetration under a fixed thrust load. Thus the rate of penetration of the test drill is used as a criterion of machinability. Control tests are made on a bar of standard machinability before and after each sample to check the sharpness of the drill point.

Typical of the long term tool life tests for the

measurement of machinability is the milling test developed by Field and Bullock.<sup>5</sup> This test consists of machining heavy cast iron test blocks (6½ by 3½ by 20½ in.) on a large hydraulically controlled milling machine. The milling cutter consists of a single-tooth carbide-tipped cutter mounted in a 500-lb. fly wheel. Tool life is determined by measuring the degree of wear on the clearance land of the cutter with a Brinell microscope after each pass. The test is considered ended when the width of the wear land reaches 0.001 in. From the data obtained in this test the authors have been able to evaluate tool life in terms of the following criteria: (a) actual cutting time to dull one tooth; (b) the number of chips required to dull one tooth; (c) the total distance traveled by the tooth; (d) the total volume of metal removed per tooth; and (e) cubic inches of metal removed per inch of cutting edge per tooth.

Another tool life test for evaluating machinability is the lathe test developed by Professor O. W. Boston.<sup>6</sup> In this test the specimen is turned on a lathe under standard machining conditions and the length of time required for failure of the cutting tool is measured at various cutting speeds. Curves are then plotted relating tool life to cutting speed for the particular materials being studied.

#### Relation of Machinability of Cast Iron to Microstructure

For a number of years qualitative relationships between microstructure and machinability in the various types of cast irons have been recognized. In the American Society for Testing Materials symposium on cast iron published in 1933,<sup>7</sup> it is pointed out that cast iron varies from one of the most readily machinable to one of the most difficultly machinable of all ferrous alloys. These differences in machinability are directly attributed to the variety of microstructures to be found in the various types of cast iron. Fully annealed permanent mold castings possess excellent machinability because the microstructure consists of iron containing virtually no combined carbon with the matrix structure broken up by a high percentage of finely divided graphite flakes. On the other hand, chilled or white cast iron is extremely difficult to machine because its structure is composed chiefly of hard iron carbide, both in the form of pearlite and as free cementite with the matrix unbroken by any graphite flakes. The wide range of machinability possessed by the several types of cast iron between these two extremes depends upon the proportions of the various microconstituents present and their distribution in the matrix structure.

More recently, Field and Stansbury,<sup>8</sup> Mahin and Lownie,<sup>9</sup> and others have made careful studies of the relation between machinability and other properties of gray cast iron to its microstructure. It has been observed that cast irons having similar physical properties and chemical analyses often differ drastically in their machining characteristics, while in wrought steel there exists a close relationship between microstructure and physical properties. Thus the microstructure of cast irons, being more complex in nature, are not accurately indicated by their physical properties.

ies. In the work of Field and Stansbury this lack of parallel between machinability and physical properties in cast iron led these workers to believe that the microstructure of cast iron rather than tensile strength or hardness might have an orderly influence on the machining properties of the metal.

Previous experiences in the milling of cast iron had shown that two different groups of alloy cast iron test blocks with virtually identical chemical analyses and physical properties gave widely different values of machinability as evidenced by tool life. An examination of the microstructure of these two groups revealed that samples having inferior machinability contained free carbide segregations, while the superior machinability group contained only negligible traces of free carbide. It appeared, therefore, that the presence of large, hard particles of free carbide seriously impaired tool life.

On the other hand, in another investigation it was observed that two alloy cast irons with marked differences in chemical analysis and having been cast in different section thicknesses nevertheless gave almost identical tool life values when subjected to machinability tests. Microscopic examination revealed a striking similarity in their microstructures. These indications of a dependence of tool life on microstructure became the starting point of an important investigation of the effect of microstructure on the machinability of cast iron. The specific object of this investigation was to determine the quantitative relationship between microstructure and machinability and thereby to obtain a reliable index for evaluating the machinability of cast iron. The machinability test, employed by Field and Stansbury was identical to that described by Field and Bullock,<sup>5</sup> the machinability rating for a given specimen being based on tool life measurements.

The presence of free carbides in an otherwise normal ferrite-graphite structure was shown to have relatively little effect on tool life at the lower cutting speeds (up to 200 ft. per min.) but seriously reduced tool life in the higher speed ranges (250 to 1,000 ft. per min.). The complete graphitization of cast iron by annealing resulted in a material containing only ferrite and graphite which exhibited tool life values as much as fifty times greater than that of the normal graphite-pearlite structure. No specific data are given by Field and Stansbury regarding the effects of high phosphorus content upon machinability, although it is recognized that the phosphide eutectic steadite would have a similar effect to that produced by free carbide.

#### Use of Alloying Agents and Inoculants to Improve Machinability in Gray Cast Iron

While maximum machinability in gray iron may be obtained through annealing,<sup>8, 10</sup> it is nevertheless desirable to produce castings having good machining qualities at somewhat lower cost than can be obtained in annealed iron. Consequently, the use of alloying addition agents to accomplish this end has been studied by a number of investigators and a wide variety of alloy treatments has been recommended.

Alloy addition agents for the purpose of improving machinability or otherwise modifying the properties

and structure of cast iron may be divided into two general classes: First, alloying agents that are commonly added in sufficient quantity to produce significant changes in the composition of the iron. In this group may be classed such elements as chromium, nickel, copper, molybdenum, tungsten, vanadium and titanium. In the second group are materials whose addition to cast iron effects drastic changes in its structure and properties although the amount added is not sufficient to appreciably change the composition. Addition agents belonging to this group are commonly spoken of as inoculants. Among the more frequently used inoculants are calcium silicide, ferrosilicon, graphite, and numerous combinations of these materials with other alloying elements such as chromium, manganese, zirconium, molybdenum, and nickel. It should be pointed out that there exists no hard and fast line of demarcation between the first type of alloying elements and inoculants. The distinction lies in the manner in which the addition agent modifies the structure of the resulting iron.

#### Mechanism of Inoculation

According to generally accepted theory,<sup>11</sup> the mechanism of inoculation is primarily a nucleation process by which microscopic or submicroscopic particles of the inoculant become dispersed throughout the molten metal. During solidification, these particles provide nuclei for the formation of normal graphite flakes. In the absence of such nuclei, molten iron frequently becomes super-cooled to a temperature appreciably below its freezing point before graphite separation is initiated. Under such conditions the formation of graphite when it does begin, proceeds quite rapidly. Rapid graphitization then results in the production of extremely fine graphite flakes whose size and distribution is quite variable, depending upon the section thickness and the rate of heat extraction from the molten metal. The addition of an inoculant to the molten metal provides artificial nuclei, which effectively initiate normal solidification thereby preventing super-cooling and its attendant undesirable effects upon the structure and properties of the casting.

Although the literature on inoculation on gray cast iron is voluminous, relatively little information is available on the evaluation of inoculants in terms of specific machinability tests. McElwee and Barlow<sup>12</sup> and others have recognized two types of inoculants, the graphitizing type and the carbide stabilizing type. Graphitizing inoculants contain only those elements known to promote decomposition of iron carbide during the solidification of the iron, i.e. silicon, calcium, carbon, zirconium, etc. Stabilizing inoculants on the other hand contain carbide forming elements, such as chromium or molybdenum, that are capable of developing increased strength and hardness in the matrix structure of the iron. Although both types of inoculants improve uniformity and decrease chilling tendency, it is apparent that the graphitizing type will give rise to a softer microstructure and will consequently be most effective in improving machinability.

Among the more successful graphitizing inoculants,



the silicon-manganese-zirconium type has been extensively studied. Lownie<sup>13</sup> describes the use of an inoculant of this type containing 7 per cent manganese, 7 per cent zirconium, and 75 per cent silicon. Experimental work on the influence of this inoculant and of 95 per cent ferrosilicon upon fluidity, chilling tendency, and other physical properties of cast iron led to the conclusion that very definite advantages are to be derived from the use of this silicon-manganese-zirconium ladle inoculant in the manufacture of high strength gray iron. It was established that 0.35 per cent silicon as a ladle addition either in the form of the silicon-manganese-zirconium alloy or as ferrosilicon was an optimum addition quantity.

#### Inoculating Agents

One of the more extensively used inoculants in gray iron production is calcium silicide. As reported by A. F. Meehan in his basic patents<sup>14</sup> it was found that the addition of a silicide of an alkaline earth metal, such as calcium or magnesium, will materially improve gray iron castings. The introduction of such a silicide in the molten metal acts to precipitate additional carbon in spite of which the tensile strength and transverse strength will be materially increased. A well-known proprietary process consists essentially of calcium silicide inoculation of a suitable gray cast iron composition which, through control of graphitization, gives rise to a uniformity of structure and improvement of physical properties which are less readily obtainable by other methods.

Another inoculating agent that has received considerable attention during the past 10 or 12 years is silicon carbide. Brown<sup>15</sup> recommended the use of silicon carbide treatments for gray cast iron in which its addition was made as a part of the cupola charge. Used in this manner, it is claimed that 10 to 20 lb. of silicon carbide per ton of metal will increase the pouring temperature due to an exothermic reaction and will enable up to 100 per cent scrap metal to be charged without impairing casting qualities. It is further claimed that silicon carbide treatment will stabilize the pearlitic microstructure so that a wider range of section thicknesses will freeze as pearlite with minimum amount of either free cementite or free ferrite.

Boyer<sup>16</sup> after conducting further investigations upon the use of silicon carbide addition to cast iron reported that its influences are found to be considerably greater than would be expected from direct effects of the resulting increases in silicon and carbon contents of the metal. Additions of from 1.0 to 1.5 per cent of silicon carbide briquettes to the cupola charge were found to result in the breaking up of massive carbide and phosphide networks in the microstructure in such a manner as to greatly improve machinability.

Loria and Thompson<sup>17, 18</sup> have continued the study of silicon carbide addition to gray cast iron. In these studies no radical departures are made from the original procedure of adding silicon carbide in the form of briquettes of the finely powdered material to the cupola charge. Studies of the influence of silicon carbide treatments upon the chilling tendency in high

carbon gray irons and upon hardness, transverse strength, porosity, and machinability of automotive cylinder iron have indicated definite improvements in each of these properties. Unfortunately, none of the literature on silicon carbide treatments gives the results of specific tests for evaluating the influence of the treatment on machinability.

Among the alloying agents suggested for improving the machinability of gray cast iron the element titanium has received considerable attention. It has been reported<sup>19</sup> that when titanium in the form of 45 per cent ferrotitanium is added to gray iron in amount sufficient to produce between 0.08 per cent and 0.25 per cent residual titanium content, the hardness is usually increased slightly due to its effect in reducing the size of the graphite flakes. In spite of increased hardness, titanium in this range of composition is reported to improve the machinability of cast iron both by eliminating the occurrence of hard, chilled structures at thin sections or corners, and by general improvement in the graphite flake structure and distribution.

#### Influence of Titanium

Starkweather<sup>20</sup> lists specific data to illustrate the influence of titanium treatment upon the machinability of alloy cast irons containing chromium and molybdenum. The machinability ratings are based upon experiments employing the tool dynamometer technique developed by O. W. Boston<sup>21</sup> in which the cutting force is measured under standard conditions of machining on a lathe. It was found that titanium treated gray iron showed appreciable improvement in machinability over the untreated iron in spite of the fact that all the samples showed practically identical tensile strength and Brinell hardness.

The data reported by Starkweather for this improvement in machinability are based on a 1.0 per cent addition of ferrotitanium to a base iron of the following analysis and properties:

Total carbon, per cent.....	3.4
Silicon, per cent.....	2.21
Combined carbon, per cent.....	0.72
Chromium, per cent.....	0.54
Ultimate tensile strength, psi.....	38,700
Brinell hardness number.....	207

Figures listed for comparative resistance to cutting in the lathe test are: 22.5 for the titanium treated iron as compared to 28.0 and 27.0 for the untreated iron under a condition of 0.003-in. feed. In a similar lathe test at 0.007 in. feed the comparative resistance to cutting is listed as 37.5 for the titanium treated iron as compared to 41.0 and 39.0 for the untreated iron.

A somewhat different modification of the use of titanium as an alloying agent in gray iron is presented by Holzworth<sup>22</sup> in which titanium is introduced into the iron by smelting a suitable mixture of titanium ores and iron ores in the blast furnace. In accordance with this patented process the blast furnace is so operated that the temperature reached in the reduction zone and in combustion zone is high enough to reduce titanium ores such as rutile or ilmenite. A practical optimum range for titanium content in the pig iron is between 0.30 and 0.50 per cent.

Employing this process it is claimed that "the finely divided character of the graphitic carbon and its uniform distribution in the pig iron carries through to and persists in castings made therefrom, and while the casting is slightly higher in Brinell hardness because of its relatively fine pearlitic structure, it is a more machinable casting, having a fine tight grain and one which increases tensile and transverse strength."

An important recent development in the improvement of machinability of gray cast iron has been reported by the British Cast Iron Research Association. According to Hartley<sup>23</sup> nodular or spherulitic graphite (similar to the temper carbon of malleable iron) may be produced in as-cast gray iron by a specialized treatment. Recent research by British foundrymen has led to the development of a revolutionizing process for the production of gray cast iron having remarkably improved physical properties and ductility. This material, although not entirely equivalent to malleable iron, possesses a measurable elongation on the order of 1 to 2 per cent.

Although the details of the newly developed process remain confidential at this writing it is reported that a heat of hematite pig iron of 3.9 per cent carbon and 2.6 per cent silicon upon treatment gave ultimate tensile strength values of 15.5 and 26.5 tons per square inch on a 0.875-inch diameter test bar. Excellent reproducibility of physical properties and freedom from defect are claimed for the new iron.

Phosphoric cast irons may be treated by this method. However, it is pointed out that the special effects resulting in nodular or spherulitic graphite do not offset the dominating influence of phosphorus upon the iron. While definitely improved machinability is reported, no specific data concerning machinability tests have been made available as yet.

#### Experimental Methods

The experimental phases of this investigation consisted of a systematic study of the influence of a number of alloy additions and treatments upon the properties and microstructure of a selected base iron. The gray iron chosen for this purpose conformed (except for its phosphorus content) to Specification No. 110, "Automotive Soft Iron," of the Society for Automotive Engineers (Table 1).

TABLE 1—COMPARISON OF BASE IRON WITH S.A.E. SPECIFICATION NO. 110<sup>24</sup>

S.A.E. No. 110	Min.	Max.	Base Iron
Total carbon, per cent	3.40	3.70	3.54
Silicon, per cent	2.30	2.80	2.56
Manganese, per cent	0.50	0.80	0.69
Sulphur, per cent	...	0.12	0.069
Phosphorus, per cent	...	0.25	0.65

The selection of S.A.E. Specification No. 110 was made because of the relatively great demand by the automotive industry for readily machinable castings of this composition. It was felt that significant improvement in the machinability of a high phosphorus

iron of this type would point the way to increased utilization of southern pig iron in automotive castings, a field now definitely restricted to low phosphorus irons.

Each of the several alloy addition agents and treatments was studied by preparing a series of laboratory induction furnace heats of the base iron incorporating varying proportions of the alloying agent under consideration. Specimens from these heats were subjected to physical and metallographic examination as well as to chemical analysis in order to evaluate the effects of the treatments.

The melting equipment used consisted of a 20 kw mercury gap frequency converter equipped with a 12-lb. tilting type induction furnace. A melting procedure was established whereby 10-lb. heats of the selected base iron were melted under standard conditions and predetermined additions of the alloying materials were made.

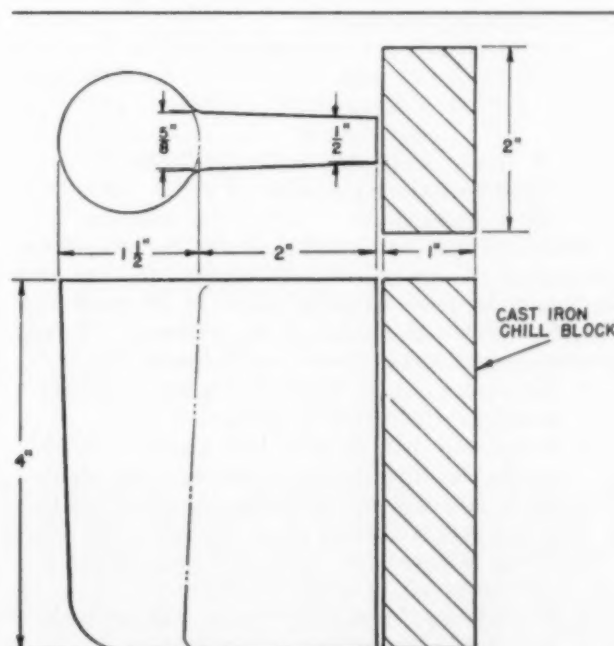


Fig. 1—Keyhole type chill test specimen showing relative position and dimensions of chill block.

Test specimens consisting of "keyhole" type chill wedges (Fig. 1) were cast in dry sand molds after allowing a suitable holding time for homogenization of the melt. These specimens were fractured for determination of chill depth. The broken pieces then provided ideal samples for metallographic examination and tests for hardness, machinability, and chemical analysis.

For metallographic study a segment of the circular cross section of the specimen was cut so as to include a portion of the cylindrical section farthest from the chilled zone. The metallographic examination included the following: an evaluation of the graphite structure according to ASTM designation A-247-41T,<sup>25</sup> an approximation of the relative proportions of pear-

ite and ferrite in the matrix structure, and an appraisal of the mode of occurrence of the steadite. Photomicrographs were made as required, to illustrate important trends in the results. Hardness values were derived by making a hardness survey of the circular cross section of each specimen from its center outward to the surface at a point diametrically opposite the chilled section (Fig. 1). The average of these readings was reported as the hardness of the specimen.

In order to evaluate the effectiveness of the various treatments in improving machinability, it was necessary to conduct a suitable machinability test in addition to the examinations and chemical analyses. For this purpose, a machinability test patterned after the Keep and Loren drill penetration test<sup>4</sup> was selected.

A standard model floor drill was modified so that the following test conditions could be consistently maintained throughout the entire series of tests:

1. Drill size:
  - Pilot drill..... $\frac{1}{8}$  in.
  - Test drill..... $\frac{3}{8}$  in.
2. Drill point type:<sup>26</sup>
  - Standard two fluted twist drill
  - Point angle..... $118^\circ$
  - Lip clearance..... $12^\circ$
  - Cutting angle..... $135^\circ$
3. Drill point pressure.....50 lb.
4. Revolutions per min.....487
5. Cutting fluid.....None

The drill test was conducted on the cylindrical portion of the fractured chill wedges. In each case the test holes were carefully placed in identical locations in the circular section of the specimen. The test procedure consisted of the following specific steps:

1. Drill pilot hole to depth of  $1\frac{1}{2}$  in. using care to keep drill in vertical alignment.
2. Locate test drill on pilot hole and with dead-weight loading device operating, start drill.
3. After test drill has penetrated approximately  $\frac{1}{4}$  in., begin timing; clock accurately the interval required for test drill to penetrate exactly one inch, or other specified depth.
4. Record rate of penetration in seconds per inch, and convert to millimeters per minute.

The constancy of conditions, including drill sharpness, was checked before and after testing each specimen by making a series of tests on a standard test bar of predetermined machinability. It was found that maintenance of drill sharpness was a critical factor. However, with reasonable care in sharpening, using a precision drill grinding fixture, it was possible to obtain satisfactory reproducibility. The test drill was resharpened whenever the rate of penetration in the standard test bar fell below a predetermined limit.

Under the test conditions outlined above, typical penetration ratings ranged from 50 mm. per min. for the untreated base iron to slightly over 100 mm. per min. for fully annealed specimens, thus providing an adequate spread of values for comparison of the various specimens.

#### Discussion of Results

Three fundamental methods of approach have been followed in investigating the problem of improving

machinability of high phosphorus gray irons. They may be summarized as follows:

First, modification of the mode of occurrence of the phosphide constituent in the microstructure. This approach has as its object overcoming the abrasiveness of iron phosphide by changing its structure, properties, and composition through alloy additions.

Second, control of microstructure exclusive of the phosphide constituent. The object of this method of attack is to produce a soft ferritic matrix structure instead of the relatively hard pearlitic matrix normally characteristic of high phosphorus gray irons. A structure of this type would exert less abrasive or dulling action on cutting tools during machining operations.

A third approach to the problem deals with outright removal of constituents known to have adverse effects on machinability. Desulphurization and dephosphorization treatments of the molten iron, and combinations of these treatments with alloy additions are included under this heading.

A total of 76 experimental heats were prepared in studying these three methods of approach, and although full discussion of the results of this work will not be attempted in this condensed report, several important observations deserve consideration.

#### Modification of Mode of Occurrence of Steadite

In investigating the possibility of modifying the mode of occurrence of steadite, it was desirable to evaluate and compare the hardness of the various constituents occurring in the microstructure of high phosphorus gray iron. In this connection, microhardness measurements were made on typical specimens of the iron under study. For this work the Knoop microhardness test<sup>27</sup> was employed. Table 2 summarizes the results obtained.

TABLE 2—MICROHARDNESS OF STRUCTURAL CONSTITUENTS OF GRAY CAST IRON

Constituent	Knoop Hardness No. <sup>a</sup>
Cementite	1000-1100
Steadite (massive)	600-900
Steadite (open honeycomb)	500-600
Pearlite	350-450
Silico-Ferrite (2.0% Si)	150-250

<sup>a</sup> 50 gm load on standard Knoop indenter

It was observed that the hardness of the phosphide constituent, steadite, is somewhat variable depending upon its physical constitution. Microhardness measurements revealed that massive steadite particles having a dense structure approach the hardness of free cementite, while the more open honeycomb type structure is generally somewhat softer. In any event, the hardness of steadite in comparison to that of the matrix structure is sufficiently high to account for its extreme abrasive action in high speed machining operations. This is especially true when steadite occurs as a practically continuous network throughout the structure of the iron, as is often the case in irons containing in excess of 0.60 per cent phosphorus.

Fig. 2—A  
phorus be  
in 2.5 pe  
essentially



On the basis of the foregoing facts it is clear that a modification in the mode of occurrence of steadite in high phosphorus gray iron would produce considerable improvement in tool life machinability provided either of the following changes could be effected:

- Reduction in steadite particle size, by breaking up of continuous network formations.
- Reduction in hardness of the phosphide constituent through changes in its composition. Such changes may involve simple decomposition of iron carbide in the ternary iron-iron phosphide-iron carbide eutectic; or they may involve the formation of phosphides of elements other than iron through the addition of alloying agents.

A series of experimental melts was prepared to determine whether or not appreciable modifications of the mode of occurrence of steadite could be produced through alloy additions to the selected base iron. For this purpose additions of aluminum, calcium, titanium and zirconium were studied.

The results of this series of experimental heats were unpromising in that no significant change could be detected in the microstructural occurrence of steadite in the treated specimens. It may be concluded, nevertheless, that iron phosphide in the form of the eutectic, steadite, is a highly stable constituent of gray cast iron. Moreover, the law of mass action preponderantly favors the formation of iron phosphide in preference to the phosphide of any element introduced as an addition agent. In view of these facts it is believed that it is not practicable to alter significantly the mode of occurrence of the phosphide constituent in high phosphorus gray cast irons through additions of mod-

erate amounts (up to 2.0 per cent) of the alloying elements studied.

#### Control of Microstructure Exclusive of Steadite

Since the nature of the matrix structure has an important bearing on the machining qualities of cast irons, several treatments have been evolved for improvement of machinability through modification or control of microstructure. One of the most effective procedures of this type is annealing or normalizing of castings prior to machining. Aside from annealing, the control of microstructure in as-cast gray iron for improvement of machinability may be brought about by adjusting the composition so as to favor graphitization of combined carbon. This may be accomplished to a certain degree by increasing carbon and silicon contents insofar as is compatible with satisfactory casting qualities, and by maintaining a proper balance between manganese and sulphur in order to inhibit the carbide stabilizing tendency of sulphur. Somewhat more effective improvement in machinability in as-cast gray irons may be attained through the addition of inoculants or alloying agents known to favor carbide decomposition.

Experimental studies of each of the above methods of microstructural control have been undertaken with particular emphasis on their effectiveness in improving machinability in the selected high phosphorus base iron. A comparative analysis of the results obtained reveals that annealing is definitely the most effective treatment for obtaining the desired ferritic microstructure. Unfortunately, annealing is also the most expensive treatment.

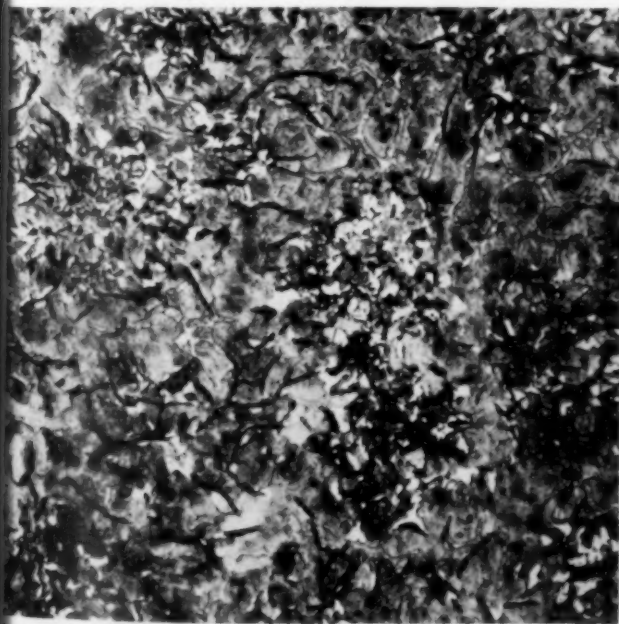


Fig. 2—As-cast structure of the 0.65 per cent phosphorus base iron studied in this investigation. Etched in 2.5 per cent nital, magnification 200X. Matrix is essentially pearlitic (gray background) with dendritic network of steadite.

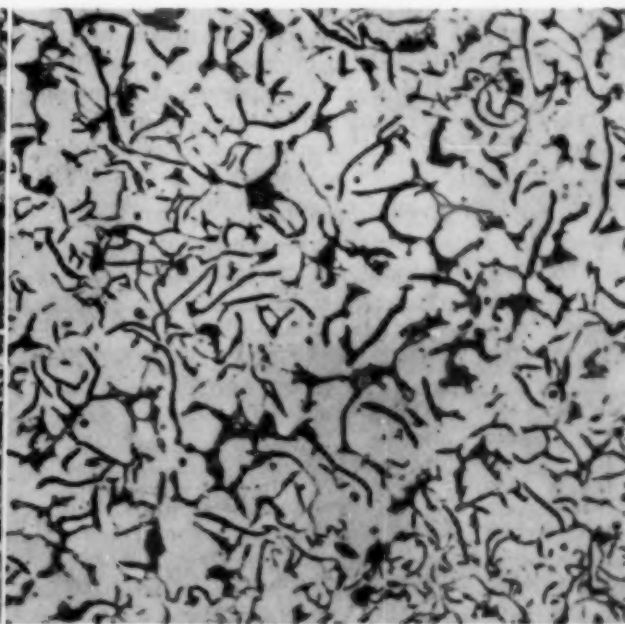


Fig. 3—Fully annealed structure of the same specimen shown in Fig. 2. Etched in 2.5 per cent nital, magnification 200X. Matrix is entirely ferritic (light background) with steadite network unchanged.

In spite of high costs, annealing or normalizing procedures are widely used for improving machinability, especially in small castings where relatively rapid cooling of thin sections is likely to give rise to chilled corners or other areas containing free cementite. The usual annealing treatment, where free cementite is known to be present, involves heating to above the critical range, 1500-1600 F (815-870 C), following by slow cooling. In pearlitic irons containing no free cementite, a sub-critical anneal at 1200-1400 F (650-670 C) followed by cooling in still air will produce satisfactory softening for machining.

Drill penetration machinability tests were conducted on as-cast and fully annealed specimens (Figs. 2 and 3) of the base iron in order to evaluate the actual improvement in machinability resulting from annealing. Table 3 gives the results of these tests.

TABLE 3—EFFECT OF ANNEALING AT 1400 F (760 C) ON PROPERTIES AND STRUCTURE OF BASE IRON

	As-Cast	Annealed
Total carbon, per cent	3.52	3.52
Combined carbon,* per cent	0.80	0.00
Silicon, per cent	2.51	2.51
Manganese, per cent	0.56	0.56
Phosphorus, per cent	0.62	0.62
Hardness, Rockwell "B"	93	75
Drill test machinability, mm/min	53.5	109.0
Approximate per cent pearlite in matrix structure	95	0.00

\* Microscopic estimation

It is pointed out that there is a marked improvement in drill test machinability corresponding to the transition in matrix structure from pearlite to ferrite. The results of this experiment confirm the initial premise that microstructural control to produce a ferritic matrix will result in improved machinability. The data obtained also provide a means of comparative evaluation of machinability ratings resulting from subsequent experimental treatments and alloy additions. Although the cost of annealing is high (\$10 to \$15 per ton of finished castings under the most

favorable conditions), it nevertheless represents a highly satisfactory means of improving machinability, and as such provides a goal for attainment by other less expensive treatments.

#### Control of Microstructure Through Control of Composition

Since the microstructure of gray cast iron is closely related to its composition, the possibility of attaining the desired ferritic microstructure through direct control of composition presents a simple, straightforward approach to the problem of machinability. Unfortunately, several disturbing factors enter the picture to prevent the achievement of this end directly.

Due to the fact that the formation of ferrite in cast iron presupposes the graphitization of combined carbon, it is obvious that in unalloyed gray irons higher percentages of those elements known to promote carbide decomposition (i.e., carbon and silicon) will tend to favor solidification in the stable iron-graphite system, thus producing the desired ferrite-graphite microstructure.

On the other hand, rapid cooling, as normally encountered in small thin-sectioned castings, acts strongly to counteract the graphitizing effects of high carbon and silicon contents. Another noteworthy factor tending to inhibit graphitization in irons where late additions are employed to adjust the composition is the inoculating effect of certain addition agents. In addition to the effects of rapid cooling and inoculation, the presence of sulphur in excess of about 0.01 per cent also exerts a definite stabilizing influence upon combined carbon. Consequently, due to the combined influence of these factors, it is not altogether feasible to produce a ferritic matrix in gray iron through simple adjustment of carbon and silicon contents.

Nevertheless, for the purpose of obtaining comparative data on machinability and microstructure, a study was made of the effects of variations in carbon and silicon contents of the base iron as produced by the three following addition agents: ferrosilicon, silicon carbide, and a proprietary mixture of graphite and ferrosilicon. Table 4 gives a summary of the results obtained through additions of varying quantities of

TABLE 4—EFFECTS OF INCREASING CARBON AND SILICON CONTENTS OF BASE IRON

Specimen No.	Composition, per cent						Chill Depth, In.	R <sub>B</sub>	DTM <sup>c</sup>	%F <sup>d</sup>	% Addition	Addition Agent
	T.C.	C.E. <sup>a</sup>	Si	Mn	S	P						
7	3.18	4.29	2.79	0.55	0.04	0.91	0.12	96	52	0-5	0.0	Base iron
14	3.17	4.39	3.17	...	...	0.89	0.15	100	...	10-15	6.5	15% Ferrosilicon
13	3.43	4.64	3.13	...	...	0.92	0.00	88	69	10-20	2.5	Silicon Carbide
16A	3.39	4.92	4.21	...	...	0.91	0.00	77	...	90-95	5.0	Silicon Carbide
18-0	3.36	4.43	2.51	0.52	0.04	1.08	0.12	91	...	10-15	0.0	No additions
18-1	3.47	4.50	2.77	0.53	...	1.01	0.00	86	...	20-30	0.95	Graphite and Ferrosilicon
18-2	3.49	4.72	3.06	0.55	...	1.04	0.00	81	71	50-60	1.8	Graphite and Ferrosilicon
18-3	3.56	4.80	3.14	0.55	...	1.00	0.00	81	...	70-80	2.6	Graphite and Ferrosilicon

<sup>a</sup> C.E. = Carbon equivalent, total carbon plus 0.3 (Si + P)

<sup>b</sup> Hardness Rockwell "B" scale 1/16-in. steel ball penetrator 100 kg load.

<sup>c</sup> Drill Test Machinability, millimeters penetration per minute.

<sup>d</sup> Estimated per cent ferrite in matrix structure.

these materials to experimental melts of the base iron.

The effects of inoculation and relatively rapid cooling rates in overcoming the graphitizing influence of high carbon and silicon contents are apparent from the proportion of ferrite in the various specimens listed in Table 4. In Specimen No. 16A, although the silicon content was 4.2 per cent with total carbon at 3.39 per cent, metallographic examination of the 1½-in. diameter cylindrical section showed appreciable quantities of combined carbon in the form of pearlite. This is conclusive proof that a completely ferritic matrix is not readily produced in light castings of normal gray iron analyses simply through adjustment of total carbon and silicon contents.

On the other hand, significant softening of the base iron with corresponding reduction in chilling tendency and improvement in drill test machinability was accomplished through additions of silicon carbide and the proprietary mixture of graphite and ferrosilicon. With either of these materials the introduction of sufficient carbon and silicon to give approximately 3.4 per cent total carbon and 3.1 per cent silicon in the finished casting produced significant improvement in machinability.

It should be pointed out that the use of silicon carbide as a ladle addition to cast iron is not recommended. According to its manufacturers, maximum effectiveness is achieved through the addition of silicon carbide briquettes as part of the cupola charge.

#### Control of Microstructure Through Ladle Additions

The utilization of alloy additions and inoculants for the purpose of improving the properties and structure of gray cast iron has been studied exhaustively both in this country and abroad. The majority of this work has been directed toward improvements in mechanical properties in general and a portion of it has dealt specifically with improvements in machinability. Nevertheless, essentially no references are to be found in the literature regarding the use of alloy additions for the specific purpose of producing a ferritic matrix in gray cast iron. The primary reason for this is the fact that ferrite, being a soft low strength constituent, causes undesirable reductions in strength and other properties of the iron; hence its presence is normally avoided.

However, in many applications where service stresses are low, good machinability is of greater importance than high mechanical properties. For such applications the use of alloy additions for the purpose of producing a soft, easily machinable structure is considered entirely justifiable. On this basis four series of experimental heats were prepared for the purpose of studying the effects of certain alloy additions upon the structure and machinability of the selected high phosphorus base iron.

The number of elements known to favor the decomposition of combined carbon in gray iron is relatively small; they include silicon, carbon, calcium, aluminum, copper, nickel, titanium, zirconium, and in small quantities, manganese. Of these, previous experiments had indicated a lack of noteworthy effect on the base iron in all save copper, nickel, zirconium

and manganese. Since copper and nickel are known to be rather weak graphitizers, it was decided to investigate the effects of additions of zirconium in the form of certain zirconium alloys, and alloy combinations. For this purpose, additions of zirconium in combination with silicon, manganese, and aluminum were made employing the following alloying agents:

1. 37 per cent zirconium, 50 per cent silicon, 10 per cent iron (37 per cent zirconium-silicon).
2. 60 per cent silicon, 6 per cent manganese, 5 per cent zirconium, balance iron.
3. 60 per cent silicon, 6 per cent manganese, 6 per cent zirconium alloy (as above) plus commercially pure aluminum.
4. Mixed alloys for studying various proportions of silicon, manganese, and zirconium.

Four series of 10-lb. induction furnace heats prepared according to the previously described melting procedure were treated with varying quantities of these addition agents.

The results of the experimental series employing 37 per cent zirconium, 50 per cent silicon alloy revealed a definite graphitizing action on the high phosphorus base iron, with corresponding improvement in drill test machinability. For example a 2.0 per cent addition of the 37 per cent zirconium alloy gave rise to a residual zirconium content of 0.18 per cent and the drill test machinability rating was increased from 52.0 mm per min. in the untreated base iron to 64.3 mm per min. in the treated specimen.

However, since a 2.0 per cent addition of 37 per cent zirconium alloy is equivalent to 0.74 per cent added zirconium, the 0.18 per cent residual zirconium content indicates a recovery of less than 25 per cent of the zirconium added. Due to this poor recovery the use of the 37 per cent zirconium alloy was not considered economically practicable.

In the second series of experimental heats zirconium alloy additions were made in the form of a 60 per cent silicon, 6 per cent manganese, 5 per cent zirconium alloy. It was found that the addition of zirconium in this form offers several advantages. Being substantially lower in zirconium content the alloy is less susceptible to oxidation losses; the percentage recovery was observed to be approximately 75 per cent instead of 25 per cent as before. Moreover the presence of manganese enhances the graphitizing effect by counteracting the carbide stabilizing tendency of sulphur in the iron. In addition to these advantages this 5 per cent zirconium alloy is appreciably cheaper than the 37 per cent zirconium alloy.

#### Effects of Si-Mn-Zr Additions

The effects of additions of 60 per cent silicon, 6 per cent manganese, 5 per cent zirconium alloy are shown graphically in Fig. 4, 5, and 6. The close correlation of hardness, drill test machinability and microstructure is clearly illustrated in these curves. A rapid change in each of the properties, hardness, machinability, and per cent ferrite in the matrix, is noted with additions of between 1.0 and 2.0 per cent, while a distinct leveling off occurs with additions of more than 2.0 per cent of the alloy.



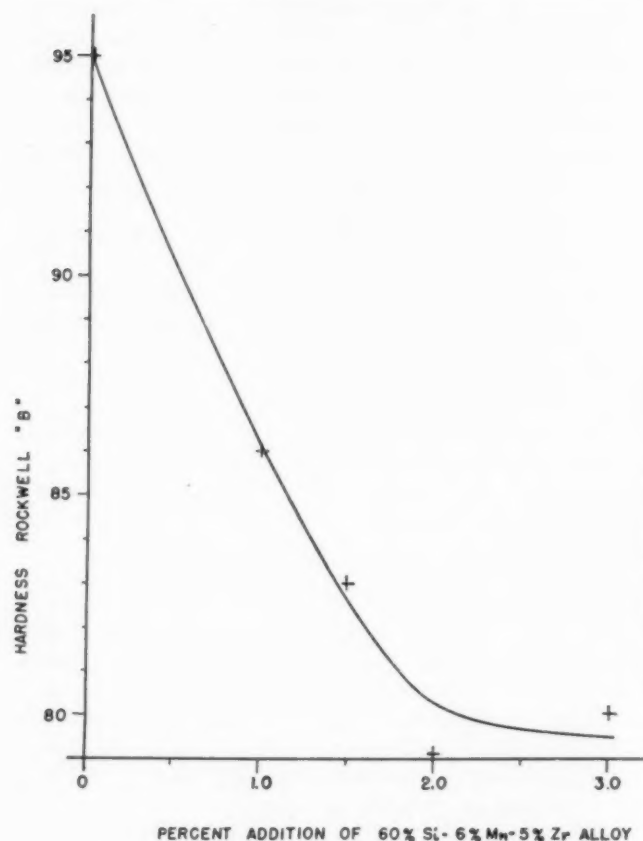


Fig. 4—Effect of increasing additions of 60 per cent Si, 6 per cent Mn, 5 per cent Zr alloy on Rockwell hardness of the 0.65 per cent phosphorus base iron.

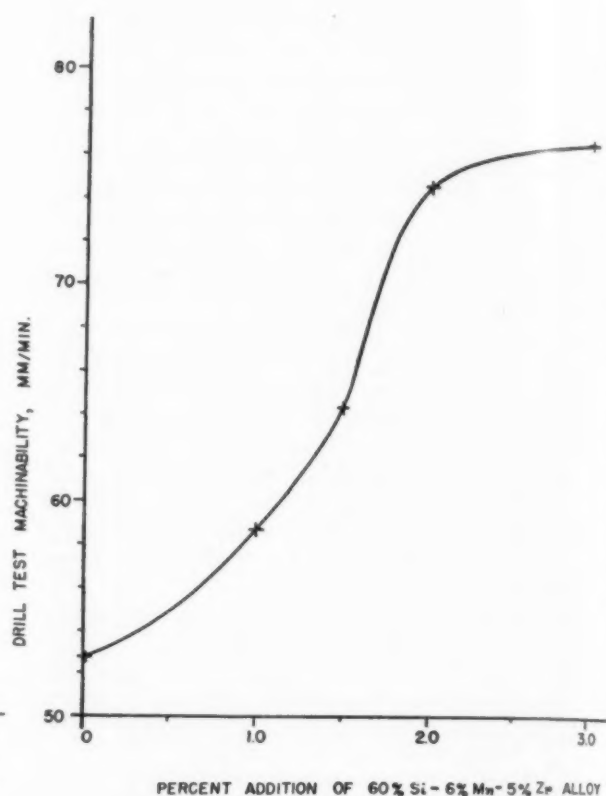


Fig. 5—Effect of increasing additions of 60 per cent Si, 6 per cent Mn, 5 per cent Zr alloy on proportion of ferrite in matrix structure of the 0.65 per cent phosphorus base iron.

TABLE 5—TYPICAL ANALYSIS OF 0.65 PER CENT PHOSPHORUS BASE IRON FOLLOWING 2.0 PER CENT ADDITION OF 60 PER CENT SILICON, 6 PER CENT MANGANESE, 5 PER CENT ZIRCONIUM ALLOY

Element	Per Cent
Total Carbon	3.44
Silicon	3.62
Carbon Equivalent <sup>a</sup>	4.72
Manganese	0.69
Sulphur	0.04
Phosphorus	0.65
Zirconium	0.09

<sup>a</sup> Carbon equivalent = T.C. + 0.3 (Si + P)

Thus it is seen that satisfactory improvement in the machinability of the high phosphorus base iron may be attained through a minimum addition of 2.0 per cent of this alloy. It must be pointed out that the high silicon content and the high carbon equivalent of iron treated in this manner (Table 5) place it in the hypereutectic range. Irons of this type are characterized by sub-normal casting qualities. Their fluidity is poor and they are rather susceptible to shrinkage. Consequently, further modification of the treatment and/or addition agent were considered nec-

essary before a satisfactory treatment could be defined.

In an effort to establish such a modification that would permit the introduction of a smaller quantity of silicon into the melt, a study of the effects of simultaneous additions of the silicon-manganese-zirconium alloy and aluminum was made. In another series combined additions of several alloys containing varying proportions of silicon manganese and zirconium were studied, the objective being to determine the optimum proportions of these elements for maximum improvement in machinability. However, comparison of the results of these experiments indicated that the original 2.0 per cent addition of the 60 per cent silicon, 6 per cent manganese, 5 per cent zirconium alloy as described above produces greater improvement in drill test machinability than could be obtained in any of the modified treatments.

Experimental results at this stage of the investigation indicated that, aside from the well recognized expedient of annealing, significant improvement in the machinability of the high phosphorus base iron may be attained through a combination of high carbon and silicon contents with the introduction of approximately 0.10 per cent zirconium. The presence of sufficient manganese to adequately counteract the carbide stabilizing tendency of sulphur is also recog-

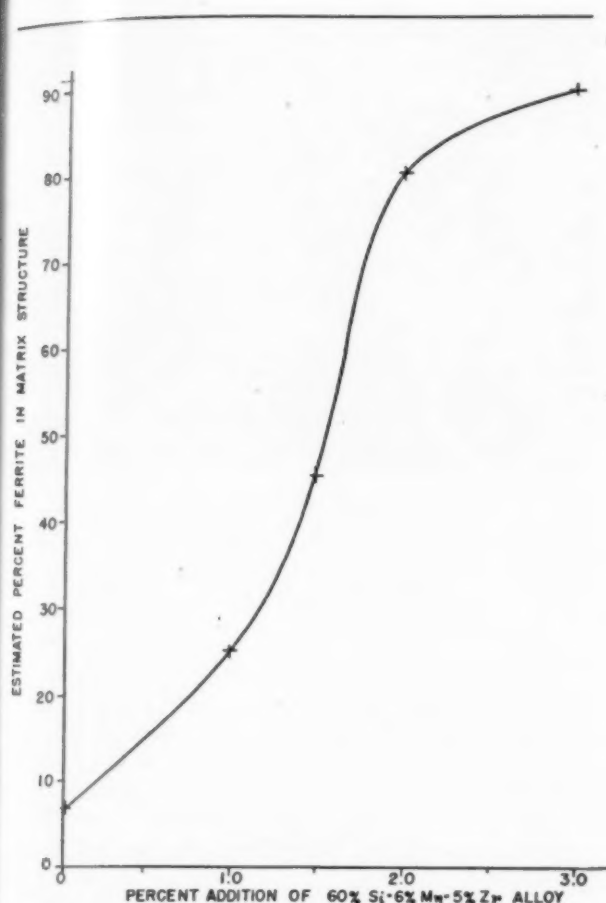


Fig. 6—Effect of increasing additions of 60 per cent Si, 6 per cent Mn, 5 per cent Zr alloy on proportion of ferrite in matrix structure of the 0.65 per cent phosphorus base iron.

nized as a significant factor in the treatment. The outstanding importance of control of sulphur content is covered more fully in the section immediately following.

#### Control of Microstructure Through Desulphurization and Alloy Additions

In reviewing the literature it was observed that Boyles<sup>28</sup> reported the occurrence of a soft easily machining structure of ferrite and fine graphite in a synthetic cast iron having extremely low sulphur and manganese contents. According to Boyles, the so-called normal structure consisting of pearlite and flake graphite in gray irons is due largely to the presence of appreciable quantities of these two elements. Cast iron in the absence of sulphur and manganese was found to revert to a highly abnormal structure.

In view of these findings and their implications regarding the prospect of improved machinability, a study was made of the effects of desulphurization alone, and desulphurization plus alloy additions, upon the structure and machining qualities of the base iron.

The desulphurization of cast iron through treatment of the molten iron with a sodium carbonate slag has been carried out as a routine procedure in commercial foundries for many years. For this purpose additions of fused sodium carbonate on the order of five to ten pounds per ton of molten iron are commonly employed in a semi-continuous process treatment.

A preliminary test was made to determine the effectiveness of this treatment under laboratory melting conditions (i.e., a small scale batch process as opposed to the large scale continuous foundry process). It was shown that initial sulphur contents on the order of 0.06 per cent could be reduced to 0.010 to 0.015 per cent by holding the molten iron at 2500 F under a sodium carbonate slag equal to 2.5 per cent of the cent of the weight of metal for a period of ten minutes.

In order to obtain data on the effects of low manganese simultaneously with low sulphur contents, several "synthetic" heats were prepared by melting appropriate quantities of Armco ingot iron, ferrosilicon and carbon black. In this manner it was possible to obtain a gray iron containing the usual quantities of carbon and silicon but with manganese and sulphur

TABLE 6—EFFECTS OF SODIUM CARBONATE DESULPHURIZATION AND ZIRCONIUM ALLOY ADDITIONS ON COMPOSITION AND MACHINABILITY

Specimen No.	Composition, Per Cent							Hardness <sup>b</sup>	DTM <sup>c</sup>	Chill Depth, In.	Remarks
	T.C.	C.E. <sup>a</sup>	Si	Mn	S	P	Zr				
63-0	3.41	4.34	2.48	0.59	0.059	0.61	...	B-95	50.2	0.12	Remelted base iron, no additions
70-1	3.47	4.32	2.23	0.58	0.014	0.63	0.116	B-88	70.6	0.37	Desulphurized plus 1.0% addition
72-1	3.48	4.31	2.17	0.55	0.010	0.63	0.093	B-90	62.3	0.43	Desulphurized plus 0.75% addition of 13% Zr-Si Alloy
74-1	3.53	4.41	2.36	...	0.014	...	0.109	B-90	65.7	0.10	Same as above, 1.0% added
75-2	...	...	...	...	...	...	0.112	B-84	76.2	0.10	Same as above, 1.0% added
76-2	...	...	...	...	...	...	0.100	B-85	72.6	0.12	Same as above, 1.0% added

<sup>a</sup> Carbon Equivalent = T.C. + 0.3 (Si + P)

<sup>b</sup> Rockwell "B" Scale 1/16-in. Ball Penetrator, 100 kg Load

<sup>c</sup> Drill Test Machinability, millimeters penetration per minute

<sup>d</sup> Zirconium Ferrosilicon, Zr-13%, Fe-43%, Si-40%

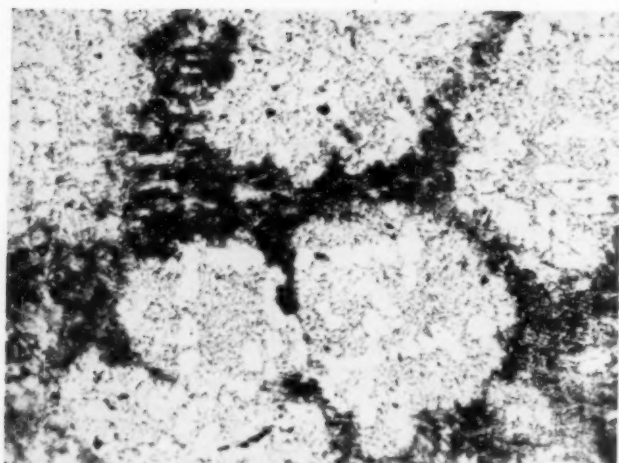


Fig. 7—Structure of "synthetic" low sulphur, low manganese iron with 1.0 per cent addition of 37 per cent zirconium ferrosilicon. Etched in 2.5 per cent nital, magnification 100X.

contents on the order of 0.02 per cent to 0.03 per cent each. It was found that specimens of this "synthetic" iron containing 0.019 per cent sulphur and 0.030 per cent manganese possessed a microstructure of normal graphite in a pearlitic matrix. However, metallographic examination revealed that the addition of 1.0 per cent of 37 per cent zirconium-silicon alloy to this iron gave rise to the desired ferrite-fine graphite structure throughout the entire cross-section (Fig. 7). Specimens treated in this manner showed higher machinability ratings than any as-cast samples yet produced (i.e., drill penetration values of 80 to 85 mm per min. as compared to 50 mm per min. in the untreated base iron).

Finally, experimental heats were prepared with zirconium alloy additions. The results of these experiments showing the effects of desulphurization and zirconium alloy additions are summarized in Table 6.

From the standpoints of economy, as well as properties of the treated iron, the combined desulphurization-zirconium alloy addition treatment is definitely superior to the straight silicon-manganese-zirconium alloy addition previously described. With carbon equivalent at the eutectic or slightly hypereutectic, the casting qualities and structure of the iron are far more likely to be satisfactory. Consistent improvements in drill test machinability of 40 to 45 per cent over that of the untreated iron were obtained following the addition of 1.0 per cent of zirconium ferrosilicon. The microstructure of the base iron following this treatment is shown in Fig. 8.

The combined influences of low sulphur content and zirconium additions in producing the desired effect upon the microstructure and machinability of the base iron may be attributed to several factors. First of all, the elimination of a large portion of the sulphur, a known carbide stabilizer, enhances the metastability of combined carbon in the iron. This naturally allows the normal graphitization reaction,  $\text{Fe}_3\text{C} \rightarrow 3\text{Fe} + \text{C}$ , to proceed with considerably less external stimulation.

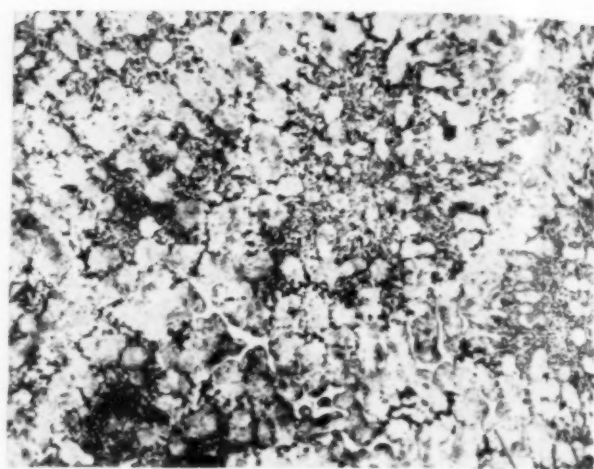


Fig. 8—Microstructure of 0.65 per cent phosphorus base iron following combined desulphurization-zirconium alloy addition treatment. Etched in 2.5 per cent nital, magnification 100X.

The addition of zirconium further diminishes the effect of sulphur through formation of inert zirconium sulphide ( $\text{ZrS}_2$ ) in the melt. The formation of zirconium sulphide in ferrous alloys was verified by Feild<sup>29</sup> who showed that zirconium combines with oxygen, nitrogen, and sulphur, and also neutralizes to a certain extent the embrittling effect of phosphorus in steels.

The refining action of zirconium upon the graphite structure of gray iron is also believed to be related to the formation zirconium carbide in the melt. According to Morrogh<sup>30</sup> it has been shown that the addition of small quantities (less than 0.10 per cent) of zirconium to gray iron causes the formation of a complex sulphide phase. When 0.10 to 0.20 per cent residual zirconium is present zirconium carbide ( $\text{ZrC}$ ) appears. In other words as long as an excess of sulphur is present zirconium additions tend to form zirconium sulphide in preference to the carbide. Once an excess of zirconium is established over and above the amount required to form zirconium sulphide, zirconium carbide is produced, and zirconium only seems to exert a positive refining effect when this carbide phase is present. This accounts nicely for the effects observed in the present investigation. When low sulphur contents were maintained the presence of 0.10 per cent residual zirconium was sufficient to produce the refined graphite structure and the corresponding improvement in machinability.

It is pointed out that the experimental work described was done on a laboratory scale. Cost estimates based on this work indicate that the proposed desulphurization-zirconium alloy addition treatment may be carried out at a cost of less than \$2.50 per ton of treated iron. The investigation is currently being expanded to include full scale foundry tests, and accepted tool life machinability evaluations.

### Conclusions

The following specific conclusions may be drawn from the results of this investigation.



1. The abrasive nature of steadite in high phosphorus gray cast irons has been verified through microhardness tests of this constituent in comparison with the other structural components of gray iron. This study revealed that massive steadite approaches the hardness of free cementite.

2. In an attempt to alter the mode of occurrence of steadite in gray iron it was found that additions of up to 2.0 per cent of aluminum, calcium silicide, ferrotitanium, or zirconium silicon, produced little significant change in the phosphide constituent.

3. The control of microstructure in gray iron through annealing was studied. A sub-critical anneal at 1200 to 1400 F was shown to be satisfactory for the production of the desired ferritic matrix structure in pearlitic irons containing no free cementite. An improvement of 85 per cent in the drill test machinability rating resulted from the application of this treatment to the base iron selected for investigation.

4. An investigation was made of the possibility of controlling the microstructure of this iron through adjustment of its carbon and silicon contents. It was found that although additions of silicon carbide and a proprietary mixture of graphite and ferrosilicon produced appreciably softer structures, the amounts of these agents required to produce a ferritic structure were in excess of commercially practicable additions.

5. A study was made of the effects of a variety of alloying agents upon the properties and structure of the base iron. These experiments indicated that an addition of zirconium sufficient to yield approximately 0.05 per cent residual zirconium would exert a beneficial influence upon machinability of the base iron provided the silicon content was maintained in the neighborhood of 2.75-3.00 per cent. Further studies revealed that the addition of 0.10 per cent zirconium (residual) in the form of the 60 per cent silicon, 6 per cent manganese, 5 per cent zirconium alloy produced a still greater improvement in machinability. In the latter case additions of 2.0 per cent of this alloy consistently produced increases of from 40-45 per cent in the drill test machinability rating of the base iron.

#### Desulphurization of Base Iron

6. Desulphurization of the base iron, with and without alloy additions, formed the basis of the final phases of the investigation. It was found that desulphurization alone would produce appreciable improvement in the machinability rating of the base iron (i.e., an increase of 24 per cent over that of the untreated iron).

7. In a series of "synthetic" irons having extremely low sulphur and manganese contents, as-cast machinability ratings were essentially the same as those of the untreated base iron. However, the addition of 0.10 per cent zirconium (residual) to this iron resulted in an outstanding improvement (86.5 per cent) in the drill test machinability rating.

8. Sodium carbonate desulphurization of the base iron to give a residual sulphur content of 0.010 to 0.015 per cent followed by the addition of zirconium as 13 per cent zirconium ferrosilicon resulted in decided improvement in machinability. A 1.0 per cent

addition of the 13 per cent zirconium alloy was shown to be sufficient to yield 0.10 per cent zirconium (residual) in the treated iron. Specific conditions arising from low sulphur content and the zirconium addition were found to give rise to an improvement of 40-45 per cent in the machinability of the base iron with but little alteration of its composition other than sulphur.

9. This treatment is believed to be commercially practicable and may be carried out at an estimated cost of less than \$2.50 per ton of treated iron.

#### Acknowledgment

This paper is a condensation of a thesis submitted to the faculty of the Graduate School of Vanderbilt University in partial fulfillment of the requirements for the degree of Doctor of Philosophy. The author wishes to acknowledge the invaluable assistance of the following men whose cooperation and support made this work possible: Mr. H. A. Berg, President, Woodward Iron Company, Dr. W. P. Fishel, Professor of Metallurgy, Vanderbilt University, and Dr. James T. MacKenzie, Technical Director, American Cast Iron Pipe Company.

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## DISCUSSION

Chairman: J. T. MacKENZIE, American Cast Iron Pipe Co., Birmingham, Ala.

Co-Chairman: W. W. LEVI, Lynchburg Foundry Co., Radford, Va.

O. W. BOSTON (Written Discussion):<sup>1</sup> The writer finds this paper to be a valuable contribution to the machinability study of cast iron. The author lists three general methods of evaluating the machinability of the several heats and treatments of irons and summarizes them by listing (a) the drill penetration tests, (b) the life of a carbide-tipped face milling cutter, and (c) the tool life of a single point tool in turning. The writer questions the value of the first test but in study of this same project in which the writer participated it was found that ratings based on tool life in turning with high speed steel tools did correspond favorably with the ratings obtained by the drill penetration test. This will presumably be the subject of a later paper by the author. The conclusions arrived at by the author are believed to be of definite commercial value and are also of extreme interest.

HANS ERNST (Written Discussion):<sup>2</sup> Dr. Austin is to be congratulated on the presentation of a significant contribution to the literature of cast iron machining. His paper gives ample evidence of a painstaking research and a careful review of previous work by other investigators.

It is of particular interest to the writer to note that the improvements in machinability reported in the paper—as by desulphurization and zirconium alloy additions—are associated with increases in the percentage of ferrite in the matrix structure, and that it therefore appears that such treatments may in part be the equivalent of annealing, at a considerably lower cost. This is also significant in that it substantiates the findings reported by Feild and Stansbury<sup>8</sup> that the relative machining properties of various cast irons may be inferred from their microstructure.

A question evoked by the paper is whether the proposed treatment can be applied equally well to base irons of lower

phosphorus content similar to SAE No. 110 and thus further improve their machinability. As indicated by Feild and Stansbury,<sup>8</sup> a very large improvement in tool life is obtained by transformation of the normal pearlite matrix to ferrite, by a full anneal. If a similar effect can be obtained by the desulphurization-zirconium alloy addition treatment, this should find wide commercial use. Many applications of cast iron exist in machines and structures where high strength and hardness are not essential, and where appreciable cost reduction may therefore be obtained by any treatment which will improve machinability.

Another question concerning the proposed treatment is its effect on the physical properties of the base iron. Values for these properties are not given in the paper.

In criticism, we would question the value of the drill penetration test as a good measure of machinability. In measuring penetration time with a standard drill shape and speed, under a constant axial force, we are actually determining the feed per revolution (or undeformed chip thickness) which will produce that force with a given material. The chip thickness will, therefore, be different for materials with different machining characteristics. As the axial force (or thrust) is not a constant function of the chip thickness, the penetration time will not indicate relative values of ease of machining.

Other objections to the drill penetration test are that the peripheral speed and chip thickness at the cutting edge are not constant but vary over a wide range; the rate of penetration is also affected somewhat by friction and interference of chip and drill with the periphery of the hole.

In place of the drilling test we would suggest a turning test on the end of a tubular sample approximately 1 1/4-in. OD x 3/4-in. ID x 3-in. long. This sample could be machined from the cylindrical portion of the keyhole test specimen used by the author. If the turning test is made with some form of tool dynamometer, using a procedure which has recently been developed for this work,<sup>31</sup> it is possible to obtain relative "machinability" values based on the fundamental physical quantities which govern the machining operation. In this way the tests will not only indicate the relative benefit obtained by a particular treatment but also the basic reason for the improvement by its effect on either the coefficient of friction  $\mu$ , the initial shear strength of the material  $S_s$ , or the machining constant  $C$  (which indicates the rate of change of shear strength with respect to compressive stress). Thus the tests will tell not only what happens but why.

Furthermore the forces (or power) involved in metal removal represent only one aspect of the complex quantity which we call machinability. In addition, and often of greater importance, are the aspects of finish and tool life. While both of these are related to coefficient of friction, shear strength, and the machining constant, they are also influenced by other quantities such as work hardenability (which is usually measured by the Meyer exponent  $n$ ) and, in the case of tool life, by what may be called the inherent abrasiveness of the material.<sup>32</sup> The latter is actually determined by the inherent hardness of the individual micro-constituents of the material at their operating temperature. Obviously, tool life determinations cannot be made during the laboratory scale stage of an investigation, such as reported by the author in this paper, but will be essential in evaluating the results of future full scale foundry tests. It is conceivable that a particular treatment might appreciably decrease the forces involved in machining by decreasing the shear strength of the metal, yet, if it introduced a highly abrasive micro-constituent, it might actually decrease the tool life.

A word of caution seems in order with respect to the author's first conclusion, namely that "The abrasive nature of steadite in high-phosphorus gray cast irons has been verified through micro-hardness tests of this constituent in comparison with the other structural components of gray iron." It must be realized that these micro-hardness tests were necessarily made at room temperature, whereas the particles of steadite which actually contact the cutting tool are at a very high temperature, possibly on the order of 900 F. As Feild and Stansbury<sup>8</sup> have pointed out: "Steadite is the iron-iron phosphide eutectic and hence is the lowest melting constituent present in the iron. The high temperature at the tool point during the milling operation should be sufficient to reduce considerably the hardness of the steadite while cutting. This reduction in hardness may even be

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so great as to allow the steadite to smear over the tool surfaces to some extent, thus in some cases protecting these surfaces instead of abrading them as the hard carbides do." This seemed to account for the comparatively small differences in tool life which they found when machining two cast irons of similar matrix structures but containing, in one case, approximately 5 per cent steadite.

While Feild and Stansbury's studies were confined to cast irons having phosphorous contents under 0.4 per cent, it seems likely that this would also apply in some degree to the higher phosphorous iron used by the author.

It is hoped that the author will find ample support for continuing this promising investigation.

F. W. BOULGER (*Written Discussion*):<sup>2</sup> Mr. Austin's useful contribution is interesting from several viewpoints. It presents additional evidence confirming the principle that microstructure, rather than composition, controls the mechanical properties of metals. Proceeding on this premise, he found several methods of improving the properties of an iron generally considered unsuitable for production machining operations because of its high-phosphorus content. It is worth repeating that the author found it unnecessary to change the phosphorus content to improve the machinability. The addition of zirconium to, or removal of sulphur from, the base iron changed the structure enough to cause a decided improvement. The treatments suggested should be worth while in many applications because they are inexpensive.

It is of interest to students of that complex of properties called "machinability" that such useful information was obtained by a simple drilling test. Testing methods do not have to be complicated to be useful and reliable. The differences in cutting rates produced by different treatments were so marked that they are quite convincing. Furthermore, the correlation between structures and machinability agrees in general with the conclusions of Feild and Stansbury based on other testing methods.

In closing, I would like to ask whether the drills dulled

appreciably faster during tests on poor irons than they did during tests on good irons. If no appreciable wear occurred during the approximately 37-sec period required for drilling the base iron, then wear did not influence the drill test ratings. Therefore, they would not provide information on the "abrasiveness" of steadite and iron carbide. Tests made under conditions where tool wear is not a factor, often agree with machinability ratings based on tool life. It is reasonable to believe, therefore, that abrasive particles do not have as much influence on tool life as is sometimes conjectured.

DR. AUSTIN: I should like to thank Mr. Ernst for this valuable discussion. With regard to the mechanical properties of the iron, unfortunately, there was insufficient material available in the small heats that we were making to produce test samples. A complete evaluation of mechanical properties of the treated iron is to be made at a later date.

The limitations of the drill test are definitely realized. In this connection we have had tool life machinability tests conducted at the University of Michigan by Prof. O. W. Boston. It was found that results of these turning tests compared favorably with the more qualitative evaluations by the drill test.

With regard to the abrasiveness of phosphide, I should like to point out that our phosphorus content (0.60 to 0.70 per cent) was much higher than that studied by Feild and Stansbury (0.30 to 0.40 per cent). Professor Boston's tests showed that tool life is decreased by some 35 per cent in going from a 0.2 per cent phosphorus iron to a 0.6 per cent phosphorus iron. In spite of the fact that the phosphide is the lowest melting constituent, I am satisfied that it does retain its hardness at normal machining speeds (up to 200 fpm) as evidenced by these tool life test results.

In reply to Mr. Boulger's question regarding the comparative rates of dulling of test drills, I can say that no significant differences were observed in the rate of dulling between the low phosphorus and the high phosphorus irons. This is indicative of the fact that the drill test simply evaluates the relative softness of the matrix structure and not the abrasive effect of the steadite. The abrasiveness of steadite verified by the tool life determinations previously referred to.

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# TENSILE PROPERTIES VS. COMPOSITION OF DOUBLE NORMALIZED CAST STEEL

H. A. Schwartz and W. K. Bock\*

## ABSTRACT

The authors derived an empirical equation for calculating tensile strength of double normalized cast steel from its chemical composition. A critical survey of pertinent literature and an evaluation of data was made by the authors. The paper deals only with double normalized acid electric and basic open hearth steels.

## Introduction

IT IS OF OBVIOUS PRACTICAL IMPORTANCE to be able, if possible, to calculate the tensile strength of normalized steel from its chemical composition. Walters<sup>1</sup> and Carapella<sup>2</sup> have recently published the results of their efforts in this field. We shall presently have occasion to discuss their results in greater detail.

Walters, by analogy with Grossman<sup>3</sup>, proceeded on the assumption that tensile strength is the product of a base figure and certain functions of the concentrations of the several alloying elements. These functions are somewhat complex and were given in the form of graphs only, which were extended to represent the basic observations.

Carapella, on the other hand, proceeded from certain basic assumptions as to the relation of Brinell hardness in the normalized state, to tensile strength, of the Brinell number normalized to that of martensite of the same composition in terms of the natural logarithm of Grossman's ideal critical diameter<sup>3</sup> and the Brinell hardness of martensite as a function of the natural logarithm of carbon. He derived certain constants necessary for his equation from a study of 25 steels varying greatly in hardenability.

Although Walters and Carapella's expressions seem to have little in common, it is true that on expanding the latter's equation, a product involving carbon content and hardenability appears, and since ideal critical diameter is the product of a series of factors (Grossman *loc.cit*) some degree of similarity is implicit.

It is not indispensable for practical use to have a well supported basis for an empirical equation if the

latter fits the facts but it is certainly desirable to have some fundamental knowledge of the form of such relation if one wishes to interpret the effect of the several alloys.

A satisfactory equation should:—

- (a) agree as accurately as possible with the observations,
- (b) permit extrapolation to the tensile strength of pure iron with not too great an error,
- (c) have a form having physical meaning,
- (d) be, for practical reasons, reasonably simple and accurate to evaluate.

The items are arranged in the order of importance at least for research use. For purely practical application (c) and (d) might be reversed. It is possible as will presently appear, that these requirements are not compatible with one another.

## Data

The steels underlying this investigation have been previously described by Schwartz and Bock<sup>4</sup>. The tensile tests here used constitute a very small fraction of the numerical data used in that paper. The effect of McQuaid-Ehn grain size in this connection is highly questionable in the light of more recent knowledge<sup>5</sup>. We can no longer re-study the steels in the light of other means of measurement.

For ease of reference the composition, grain size, calculated hardenability and tensile strengths are combined in Table 1. Table 2 gives data as to the tensile strength elongation relation. The tensile properties are the average of two specimens normalized from 900° C (1650° F) and renormalized from 820° C (1510° F). Both acid electric and basic open hearth steels were used. Specimens were about 1 in. in diameter when heat treated.

The statistical treatment of the data for both tensile strength and elongation and for both single and double normalized steel indicated that there is no significant difference between single and double normalizing for tensile strength, but there is a difference for elongation. The present paper deals only with double normalized material, partly because the results are more predictable. The elongation of the single

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TABLE 1—CHEMICAL ANALYSIS AND PROPERTIES OF STEELS STUDIED

No.	C	Si	Mn	Ni	Cr	Mo	V	Cu	Grain Size*	Tensile Strength, D <sub>1</sub>	Psi
1	.31	.44	.71	3.35	.02	.03	..	.07	8	2.75	10,695
2	.33	.37	1.20	.65	.02	.02	..	.10	9	1.60	10,120
3	.38	.43	1.43	.01	.03	.02	..	.10	8	1.88	10,700
4	.24	.30	.74	.03	.02	.39	..	.05	10	1.38	8,530
5	.36	.32	.73	1.67	.54	.26	..	.04	9	4.23	14,590
6	.25	.28	.81	1.37	.02	.10	.09	.07	8	1.95	9,170
7	.32	.41	.81	..	.08	.04	..	.10	11	0.98	8,955
8	.23	.38	.74	.81	.27	.22	..	.03	10	2.12	9,955
9	.35	.34	1.18	.02	.03	.02	..	.10	9	1.35	9,855
10	.40	.39	.67	.22	1.44	.47	.15	.12	8	6.50	17,300
11	.395	.89	.46	1.00	.04	.27	..	.02	7	2.29	11,255
12	.365	.42	1.20	.68	.05	.28	..	.06	8	3.57	12,780
13	.385	.48	.74	.16	1.12	.33	..	.09	6	5.77	15,300
14	.40	.51	1.58	.12	.12	.08	.13	.11	8	3.50	15,155
15	.395	.38	1.50	.09	.06	.41	..	.12	8	4.46	14,750
16	.355	.42	1.34	1.33	.33	.10	..	.12	10	4.24	15,800
17	.395	.52	1.51	.08	.16	.21	.09	.25	8	5.67	13,265
18	.38	.45	1.34	.04	.07	.03	..	.03	9	1.69	10,595
19	.355	.44	1.26	.07	.11	.11	..	1.64	9	3.45	14,375
20	.32	.29	1.30	..	.03	.43	.09	.05	9	3.65	11,980
21	.325	.47	1.41	.06	.38	.11	..	.02	8	3.46	12,635
22	.39	.39	1.60	.12	.05	.34	..	.81	9	4.88	15,890
23	.52	.58	.83	.01	.02	.06	..	.10	8	1.59	11,140
24	.44	.65	.88	.02	.02	.49	..	.35	10	3.20	12,570
25	.42	.40	.75	.03	.01	.16	..	.11	10	1.35	9,655
S	.255	.528	.551	..	..	..	..	..	8†	0.72	7,860
M	.238	.362	.604	..	..	..	..	..	8†	0.69	7,440

\*McQuaid-Ehn at 1700° F. †Assumed.

normalized steel was, in general, decidedly lower than that of the double normalized, and more erratic.

Since some difficulty was encountered in locating the curve at its lower end, the above data were supplemented with the average data for two groups of grade "B" steels, each containing some 50 heats.

#### Evaluation of Data

The significance of this paper lies entirely in the evaluation of these data. The problem is merely one of curve fitting, selecting a form of relationship and determining (except for transcendental functions) by least squares, the appropriate constants. The details are enormously voluminous and cannot be reproduced here. For many of the calculations we are indebted to James V. Anthony of this laboratory. It is assumed that the foundry reader will not be interested in the purely mathematical manipulations involved and these are here almost completely omitted for simplicity's sake.

The first and simplest assumption was that equal increments of concentrations of a given element added equally to tensile strength regardless of what was already present. It is already well known that tests in which all elements are held constant and carbon (for example) alone varied, do not indicate that this form of equation has merit. Reasons for its inclusion will appear later. This yields the following equation:—

$$T_1 = -1.03379 + 0.14496 C + 0.00738 Si + 0.04175 Mn + 0.01287 Ni + 0.05609 Cr + 0.06206 Mo + 0.08938 V + 0.01700 Cu \quad (1)$$

Here, and throughout,  $T$  is in 10,000 psi units and the symbols of the elements indicate their concentrations in 0.01 per cent. The standard error of fit is 2.31,

which is too high for utility, and the extrapolated tensile strength for iron (constant term of the equation) is negative. The linear relation thus has no merit except as a starting point.

A word of explanation as to the term "standard error of fit." This is a term used in statistics as the measure of the degree of scatter of observations about a trend line. Its mathematical significance need not be here discussed but the reader can realize its significance by remembering that about 68 per cent of all observations will fall less than the standard error away from the trend line (or from the results calculated by the equation in question).

Successive attempts to improve the fit will now be outlined.

A plot of the deviation of Eq. (1) against hardenability indicated the need of a term for hardenability and a new equation was derived thus:—

$$T_2 = -1.45479 + 0.1517 D + 0.14496 C + 0.00738 Si + 0.04175 Mn + 0.01287 Ni + 0.05609 Cr + 0.06206 Mo + 0.08938 V + 0.01700 Cu \quad (2)$$

Here  $D$  is the Grossman ideal critical diameter in inches.

The extrapolation to pure iron is still bad but the standard error of fit has been reduced to 0.839. The introduction of  $D$ , which must be calculated, has added to the labor of calculation very greatly.

Except insofar as  $D$  contains a factor  $C^{1/2}$ , Eq. (2) is a linear function of carbon, whereas the data for the tensile strength of normalized steels<sup>4</sup> over a long carbon range, indicates that carbon should enter about as the  $2/5$  power.

Graphically it appeared that we were not so much concerned with  $C$  as with  $\frac{C}{D}$  and the equation:—

TABLE 2—MECHANICAL PROPERTIES OF STEELS STUDIED

No.	Tensile Strength	Elongation	Tensile Calculated from Composition $T_5$
1	10.6	21.0	11.4
2	10.2	25.0	10.1
3	10.7	20.5	11.1
4	8.6	25.8	8.8
5	15.0	12.8	14.4
6	8.8	30.0	9.6
7	9.0	26.3	8.6
8	10.1	23.3	9.7
9	9.9	23.3	9.6
10	18.9	4.8	16.8
11	11.2	13.9	10.3
12	12.5	19.5	12.6
13	13.8	15.3	16.0
14	14.4	12.5	14.3
15	14.6	15.3	14.5
16	16.3	12.8	14.4
17	12.4	18.5	15.0
18	10.6	22.8	10.9
19	13.4	15.0	13.9
20	11.8	21.8	13.0
21	12.4	16.3	12.9
22	15.7	8.5	15.3
23	10.8	18.8	10.8
24	13.1	11.5	13.3
25	9.6	19.5	9.7

$$T_3 = 0.69077 + 0.1577 D + 0.17502 \left(\frac{C}{D}\right)^{0.8} \\ + 0.04615 \text{ Si} + 0.04467 \text{ Mn} + 0.01155 \text{ Ni} \\ + 0.04616 \text{ Cr} + 0.05320 \text{ Mo} + 0.03163 \text{ V} \\ + 0.01225 \text{ Cu} \quad (3)$$

results.

The standard error of fit has increased to 0.952 but the extrapolated iron strength has become positive, though still far too low. The deviations of the observations from Eq. (3) were found not to be correlated with grain size. The term containing  $\left(\frac{C}{D}\right)^{0.8}$  greatly adds to the labor of computation. Except for the insignificant improvement for pure iron Eq. (3) is a step backward from Eq. (2).

A plot of  $T_3$  against the observed values of tensile strength ( $T_0$ ) indicated that the curve was S-shaped and an attempt was made to find a "power function" of  $T_3$  which should better fit the results. The relation:—

$$T_4 = 34.339 - 8.404 T_3 + 0.808 T_3^2 - 0.022 T_3^3 \quad (4)$$

was developed.

The form of Eq. (4) is completely empirical and cannot have physical meaning for  $T_4$  has a minimum value (about 7.5) when  $T_3$  is about 7.5 and a maximum (about 16.3 when  $T_3$  is about 18.4). The tensile strength of pure iron (when  $T_3 = 0.69577$ ) comes out very much higher than the lower carbon and lower manganese steels of the table ( $T_3 = 7.4 \pm$  or  $7.9 \pm$ ). Except by graphic means it is not easy to use, and its only value here is in leading to the next step. The standard error of fit is, however, 0.730, the best so far encountered.

This brings one to the point of looking for a curve whose form will correct for the fact that Eq. (1) or (2) give low results at low alloy content and high results at high alloy content. It would be nice also if that curve did not necessitate using  $D$ , which would greatly simplify the calculation.

The statistical analysis of various forms of curves which suggested themselves showed that  $\left(\frac{C}{D}\right)^{0.8}$  could be dispensed with, but not  $D$  itself; hence one must go back and compute a relation between  $T_2$  and a new calculated value of  $T$ , which we will call  $T_5$ . The form of relation chosen is not related to any reasoned process but was selected as a curve which best fits the observations. The reason for insisting on a mathematical equation rather than on a curve put in by inspection is to secure a graph which interpolates (and extrapolates to some degree) according to a uniform system.

Leaving out all the intervening mathematical routine, the following equation was selected

$$T_5 = 12.28 + 5.17 \int_0^{0.20(T_2 - 12.24)} e^{-\lambda^2} d\lambda \quad (5)$$

The expression is of formidable appearance but tables exist for evaluating the integral<sup>6</sup>. Alternatively we can plot, as in Fig. 1, the line having the equation

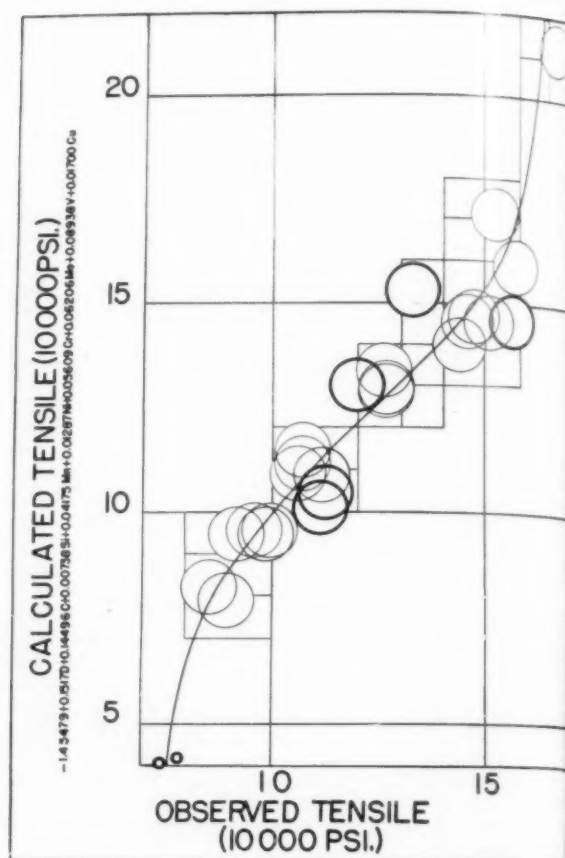


Fig. 1

and read the values of  $T_5$  corresponding to any value of  $T_2$ .

The standard error of fit of this expression is 0.63 but the strength of pure iron corresponding to  $T_2 = -14.5479$ , is far too high, 7.66.

In Fig. 1 the tensile strengths calculated from Eq. (5) are plotted against those which have been observed. The radii of the circles representing the several heats are twice the average standard deviation of the tensile strength observation, i.e., about 0.5, and indicate the range within which the true observed values almost certainly fall. The circles for the "Grade B" data are small since these observed values are the average of many observations. Circles whose horizontal diameters do not touch the trend line are made extra heavy. The line represents the relationship of equation Eq. (5).

The scatter of points about the line in Fig. 1 is nearly random that if some other function of  $T_2$  were substituted, no great improvement in fit could well be expected.

#### Comparison With Walters and Carapella

It now seems in order to compare the accuracy with which our equations predict results with the accuracy of prediction of the equations of the earlier workers. It must not be forgotten that our work pertains to cast metal and theirs to rolled products. Applying Walters' methods to our data, one is somewhat handicapped by difficulties in reading off values from the published



Walters did not choose to give equations for his curves because, as he said in discussion, he did not want to force his observations into the straight-jacket of algebra. He apparently did not consider that when data are represented by a line considered to fit them best, an equation is assumed whether its form is determined or not. An algebraic expression is less subject to errors of interpretation than a curve.

Be that as it may, working as carefully as we could, Walters' calculation has a standard error of fit of 1.46. It gives a mean strength for our 27 steels 0.451 lower than our observed mean. If 0.451 were added to each of his calculated values the standard error would be reduced to 1.39. Since his method begins with an assumed value for iron, it of course reproduces that plausible value. No further checks as to details are possible. The method requires considerable calculation, though less than ours.

Carapella's equation, as published, is the simplest of all to apply but has a standard error of fit of 0.990.

If, for Carapella's coefficients, new ones are calculated using our own data for the relation of tensile strength to hardness, which differ slightly from his, and calculating new values of his constants  $A$  and  $B$  to produce the best possible fit, we obtain an expression of his form in our units:—

$$T = \frac{10.901 \ln C + 15.8324}{-0.836 \ln D + 3.13980} + 1.797$$

which has a standard error of 0.972.

By "L'Hospital's rule" the value of this expression in the absence of any alloy is about  $-10.0$ , indicating the rapid failure of this form of equation to predict the strengths of the lower carbon and unalloyed steels.

Our Eq. (5) seems to reproduce our observations with distinctly greater accuracy than either of the published relationships, even when these are modified as best we may to fit our data. Our form extrapolates toward pure iron better than Carapella's but not so well as Walters'. It requires that the presence of about 0.035 per cent each of phosphorus and sulphur, and the existing average amounts of undetermined residual Sn, B, Ti, Al, O<sub>2</sub>, N<sub>2</sub>, H<sub>2</sub> and perhaps others raise the tensile strength of otherwise pure iron to something like 7.5 (75000 psi). We do not like this assumption but cannot avoid it.

The standard error of fit of 0.634 is made up of the standard deviation of our tensile results, about 0.325 and the actual standard error of fit of the equation. The standard error of fit of our equation compared with perfectly exact tensile observations, or the average of a very larger number of tests, is about 0.55.

This is not a study of the properties of forged steels but tests were made on the identical material used in this investigation after being forged (reduction about 60 per cent) and then double normalized. The strength of the numbered steels, except No. 20, averaged 10,330 psi; i.e., 1.033 in our notation, higher after forging. This will serve as some guide in correlating our data with data on rolled products, whose reduction is, of course, much greater.

The coefficients of Eq. (2) furthermore bear not even a qualitative relation to the separate effects of

some of the elements as shown graphically by French<sup>7</sup>. However, the relation which converts  $T_2$  to  $T_5$  is curved. The effect of this is to introduce into the final relation the concept that the effect of a given amount of a given element depends on what else is present.

#### Relation of Tensile Strength to Elongation

We have considerable indication of the relation of tensile strength and elongation of normalized steels which prompted us to believe that a logarithmic function was indicated.

Using actual observed tensile strength,  $T$  and elongation,  $E$  of double normalized specimens, we obtain

$$\log E = 1.965 - 0.06025 T \quad (6)$$

where  $E$  is in per cent and  $T$  as before, in 10,000 psi units. This equation has a standard error of about 2.6\*.

In practice one needs to know the elongation before any tests are made so a calculated tensile strength must be used. One then gets

$$\log E = 1.925 - 0.05648 T_5 \quad (7)$$

The standard error of fit has risen to 3.5\*. A correction which assumed that Eq. (7) should have a term correcting for hardenability did not materially improve matters.

Some improvement in accuracy of prediction would seem desirable. By graphic and statistical attack it was developed that the differences between values calculated from Eq. (6) or (7) were not definitely capable of correlation with any variable other than C, Si and Mn.

Successive corrections for these elements were calculated, which when applied to Eq. (6) and (7) yield

$$\begin{aligned} \log E = & 2.028 - 0.06025 T - 0.00156 \text{ Si} \\ & + 0.00091 \text{ Mn} - 0.00255 \text{ C} \end{aligned} \quad (8)$$

and

$$\begin{aligned} \log E = & 2.004 - 0.05648 T_5 - 0.00108 \text{ Si} \\ & + 0.00120 \text{ Mn} - 0.00445 \text{ C} \end{aligned} \quad (9)$$

The standard errors\* are now 1.8 and 2.76 respectively. It is unlikely that further improvement could be secured unless other variables not so far considered can be discovered.

#### Practical Application

The practical use for which this study was carried out has already been stated in the first sentence of the introduction. Obviously it is important to be able to predict the strength of a steel, or its elongation, from its composition.

The reverse problem is, of course, meaningless, for plainly there are innumerable combinations of concentrations which when set in Eq. (2) give the same value of  $T_2$  and hence the same value of  $T_5$  from Eq. (5).

Generally the problem is considerably simpler. One may wish to know what the effect of the addition of a given amount of some alloy to (say) a plain carbon steel will be.

It is easy to calculate  $T_2$  with and without the added alloys (residuals cannot be disregarded) and

\*The standard errors given for Eq. (6) to (9) inclusive are those calculated for  $E$ ; not those for  $\log E$  as our phraseology might readily be interpreted.

insert the two values in Eq. (5) and so get  $T_5$  with and without the alloy. The difference is the improvement.

Note further that if the tensile strength without the alloy is known from experience, the result obtained by adding the difference of the  $T_5$ s with and without the alloy, is more reliable than the abstract calculation.

Since the contribution made by the second term on the right of Eq. (2) depends on all the elements, moderate changes in a single element do not greatly affect  $T_2$  via the term in  $D$ .

It is, therefore, fairly easy to calculate how much a given element must be changed to produce a given change in strength. From the existing composition calculate  $T_2$  and by Fig. 1 find the corresponding  $T_5$ , change this by the amount desired and obtain, again from the figure, the needed value of  $T_2$ . The difference between the original and altered values of  $T_2$  divided by the appropriate coefficient of the element to be changed, as shown in Eq. (2), gives the required change of concentration.

If Eq. (6) and (7) were rigidly exact it would mean that regardless of composition a double normalized steel of given tensile strength could have but one elongation. The corrections made in obtaining Eq. (8) and (9) indicate the extent to which departures from this rigid proportionality are possible.

For a given value of tensile strength we have, even with the corrections of Eq. (5), a given value of  $T_2$ . Neglecting the effect on  $D$  the substitution of 4.175 parts of carbon for 14.50 of manganese, or of 4.175 parts of silicon for 0.738 of manganese, is without effect on  $T_2$  as indicated by Eq. (2). It is not, however, without effect on elongation even if the strength does not change.

From Eq. (9) the removal of 4.175 units of carbon increases  $\log E$  by about 0.019 and the addition of 14.5 parts of Mn further increases it by 0.017, making a total increase of 0.036, which represents multiplying the original elongation by about 1.08. That an increase in elongation would occur could have been quickly predicted since the algebraic sign of the terms in C and Mn in Eq. (9) are opposite so that compensating changes of the two, for strength, are cumulative for elongation.

The substitution of carbon for silicon is, however, to some extent compensating. From Eq. (2) 14.5 parts of silicon may be substituted for 0.738 parts of carbon without change of strength. The addition of 14.5 units of Si, from Eq. (9), decreases  $\log E$  by 0.0157 and the removal of 0.738 units of carbon increases  $\log E$  by 0.0033, a net decrease of 0.0124, and the substitution is undesirable from the viewpoint of elongation.

For precise work a calculation including the changes in  $D$  would be required.

We are somewhat inclined to apologize for inserting a section showing the reader the mere arithmetical consequences of our equations. He can make these and many others of the same type for himself.

One may add another practical word of caution. The relation exemplified in Eq. (5) and in Fig. 1 indicates that the question of how much the addition of a given amount of an alloy will alter the properties

of steel (in general), is unanswerable. It depends not only upon the individual concentration of each component separately but upon their joint effects.

That we have been able to produce a means of prediction with even considerable uncertainty, in so complicated a field, seems to us to be a useful forward step.

### Conclusions

We conclude that the mathematical relation of tensile strength to chemical composition of normalized and untempered cast steel specimens which best fits our observations is:—

$$T = 7.7 + 5.17 \int_{-\infty}^x e^{-\lambda^2} d\lambda$$

where  $x$  is  $-2.738 + 0.03034 D + 0.028992 C + 0.001476 Si + 0.008350 Mn + 0.002574 Ni + 0.0011218 Cr + 0.0012412 Mo + 0.0017876 V + 0.003400 Cu$ .

Here  $T$  is in 10,000 psi units,  $D$  in inches and the concentration of the elements is in 0.01 per cent. The equation is a simple transformation of Eq. (5).

One should not base too many theoretical conclusions upon the form of an empirically determined curve but if agreement with this form should be evidence that the form is correct, then:—

The tensile strength of normalized steels approaches maximum and minimum values asymptotically and the effect of a given increment of a given element cannot be related directly to the amount of that element already present but depends upon the cumulative effect of all the concentrations and of their effect (as factors) on hardenability.

Our equation, although apparently it has the lowest standard error of fit of all the suggested forms, still leaves much to be desired with regard to the precision of its predictions. This precision is not likely to be improved by further attempts to find a better form. An investigation of the potency of variables not so far included is a more likely approach to improvement.

We offer no explanation as to why the tensile strength should logically involve the probability integral function, nor can we account quantitatively for the relative values of the several coefficients by any simple atomic property.

It may be merely coincidence that the elements forming metallic carbides have relatively high coefficients.

Elongations can be calculated most accurately from known tensile strengths, especially if recognition is taken of C, Si and Mn concentrations. Calculations from composition alone using the equation

$$\log E = 2.004 - 0.05648 T_5 - 0.00108 Si + 0.00120 Mn - 0.00445 C$$

possess considerable utility.

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## DISCUSSION

Chairman: FRANK KIPER, Ohio Steel Foundry Co., Springfield, Ohio

Co-Chairman: V. E. ZANG, Unitcast Corp., Toledo, Ohio

R. A. WILLEY:<sup>1</sup> Are the calculations referred to by the authors based on normalized properties? What is taken into consideration for tempering or drawing back?

MR. BOCK: These data are entirely on 'double normalized steel. As stated in the paper, if we had used single normalizing, the tensile strength would not have changed much. We would have had erratic behavior on the elongation compared to what we have here.

In answer to your question on calculations based on the tempering back of the steel, you would have to go through the same procedure to establish the relationship between tempering temperature and the properties mentioned in the paper. It is beyond the scope of this paper, but the same method that we used here could be used to get the effect of tempering temperature. In fact, quenched steels and tempered steels are handled in the same way.

J. A. WETTERGREEN:<sup>2</sup> Did the authors' calculations take into account the presence of residuals?

MR. BOCK: Yes. The residuals were all determined and everything we found in our test samples was put into our equations.

C. W. BRIGGS:<sup>3</sup> I think the paper would have been of greater value to everyone if the authors had given an example and shown how the values are calculated. The point has been raised by several people who have reviewed the paper that they would have found it much more understandable and readable if the authors had done so even though they do say it is extremely simple.

MR. BOCK: We have lived with these calculations so long that they appear simple to us. The simple part of the calculations stems from the fact that some of the mathematics fitting those S-shaped curves have been dropped out of the paper.

As an example of the operation of these equations, assume an analysis of

C	—0.30
Si	—0.30
Mn	—0.75
Ni	—0.50
Cr	—0.02
Mo	—0.02
V	—0.10
Cu	—0.05

By use of Grossman's data we find  $D_1=1.92$ . Applying Eq. 2 we obtain

$$T_1 = -1.45479 + 0.1517 \times 1.92 + 0.14496 \times 30 + 0.00738 \times 30 + 0.04175 \times 75 + 0.01287 \times 50 + 0.05609 \times 2 + 0.06206 \times 2 + 0.08938 \times 10 + 0.01700 \times 5$$

Thus  $T_1 = 8.397$  and using Fig. 1 we find for a calculated tensile of 8.40 an observed tensile of 8.98 and so find  $T_2 = 0.98$ . We should expect the above analysis to yield a tensile

strength of 89,800 psi in the normalized condition. This value is subject to a standard error of 6340 psi.

Next, using Eq. 9 we have

$$\log E = 2.004 - 0.05648 \times 8.98 - 0.00108 \times 30 + 0.00120 \times 0.75 - 0.00445 \times 30$$

Then  $\log E = 1.421$  and therefore  $E = 26.4$

This means that our alloy double normalized will have an elongation of 26.4 percent with a standard error of 2.76 percent.

MR. WILLEY: Does the equation indicate that you would expect the same number of points of chromium to be as effective in increasing the tensile strength of a steel in the range of 0.60—0.90 percent as in a 1.35—1.65 percent? In other words, if an increase of 20 lb of chrome will give you a definite increase of tensile strength of say an SAE 4335 steel, will 20 lb of chrome give you the same increase in tensile strength in an SAE 3335 steel?

MR. BOCK: Probably not. In the first equation the relation was written with the idea that the addition of say 20 points would increase the strength a given amount. We did not believe that to start with. We threw it out. It was an absurd equation, except for a starting point. In the final equation chrome enters in two places. It occurs in the composition part of that equation and also in the hardenability. That is where the effect of the amount of chrome and of the amount of other elements already present enters, so that the effectiveness at any given level of adding to that level will change mostly due to that term involving the hardenability. That is why we had to leave it in, even though we did not like to do so and complicate the calculations.

The essence of this paper is to show that a given increment of any alloy does not produce the same increment of tensile strength regardless of what else is present, but that the effect for any and all alloys depends on the tensile strength existing without that increment.

R. E. KERR:<sup>4</sup> How do you account for the fact that the upper part of your S-curve approaches vertical and therefore lacks sensitivity?

MR. BOCK: That S-curve is strictly empirical. We observe from plotting our points that we have to have some kind of correction which will move the high values to the axis of the chart and move the low values in the other direction. We could have fitted any other curve which would have given that general shape and probably have gotten about the same results. However, the curve we used we thought was the easiest to handle because of the tables.

As to losing sensitivity at the high end of the curve, we are dealing with low-alloy steels as specified in the paper. We do not expect to extrapolate much beyond that top point which was about 180,000 psi, not for normalized steels. Although we lose sensitivity in there, the physical limitations of normalized low-alloy steels means that we would have few or no heats with strengths in the range of low sensitivity.

Any form of curve we selected would have to meet the observed condition that equivalent increments of alloy content produce in weak steels a slight increase in strength. The effect of the increment first increases as the strength decreases and then decreases. As the strength gets to a certain point alloying adds little or nothing.

CHAIRMAN KIPER: Has any work been done to indicate what another type of heat treatment such as a quench and draw to a light hardness would do to the equation?

MR. BOCK: Yes. This same set of steels was examined for the relationship between tensile and elongation. In that work we used not only the normalized but the normalized and tempered and the quenched and tempered steels. These results appear in the paper entitled "Tensile Properties of Medium-carbon Low-alloy Cast Steels" published in the *Trans. AIME*, vol. 158 (1944). I do not believe that there were any of these corrections necessary on that work.

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# CAN CASTINGS BE ENGINEERED?

By

F. G. Tatnall\*

## ABSTRACT

*The author describes thin wire strain gages and strain gage technique for analyzing strain in metal components. He also discusses brittle lacquer technique for stress analysis in components to show where design improvements are necessary. The author outlines in the paper a procedure for design improvement.*

THE APPLICATION to casting design of the results of stress measurements is a comparatively recent development in the "engineering of castings." You will find in the instrumentation described a new tool for the foundry which might become as useful as well as revolutionary approach to design improvement in castings.

So as to let invention have full play, it seems logical to first describe the instrument in basic fundamentals and simple language, and then to suggest possible usage.

A wire  $\frac{1}{1000}$ th inch in diameter is one-third the thickness of the human hair, so fine that you can hardly see it. When working with it, if you lose the end of the coil, you might as well throw the whole spool away.

## Versatile Measuring Means

If 5 in. of such wire of copper-nickel composition be fixed to a piece of thin paper like cigarette paper, with a plastic cement like Duco, in the form of a grid, like Fig. 1, and if its ends be attached with a drop of solder to heavier wires leading to an external circuit, you will find that you have the most surprisingly versatile measuring means of all time.

The elongated grid cancels all components of strain other than that along its axis, as a good component divider should do.

The cement holds the wire on its carrier, the paper strip, like concrete holds a reinforcing rod. Since the tiny wire has a surface area in an inch of length equal to 2000 times its cross-section, the cement finds its grip so much stronger than the wire that, when bonded to any kind of a surface, it can stretch the wire as the surface stretched or compress the wire and

prevent its buckling, and at the same time electrically insulate the wire from the surface to which it is bonded. When the bonding agent is thus stronger than the wire, the wire is accurately deformed as the cement deforms. This is the key to this whole system of measurement.

When the wire is stretched or compressed it naturally changes in both length and cross-section, increasing resistance as it stretches or decreasing resistance as it compresses. The paper, which is acting only as a convenient means of carrying the wire, does not enter into the operation.

This wire gage is so small, so light, so flat that it does not change the weight, balance or natural frequency of the fabricated or cast part to which it is attached. It is sensitive to changes in length of one-millionth of an inch per inch, and can faithfully follow vibrations exceeding 50,000 cycles per second. When applied with the proper cementing technique its accuracy is right in line with the best of measuring instruments. Commercial applications with 1 per cent accuracy are common.

By "proper technique" is meant a cementing process which requires a little more care, but not too much more, than that of fixing a postage stamp to a letter. Technique is a short way of saying "learn it by doing it."

## Strain Gage Factor

Such a "postage stamp" gage or patch becomes so cheap and easily applied that numbers of them can be plastered all over a casting, assembly or structure to show up regions of maximum strain, maximum shear, and direction of principal strains either in static test or in service tests while measuring acceleration, repeated loading, impact, water hammer, explosions, or centrifugal forces, and give such thorough coverage that peak values do not escape.

Strains appear in terms of electrical resistance changes, which are readily measured with commercially available apparatus. To get resistance values back to strain again, a gage factor is used which is simply the ratio of "change in resistance to change in strain." For copper-nickel wire this is a convenient

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factor of "2". This factor "2" is a good figure. It means that resistance changes just twice as fast as strain, a 2 to 1 lever. Also it represents a condition of constant volume in the strained wire in the plastic range so that the gage can be stretched or compressed the unusual amount of 3 or 4 per cent and be restored unharmed and ready to go again, and at the same time the gage factor remains unchanged at "2" throughout this plastic deformation range.

No one knows why the gage factor should be 2 or why it should differ with different metallic compositions (it is 3.5 for chrome-nickel-moly wire). With Poisson's ratio at 0.3, geometry would indicate that for a stretch of 1 unit, the gage factor should be 1.7. There must be something beside geometry involved, and whatever it is, it is amazingly precise and unchanging.

We would like to use the largest gage factor possible so as to reduce the amount of electrical amplification required to operate commercial electrical instruments

from the gage output. Because of the negligible temperature coefficient of resistance of copper-nickel wire, we must be content with its small gage factor to gain the advantage of temperature stability.

This leads to the matter of temperature compensation for gages, a matter of first importance, but fortunately quite easy to achieve. Strain in a part under test is detected by the gage not only as the strain due to stress, but also as the strain due to temperature change which latter causes the part to expand or contract. A so-called "dummy" gage unscrambles the two when measuring in tension-compression. Gages back-to-back in bending will do the same thing. To explain this, the fundamental gage circuit must be described and can be best shown as an hydraulic analogy (Fig. 2).

When a pump discharges water into a line which branches into two identical smaller lines, the fluid flow in each branch is equal and there is no differential pressure indicated by the gage G. By opening or closing valves A, B, C and D some interesting flow



FIG. 1

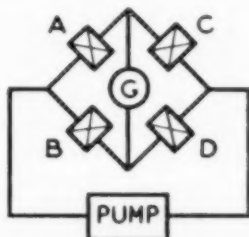


FIG. 2

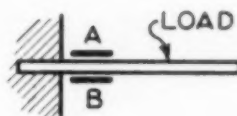


FIG. 3

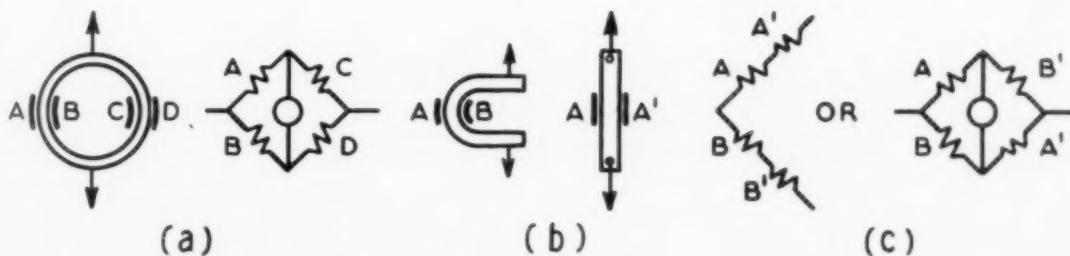


FIG. 4

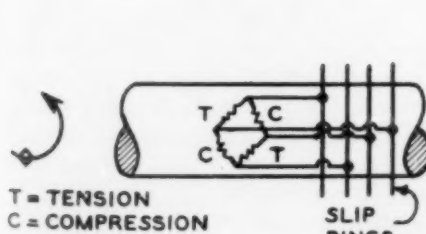


FIG. 5

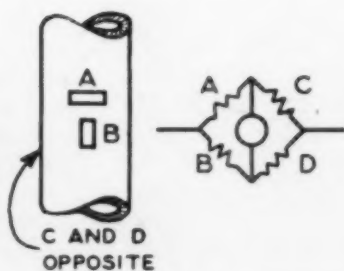


FIG. 6



FIG. 7

conditions are observed. For instance, to force more flow through B, partly close valve A. If, on the other hand, A and B are both either closed or opened an equal amount, the flow is again equally divided between the branches and the gage reads nothing. Try opening A full and closing on B, both actions add up in increasing the flow through A, doubling it, in fact. The effect on the differential pressure of opening or closing valve D is the same as for valve A, as can be seen by simply tracing the flow lines. By the same token, valve C acts the same as valve B. Hence, opening A and D and closing B and C quadruples the pressure differential.

In Fig. 2 think of the pump as a battery, the pipes as wires and the pressure gage as a millivoltmeter. The wire gages in the place of the valves completes a conventional wheatstone bridge circuit, in which the gages actually become valves, tension in the gage increasing its resistance which is the equivalent of closing the valve, compression opening it. The gages are far better than valves, they are inertialess and can be opened and closed with almost infinite speed, and precision.

This circuit can be temperature compensated easily. If A is the active gage, use B as a dummy by cementing it to unstressed material of the same kind and at the same temperature as that which carried gage A. A measures the strain due to stress plus the strain due to temperature change, while B (unstressed) measures strain due to temperature change only. So both valves are opened and closed together by temperature change thus cancelling its effect and leaving the stress response of A alone to be indicated on the millivoltmeter as unbalance due to external load.

When gages can be placed back-to-back in bending, automatic temperature compensation is obtained together with double output of the bridge. Take the cantilever beam of Fig. 3. Bending the beam downward puts the top gage A in tension and the bottom gage B in compression, the equivalent of opening valve B, closing valve A. But change of length of the beam due to temperature change will affect both gages alike, automatically cancelling out temperature, since gages are connected as in Fig. 2.

A cantilever beam like this when loaded at its free end serves nicely as a load weighing unit such as a dynamometer or scale. Four gages instead of two will again double the sensitivity of the beam (or four times the effect of a single gage). If the beam can be made quite flexible capable of tip travel of an inch or two, bridge output can be calibrated in terms of deflection, so the simple device becomes a deflectometer. If a weight be fixed on a beam end and the device calibrated in terms of acceleration we have an accelerometer.

#### Load Sensitive Devices

Load sensitive devices can take many convenient forms. Take a loop, for instance, such as a ball bearing race with gages back-to-back at points of maximum bending moment, Fig. 4 (a), (a bridge with four arms active), or if this is too stiff, cut half the loop

away, leaving a C-loop with two gages active, Fig. 4 (b), or if a loop be found too limber, use a tension bar with gages on two sides to average strain, Fig. 4 (c). Such a connection, A and A' in series, B and B' dummies, measured axial load, eliminates bending, as opposed to the cantilever beam circuit which measures bending, eliminates axial components.

To handle shear components or torque, bearing in mind that gages only measure along their axes, it becomes necessary to use a slightly different arrangement. Torsion in a shaft, for instance, can readily be measured if gages are placed at 45° to the shear planes since this is the direction of tension and compression components in a twisted shaft. This symmetrical arrangement of four gages provides a complete active bridge, Fig. 5. To indicate horsepower directly, the output of an electric tachometer can be fed into the circuit. Torsional vibration as well as static torque can be measured readily.

The gages can be arranged to measure fluid and gas pressures just as easily. Pressure changes in a pipe are accompanied by extremely small changes in pipe diameter. Wire gages, measuring in millionths of an inch readily pick up these hoop strains in terms of fluid pressure. Figure 6 is the gage arrangement, wherein gage A and gage D on opposite sides of the pipe are sensitive to diameter change; gages B and C (C on rear of pipe) cemented longitudinally are unstressed, except to the extent of Poisson's ratio, and serve as temperature compensators. This type of pressure surges and explosions as well as static pressures. Diesel engine indicators have been made in this way as well as gun barrel pressure measurements.

#### Measurement of Residual Stresses

A unique use of gages of interest to foundrymen is in the measurement of residual stresses. Again the application is simple and straightforward. If a gage be fixed to the part to be investigated and a small disc or piece be cut from the part directly under the gage by drilling small holes in a circle around the gage, the strain relaxation indicated by the gage when the small piece is separated from the body of the part is a measure of the locked-up strain.

When gages are wound on paper impregnated with phenolic resin and mounted with bakelite cement instead of the usual Duco or cellulose nitrate cement, they can be used at elevated temperatures up to 500 F if the exposure is not too long, or for slightly lower temperatures for long duration exposure.

Then there are dual lead gages especially good for severe vibratory conditions, spiral gages for use on diaphragms, bi-axial gages that measure stress directly and many other forms including some wound on a ceramic base, instead of paper, good dynamically up to 1500 F and others made of special ductile materials good up to 20 per cent elongation.

Finally there are the multiple gage groups known as rosettes, for determining not only the magnitude of principal strains but their direction as well, determined by extremely simple graphical solution. Gages so used are mounted together in a fan or delta arrangement, Fig. 7.





(Courtesy, Baldwin Locomotive Works, Diesel Locomotive Dept.)

Fig. 8—Testing a permanent mold casting, a diesel locomotive piston with strain gages.



(Courtesy, Edward Valves, Inc., East Chicago, Ind.)

Fig. 9—Strain gage testing a large valve. Man with pump at left is simulating pipe line stresses. Man with pump at right is applying internal line pressure to valve. Man at instruments is reading strains in casting.

Auxiliary equipment to switch, indicate and record strain gage bridge unbalance has been developed commercially in adequate variety with good characteristics and in package units, so it is only necessary to connect bridge leads to marked binding posts. Many gage users devise their own circuits and auxiliaries and advice is available from many quarters.

It would seem that enough of the bare fundamentals of this bonded wire gage technique have been cited to stimulate the imagination of engineers and suggest solutions to problems which heretofore have seemed difficult. Experience indicates this has been the case so frequently in the recent past. Ingenious adaptations in experimental stress analysis, in built-in measuring devices, schemes for regulation and control, and tens of thousands of clean-cut engineering uses in all fields from medical research to the solution of routine industrial problems have been enthusiastically described in current technical literature. It illustrates the axiom that simple things are best, for nothing could be simpler than the bonded wire gage and nothing has enjoyed a more varied usage.

A few fields of engineering have not yet done much with the gage. The foundrymen's is one of these in which we can find little evidence of much application, although an increasing activity is noted of late, revealed in an official search for strain gage data on castings.

If the reader of this paper has not by this time captured a mental picture of a service this gage could perform, the first use that would occur to the author would be in the design improvement in castings. There is talk these days of reducing the height of ribs and spreading them out or making a real effort to do without them at all. There is the flange and bolt problem where the gage has served well, the residual stress studies, the studies of castings in service, impact, fatigue, stress raiser elimination, the detection of "designed-in" stress concentrations, a study of stress peaks set up by temperature differentials, the breakdown tests of castings, and in hosts of other problems.

#### Design Improvement

The matter of design improvement is one that has already come a long way. Eliminating assumptions and guess work has paid off. Making structures stronger by making them lighter was exemplified in aircraft and automotive designs and is now invading the railway field. Correcting empirical formulae from experimental studies, and reducing dependence on the "factor of safety" in design are processes as precise as they are varied.

In a few words we must try to outline a procedure for design improvement. It is this:

1. A casting, fabricated part, component or complete structure, designed analytically using overlapping assumptions, is to be improved.
2. The part is tested in service with strain gages applied so as to indicate service loads. Heretofore in most cases service loads have been guessed at.
3. A static test of the part or structure is made applying the service loads previously determined,

using a testing machine, or hydraulic jacks, or other conventional loading means. This test is to determine whether all members are taking their share of the load and if not to correct them by redistributing material and stresses. It is usually more effective to reduce stress peaks by redistributing stresses without increasing weight than to "beef-up" and let the stress concentrations ride.

An accepted method of experimental stress analysis is the use of stresscoat, a brittle lacquer which provides a crack pattern under load that not only gives a qualitative overall stress picture but points out where to put wire strain gages. These latter will provide the quantitative values at points of maximum stress as needed for design improvement.

4. Having logged the stress level in a part or structure it is next in line to determine whether this level is safe or dangerous under dynamic conditions of simulated service. A life test or fatigue test of the parts or components evaluates several things:—

- (a) The notch sensitivity of the material which points out the safeness of the stress level in the part.
- (b) Shows up design flaws or stress raisers.
- (c) Evaluates workmanship in connections, discontinuities and surface finish conditions.
- (d) It furnishes a dynamic proof test, proving adequate fatigue strength for the service intended, providing a good load analysis has been made and the part is loaded as it will be in service.

As to castings, if it is always borne in mind that stress flows in a part like fluid in a duct, a visual examination of contours will give some idea whether stress can be poured smoothly through the casting or whether turbulence (stress concentrations) will exist. And while looking at a casting why not try to visualize where the neutral axes are likely to be located? They sometimes give a hint as to how to redistribute material. A casting is trying to tell you where it is in trouble if you will give it heed.

#### DISCUSSION

*Chairman:* C. E. NELSON, Dow Chemical Co., Midland, Mich.

*Co-Chairman:* M. H. YOUNG, Wright Aeronautical Corp., Wood Ridge, N. J.

D. LAVELLE:<sup>1</sup> If it is not possible to use a dummy for temperature compensation, can a thermocouple be placed on the bar and temperature correction applied to the gage reading?

MR. TATNALL: Yes, a mathematical correction is possible but tedious. The purpose of the gages is not only to get mass measurements quickly but to do it with the least amount of figuring. A thermocouple correction on hundreds of readings becomes unthinkable, when automatic temperature correction methods can be so easily attained. If test data is too involved it may go into the file and a set of test data in the files is worthless.

CHAIRMAN NELSON: Everyone may not know what equipment is required to do stress analysis work. Perhaps that could be expressed in dollars and cents for the minimum equipment and facilities. One might get the idea that this work is very complicated but it is not. So that we might look to the future when almost all companies actually making mechanical devices and even foundries will have such equipment, does the author care to discuss that?

MR. TATNALL: In strain gage work you can start for about \$500. The self-contained strain indicator costs about \$450. The

<sup>1</sup> American Smelting and Refining Co., Barber, N. J.

strain gages themselves cost \$1.45 each. That is all that is required to start with. Get a good man to do the work and interpret the data and you have to pay him. Modern testing is now testing parts and components instead of specimens.

If a structure exhibits weak points, it is better to cut the strong points down rather than to strengthen the weak points. In other words, you can frequently strengthen by weakening, as could be very clearly shown by specific examples. Many companies are showing that a rib is a guy wire, and if you make ribs flatter and broader you improve the casting.

D. L. COLWELL:<sup>2</sup> Would it be possible to apply that strain gage to a die-casting die to measure the pressure on the die as the metal goes into it? The distortion of the die should be proportional to the pressure, but the die is naturally very heavy in comparison with the volume of the casting. Would the gage be sensitive enough to measure the metal pressure at different portions of the casting?

Mr. TATNALL: The part of the structure on which a gage is used must have a small enough cross section so that you get strain that is measurable. If a rapid temperature rise takes place you have the old question of temperature compensation again which must be intelligently accounted for. But it would be quite possible to find a place on the die which would deform a few millionths of an inch.

Co-CHAIRMAN YOUNG: How carefully must you control the diameter of this wire? If it is 0.001 in. in diameter it is going to be sensitive as to cross-section area variations.

Mr. TATNALL: The diameter of the wire does not bother us as much as the metal structure. We have a quality control method for inspecting gages. When we find the gage factors on a number of gages irregular we throw the spool away. It is a combination of the diameter and the metal structure and the composition of the wire which determines how uniform the gage factor is. With the gage factors showing up uniformly on every twentieth gage selected at random we do not worry about diameter except as it relates to gage resistance. That is, a certain length of wire must be of certain dimensions to give you a certain resistance and we want to get 120 ohms in 5 in. The wire is redrawn in diamond dies before use. We are thinking only in terms of gage factors and gage resistance.

Co-CHAIRMAN YOUNG: I had no thought of the composition or structure of the wire. I was thinking only of the diameter. I can see now that maybe the other factor is more important.

You can apparently fasten the gages by gluing them to any metallic surface. You do not have to take the skin off the casting, for example.

Mr. TATNALL: No, it just has to be smoothed up enough to make a cementing surface. You do not have to put the gage on as we showed you here. You can wind the wire directly on the casting if you want to. The only reason the paper is used is to carry the wire around in its grid. In some cases, especially with the high temperature gages, we wind the wire directly on the part. In case you have a very rough surface you should get it just smooth enough so it will really stick the gage down because if you take the skin off the casting you have a different condition.

Co-CHAIRMAN YOUNG: That is right, you could not measure residual stresses very well.

Mr. TATNALL: About the skin of the casting, it has to be just smooth enough to tack the gage down along its whole length. You can put these gages on concrete, wood, plastics, textiles, anything that will hold du Pont Duco cement.

MEMBER: Do you have to worry about sneak currents when you wind it directly on the casting?

Mr. TATNALL: Yes, we must have at least 50 million ohms resistance between the wire and the metallic surface to which it is bonded. That insulation is usually provided by plastic cement. 50 megohms is the minimum resistance necessary; usually 200 million ohms are desired. Sneak current is a good name for it. We call it leakage. If there is leakage it looks the same as strain on the indicator. For instance, sometimes it looks like the plastic is creeping whereas it is not creeping at all. Moisture is probably weakening the insulating ability of the cement and current is getting through. Waterproofing is very important. Waterproofing can be done with various kinds of waxes, such as Petrosene wax or various rubber compounds and roofing cements. Almost anything has been used with varying success to waterproof these gages. They have been kept under water for 30 months without losing their effectiveness if waterproofed by Petrosene A wax, covered by Ozite B, which is a General Cable product, a kind of roofing cement. The whole thing is a matter of technique. You say, "What is technique?" and we say "Learn by doing." In other words, you just try it yourself. You will not find it expensive to try a few of these things. I might say we have learned much of our technique from our customers.

<sup>2</sup> Apex Smelting Co., Chicago.



# EFFECT OF THE COMMON ALLOYING ELEMENTS ON THE TENSILE PROPERTIES OF MALLEABLE IRON

By

H. A. Schwartz and W. K. Bock \*

## ABSTRACT

The effect of alloying elements on the tensile properties of malleable iron is so much obscured by other operating variables that any deductions from routine tests and analyses are open to suspicion. Direct experimental exploration indicates that, as expected, any alloy dissolving in ferrite strengthens it. The effects of successive increments in the concentration of a given element or of the presence of several elements are not directly additive but decrease as the total alloy increases. The effects of several alloys can, however, probably be approximated by converting all into equivalent amounts of a single alloy. Silicon seems to exert an effect on graphite form which may obscure its effect in the matrix.

IT IS WELL ESTABLISHED that the tensile properties of completely graphitized malleable iron depend on the properties of the metallic matrix, and the amount and form of the included graphite. The effect of a unit of weight of graphite has been found in various studies to vary greatly, and it probably is a fair first assumption that this is due to variations in the size and sprawliness of the nodules.

In the present discussion it is proposed to exclude the effect of graphite by suitable selection of experimental conditions, and to consider only the effects of the ever-present elements—silicon, manganese, sulphur, phosphorus and chromium.

One approach to the question of the quantitative effect of chemical composition on tensile properties of any alloy is the deductive one of computing, statistically, the partial correlation coefficient of each element's concentration and the properties in question (if all other variables correlated with each concentration be kept constant). For example, additions of ferrosilicon probably affect the mechanical properties in two ways; by affecting silicon content, and perhaps by deoxidation. We must, therefore, concern ourselves with the partial correlation of silicon and strength (or elongation), ferrosilicon additions being assumed constant.

Similarly we must consider the effects of sulphur and manganese, ferromanganese and other elements of this pair being kept constant. While a correlation can be demonstrated between sulphur and sprue content in

the mix, the effect of sprue in itself on physical properties is negligible and that variable may be disregarded in this instance.

If, by the accepted methods of statistics, a partial correlation coefficient of a given element's concentration and a given property can be calculated, it is a simple matter to compute the corresponding regression coefficient which states how much the given property is caused to vary by unit change (one per cent) in the given concentration. It will presently be seen that there is, in fact, little or no consistency to the regression coefficients computed from different populations necessitating detailed study of the standard errors of these coefficients, as computed by well established means.

Since the mathematics is complex, no attempt is made in this paper to illustrate the calculations or demonstrate their reliability. The reader is asked to accept the procedure and calculation as authentic. The data represent four groups, each of constant carbon. They represent the daily average values of the relevant variables. Three plants are represented.

Table 1 gives the several regression coefficients of tensile strength of each of the four populations and the average regression coefficient weighting each plant equally.

Not only are large differences in magnitude shown, but even occasional differences in sign. Are these to be considered as real or as the result of chance? One may judge from the standard error of Table 2. The significance of the symbol ° in Table 1 is described in the discussion following Table 2.

Only the values marked ° in Table 1 are two or more times the corresponding ones in Table 2, and

TABLE 1—REGRESSION COEFFICIENTS OF TENSILE STRENGTH ON CONCENTRATION (PSI/1 PER CENT)

Group	A	B	C	D	Avg.
C, per cent	2.50-2.52	2.53-2.55	2.56-2.58	2.49-2.51	
Days, no.	52	71	84	59	
Si, .....	-2470	+4420	-6400	-17150	-7250
Mn, .....	+15800°	+4020	+81	+157	+3383
P, .....	+4930	+44500°	+27500°	-19500	+24220°
S, .....	+23400	+31800	+46000°	-5235	+22788
Cr, .....	+34150	+17900	+15450	+24700	+17187

\* Respectively Manager of Research and Research Engineer in the Research Laboratory of the National Malleable and Steel Castings Company, Cleveland, Ohio.

therefore have a reasonable certainty of being real and in the observed direction (2.5 per cent chance of accident).

These very large standard errors have the effect of making necessary very large differences in Table 1 between the regression coefficients in different populations before these differences are certainly real. Of all these differences only that between the regression coefficients of tensile strength on phosphorus for *B* and *D* can be taken as certainly real. The general significance is that the scatter of values due to any particular element is so small compared to the total scatter that we are unable to get any data from any of our single populations sufficiently accurate to warrant carrying over into another.

In such cases one hopes for better precision from average values. Comparing the averages of Table 1 with their standard error in Table 2, it is seen that the chance that the regression coefficient on silicon is positive instead of negative, i.e., that silicon actually does not weaken the iron, as it seems to do, is (from the appropriate probability tables) about 0.062, or 62 in 1000. Similarly, the chances that the other elements do not strengthen the iron are: manganese, 0.195; phosphorus, 0.0005; sulphur, 0.015, and chromium, 0.145. The chances are, therefore, reasonably good, but convincing only in the case of phosphorus, that silicon decreases tensile strength and all the other elements increase it. The average values of Table 1 are the best quantitative estimates of their effects, but except for phosphorus and sulphur precision is low.

Tables 3 and 4 have the same relation to elongation as do Tables 1 and 2 to tensile strength. The meaning of the symbols \* is as previously described.

It is at once evident that none of the values in Table 3 is numerically as great as the corresponding numbers in Table 4, and hence it cannot, by that test, be stated with certainty that composition affects elongation. Even the standard errors of the averages are large. The probabilities that silicon and chromium do not decrease elongation are 0.295 and 0.450, respectively, and that the other elements do not increase it are 0.255, 0.315 and 0.160 for manganese, phosphorus, and sulphur, respectively.

The conclusions, such as they are, are valid only within the range of composition for which the data were gathered. This represents a region of the composition shown in Table 5. The reason for this restriction will be further evident later in the paper.

The great uncertainties in the conclusions of this bulletin arise out of the impossibility of excluding in any commercial process all the variables except a very few. They could presumably not be eliminated by any method which did not secure greater constancy with regard to even the unknown factors. It is scarcely feasible to operate for an extended period at, say, a low level of phosphorus, and for another at a high level, and still less so for elements like silicon whose usable range is limited. Operating with furnace, or ladle, additions to part of each day's run introduces a new and unknown variable; thus it would seem that the possibilities of improving the data on which such studies might be based are exhausted.

An inductive approach based on direct experiment may also be attempted by measuring the effects of the

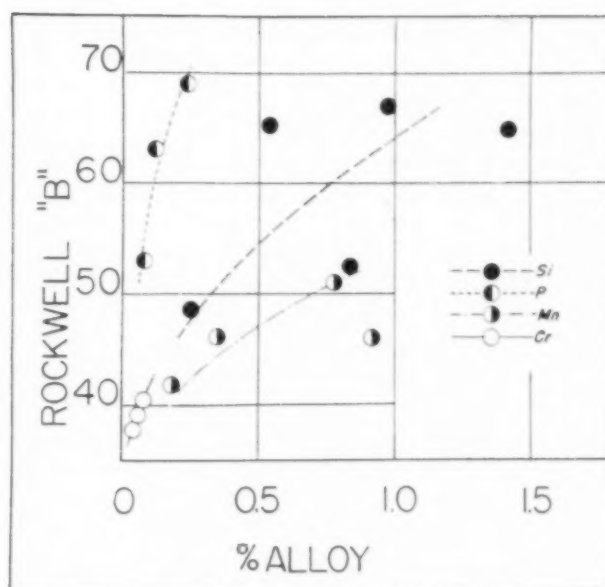


Fig. 1—Effect of various amounts of alloying elements on Rockwell "B" hardness values of malleable iron.

TABLE 2—STANDARD ERRORS OF REGRESSION COEFFICIENTS OF TENSILE STRENGTH (PSI/1 PER CENT)

Group	A	B	C	D	Avg.*
C, per cent	2.50-2.52	2.53-2.55	2.56-2.58	2.49-2.51	
Days, no.	52	71	84	59	
Si.....	7432	5746	7284	10183	4719
Mn.....	7824	7540	5489	5242	3932
P.....	4910	9466	12496	16390	7277
S.....	21097	18262	6122	23919	10523
Cr.....	46594	36366	20076	15222	16170

\* Computed as  $\frac{1}{2}$  the square root of the sum of the squares of the standard error for each group and representing the standard error of the averages of Table 1. NOT the average of the standard errors of the averages of Table 1.

TABLE 3—REGRESSION COEFFICIENTS OF ELONGATION ON CONCENTRATION (PER CENT/PER CENT)

Group	A	B	C	D	Avg.*
C, per cent	2.50-2.52	2.53-2.55	2.56-2.58	2.49-2.51	
Days, no.	52	71	84	59	
Si.....	-5.47	-1.65	+3.40	-2.11	-0.76
Mn.....	+6.95	+4.38	+0.17	-0.43	+1.80
P.....	+1.10	-3.29	+2.17	+11.1	+2.77
S.....	+32.8	-2.15	-6.77	+11.65	+6.73
Cr.....	+18.0	+14.6	-0.65	-21.7	-2.01

\* See footnote of Table 2.

TABLE 4—STANDARD ERRORS OF REGRESSION COEFFICIENTS OF ELONGATION (PER CENT/PER CENT)

Group	A	B	C	D	Avg.*
C, per cent	2.50-2.52	2.53-2.55	2.56-2.58	2.49-2.51	
Days, no.	52	71	84	59	
Si.....	4.77	4.46	2.49	9.86	3.72
Mn.....	5.54	6.81	3.02	4.99	2.71
P.....	3.23	8.97	3.34	15.79	5.76
S.....	14.94	16.49	3.37	10.10	6.72
Cr.....	30.58	30.76	22.72	14.53	13.6

\* See footnote of Table 2.

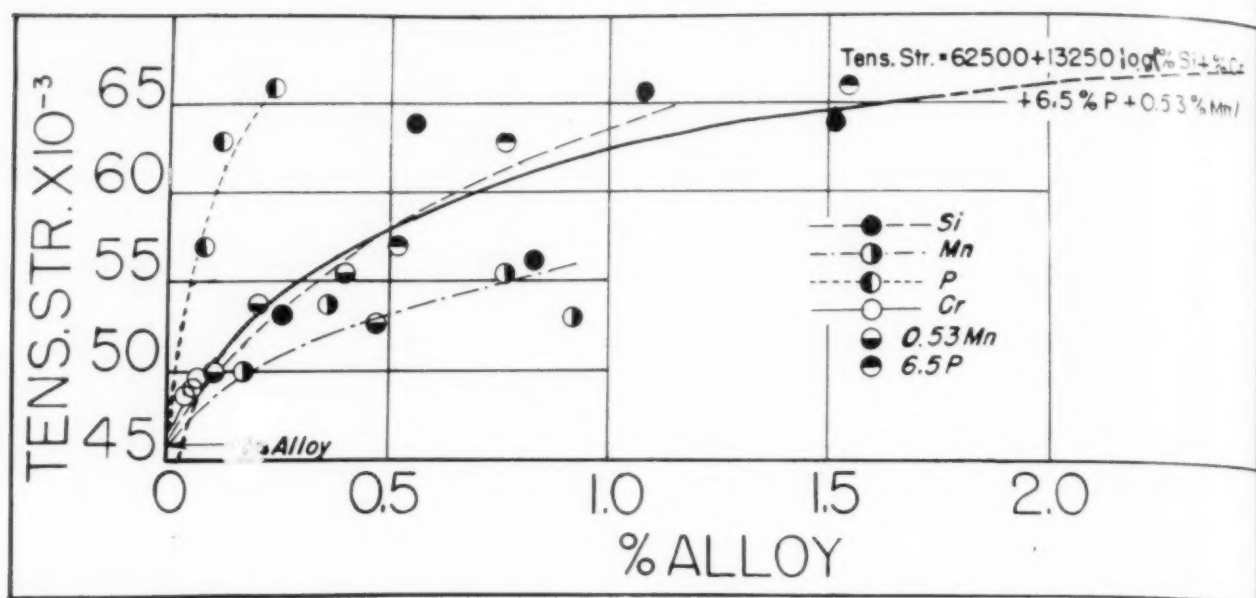


Fig. 2—Malleable iron tensile strength curves showing effect of various percentages of alloying elements.

several alloying elements separately. It is entirely reasonable that all four elements, when dissolved in ferrite, should strengthen the latter, and hence malleable iron. The present experiments were conducted to obtain some direct confirmation of this assumption in somewhat quantitative terms.

Armco iron was melted in vacuo in zircofrax crucibles with suitable amounts of ferrophosphorus, pure manganese, pure chromium or ferrosilicon to give alloys of the desired chemical composition. The melts were allowed to solidify in vacuo and were found to freeze with very large grains, some almost 1 cm in diameter. By quenching from above their  $A_3$  points and renormalizing from above  $A_3$ , they were given a (ferrite) grain size comparable with that of malleable

TABLE 5—MEAN COMPOSITION AND ITS STANDARD DEVIATION

Group	A	B	C	D
C, per cent. ....	2.50-2.52	2.53-2.55	2.55-2.58	2.49-2.51
Si, per cent. ....	0.99	0.99	0.98	1.08
$\sigma$ Si, per cent. ....	0.041	0.041	0.037	0.029
Mn, per cent. ....	0.51	0.51	0.44	0.37
$\sigma$ Mn, per cent. ....	0.053	0.054	0.025	0.054
P, per cent. ....	0.119	0.116	0.150	0.125
$\sigma$ P, per cent. ....	0.056	0.0181	0.0232	0.0169
S, per cent. ....	0.174	0.173	0.156	0.098
$\sigma$ S, per cent. ....	0.0214	0.0239	0.0239	0.0118
Cr, per cent. ....	0.035	0.036	0.031	0.027
$\sigma$ Cr, per cent. ....	0.0059	0.0050	0.0108	0.0181

TABLE 6—TUKON HARDNESS OF FERRITE

Heat	Alloy	Rockwell "B"	Knoop
2	None	30.5	135
7	1.42% Si	64.8	187
9	0.24% P	68.9	182
12	0.78% Mn	50.9	142
20	0.06% Cr	39.3	158

iron. The ingots were about  $\frac{3}{4}$  in. in diameter and the heights ranged from one to one and one-half in.

Rockwell "B" hardness was determined at numerous points on several cross sections of each small ingot and the results averaged.

In Fig. 1 these Rockwell "B" numbers are plotted as a function of the alloy content. It is reasonable, since the effect of a unit of any elements depends on the amount present, that the effect of two or more alloys was probably not directly additive. As a check on this an alloy was made containing 0.86 per cent silicon and 0.111 per cent phosphorus, which alloy had a Rockwell "B" number of 76.

In the phosphorus curve of Fig. 1 at 0.111 per cent phosphorus will be found a Rockwell "B" number of about 59, which is equivalent to a silicon content of about 0.7 per cent. Adding this silicon content to the observed value of 0.86 gives a total of approximately 1.56 per cent. The extrapolated silicon curve at this value shows a hardness of about 73. It appears, therefore, that this means of summing the effect of more than one alloy yields results about as concrete as the curves.

As a check on the relation of Rockwell "B" to Knoop hardness, four of the alloys in Fig. 1 were tested on the Tukon hardness tester with the results shown in Table 6.

The results shown in Table 6 scatter rather badly but are roughly equivalent to the statement that the Knoop hardness is equal numerically to one and one half times the Rockwell "B" hardness plus 85.

Assuming that a malleable iron contained 0.15 per cent phosphorus, 1.00 per cent silicon, 0.25 per cent manganese (in excess of manganese sulphide) and 0.03 per cent chromium, this is equivalent to about 2.08 per cent silicon by the calculation made in the foregoing. At 2.08 per cent silicon, extrapolating the silicon line of Fig. 1, should give about 78 Rockwell "B" hardness, which by the equation correlating Rockwell and Knoop numbers should be something over 200. The Knoop numbers of the ferrite of malleable iron range from 220 to 250, which seems a fair agreement with the present work. These hardness values were converted



to tensile strengths by the data of Fig. 98 of Williams' "Hardness and Hardness Measurements. It may perhaps be argued that this conversion is not too reliable. The data are given in Fig. 2.

The slopes of the curves at 0.15 per cent phosphorus, 1.0 per cent silicon, 0.25 per cent manganese and 0.06 per cent chromium, approximate values in malleable iron, were measured and found to represent an increase of 47,500, 7,500, 12,500 and 32,500 psi for an increase of one per cent in the several elements in that order. All values are somewhat lower than those given in Bain's "Alloying Elements in Steel," Fig. 40.

It should also be noted that all of these values are higher than those approximated in the deductive approach. The latter fact is perhaps best explained by the idea that the present figures relate to the properties of approximately pure iron alloys, while the former relates to the effect of each element in the presence of the average amount of the three others (and perhaps of sulphur).

Consulting Fig. 2 it is found that 0.15 per cent phosphorus (a usual value) is equal in strengthening effect to 0.88 per cent silicon; 0.25 per cent manganese (a probable excess above sulphur) is equal to 0.17 per cent silicon, and 0.05 per cent chromium to 0.09 per cent silicon. Hence a 1 per cent silicon ferrite containing these amounts of other alloys might be equivalent to 2.14 per cent silicon and no other alloy where the curve would be much flatter. The general trend line of Fig. 2 has only empirical significance. It has been seen that the hardness of a ternary silicon-phosphorus-iron alloy bore out this idea.

Here is experimental support for the assumption that silicon strengthens ferrite. If it nevertheless weakens malleable iron, this is an indication that silicon affects the form of graphite nodules.

It seems a reasonable assumption that a given change in composition will produce the same proportional change in ferrite and in malleable iron. Ferrites average perhaps 15 per cent greater strength than malleable iron, so that the effect of changes of composition on malleable iron should be about 85 per cent of the values given (in psi).

The survey made has not been sufficiently extended to attempt an extrapolation from the data on ferrite into the region of complex compositions corresponding to the ferrites of malleable iron to justify a comparison on a quantitative basis with the values reached by deduction.

### Conclusions

Based upon the foregoing discussion, the authors consider the following conclusions to be justified.

All the alloying elements strengthen the matrix of malleable iron.

Silicon probably increases the sprawliness of graphite sufficiently to overbalance its strengthening effect on the matrix.

The effect of a unit of any alloy decreases progressively as the amount of that alloy increases.

The cumulative effect of all the alloys can be approximated by converting the several concentrations into an equivalent amount of a single alloy.

Very roughly for such conversion, silicon and chromium are of equal potency, phosphorus is equivalent to about six and one half times as much silicon, and

a unit of manganese to about one half a unit of silicon.

Changes in concentration in a metal containing as much alloy constituent as malleable iron cannot be expected to produce great changes in strength.

### Acknowledgment

The authors wish to express their appreciation to J. V. Emmons, Cleveland Twist Drill Co., for providing facilities for making the Tukon hardness tests, and to J. V. Anthony, of this laboratory, for the calculations.

### DISCUSSION

*Chairman:* W. D. McMILLAN, McCormick Works, International Harvester Co., Chicago.

*Co-Chairman:* W. B. McFERRIN, Electro Metallurgical Co., Detroit.

*CHAIRMAN McMILLAN:* Silicon should strengthen the ferrite. It is common practice to add silicon to low-carbon steel to raise the elastic limit when it is not permissible to increase carbon to obtain the same effect. Phosphorus increases the strength and elastic limit in Bessemer over an open hearth steel of the same carbon content. In malleable duplex operations, using as high as 40 per cent steel in the charge, it has been the practice to add phosphorus to increase the yield, particularly in cases of a long anneal. The cumulative effect of alloy elements may well account for some of the short breaks and low elongations that are not readily explained from the standpoint of structure.

With regard to the increase in sprawliness of the graphite nodules with an increase in silicon, our experience has been that with an increase in silicon, we will have a greater number of graphite nuclei than with a lower silicon. It may be that the sprawliness observed is a matter of annealing practice as well as a matter of silicon content.

*J. J. CURRAN:*<sup>1</sup> What heat treatment was given the small vacuum-melted induction heats?

*MR. BOCK:* The heats were analyzed and then by reference to the alloy diagram for that system we found the Ac3 point, took the sample above the Ac3, water quenched it and took it above the Ac3 again and air cooled it.

*MR. CURRAN:* Did you attempt to simulate a regular malleable anneal?

*MR. BOCK:* No. We were trying to get a ferrite grain size comparable to the ferrite grain size in malleable iron.

*MR. CURRAN:* You find that the effect of the elements is additive in these particular alloys, whereas in steel, quenched and tempered, they multiply each other.

*MR. BOCK:* That is true. The elements are additive in terms of a single element.

*H. A. SCHWARTZ (authors' closure):* It appears that Mr. Curran has seriously misunderstood that portion of our paper which relates to the hardness of ferrite with various elements added to it. These alloys are not malleable iron but are an exceedingly low-carbon steel to which the desired concentration of silicon, manganese and chromium were added. Consequently there is no such thing as simulating a regular malleable anneal. Unless the special heat treatment described by Mr. Bock in the oral discussion is applied, these steels are of enormous grain size. The heat treatment we used produced a ferrite grain size the approximate equivalent of that in malleable iron. We then have a ferrite comparable to that of malleable as to grain size and containing only those alloys in solution in which we happen to be interested.

Mr. Curran further misunderstands, perhaps when he says that the effect of our elements is additive. It is true, in fact, that they are additive, but not linearly additive; that is, equal increments do not produce equal increments in hardness. If the alloys multiply them, the effect of a given increment would increase with the amount of the alloy already present, whereas the reverse is true. We are thus not in accord with Mr. Curran's assumption that in steel certainly if the material is quenched and drawn you would find that they multiply each other. Has there been confusion here between the effect of alloys on hardness and on hardenability? The two certainly are not the same thing.

<sup>1</sup> Walworth Co., Greensburg, Pa.

# SOME PRINCIPLES INVOLVED IN HEAT TREATMENT OF GRAY CAST IRON

By

Alfred Boyles \*

IT IS THE PURPOSE of this paper to review, in an elementary manner, some of the fundamental principles involved in the heat treatment of cast iron, and to point out certain basic differences between this material and steel as regards transformation behavior during heating and cooling. It has become the custom in recent years for metallurgists to disregard the iron-carbon diagram in discussing the heat treatment of steel, and to interpret results in terms of "S" curves, or time-temperature-transformation diagrams.

This does not mean that the iron-carbon diagram is no longer of importance, but rather that it has become so completely a part of the mental processes of the metallurgist that further reference to it is not needed. The practical interpretation of data is based on rates of reaction in the solid state with the tacit assumption that everyone knows all about the conditions of equilibrium in the iron-carbon system.

In the case of plain carbon steel, this attitude is probably justifiable. In the case of cast iron, however, the metallurgist soon finds that the conditions of equilibrium are not so well known and that they are quite different from those in steel. To illustrate some of these differences, a parallel will be drawn in the following discussion between the behavior on heating and cooling of a plain carbon pearlitic steel of eutectoid composition (about 0.80 per cent carbon) and a plain cast iron of nominal analysis having a pearlitic matrix.

## Transformation in Pearlitic Steel

Figure 1 represents a part of the ordinary iron-carbon diagram showing the metastable equilibrium between iron and iron-carbide. The vertical dotted line represents the composition under consideration, a pearlitic steel of 0.80 per cent carbon. At room temperature the metal consists of two phases, ferrite ( $\alpha$ ) and iron carbide ( $Ca$ ) lying side by side as pearlite lamellae. Upon heating past the horizontal line  $AB$ , the iron transforms into the  $\gamma$  phase and absorbs the carbide lamellae to form a solid solution of carbon in iron called austenite. If time is allowed this austenite becomes a homogeneous single phase  $\gamma$ . At no time does the carbon content of the matrix undergo any change. There is never more or less than 0.80 per cent carbon present.

Upon cooling slowly the reaction is reversed. Austenite breaks down into alternate layers of ferrite ( $\alpha$ )

and iron carbide ( $Ca$ ) to re-establish the pearlitic structure. Cooling somewhat more rapidly will delay the transformation and cause it to occur at a temperature below the line  $AB$ . The pearlite lamellae then become finer, or more closely spaced. Still more rapid cooling may suppress the transformation temperature to as low as 1200 F, and extremely fine pearlite lamellae will be formed.

Drastic cooling, such as quenching, will suppress the lamellar type of transformation completely and allow the austenite to persist down to a temperature below 500 F, at which time the austenite breaks down into an entirely different type of structure called martensite, which is characterized by a needle-like or acicular pattern. Between the temperature range for the formation of martensite and that of fine pearlite, a variety of acicu-

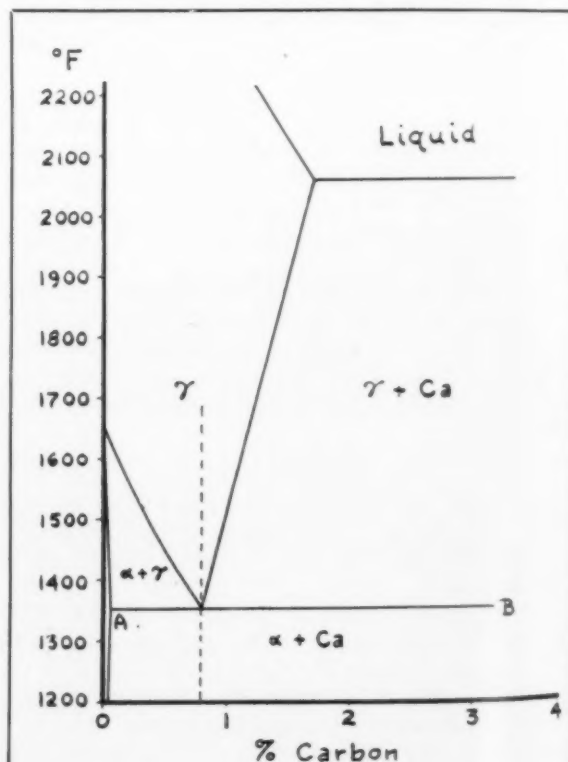


Fig. 1—Part of the ordinary iron-carbon diagram showing metastable equilibrium between iron and iron-carbide.

\* Research and Development Dept., United States Pipe & Foundry Co., Burlington, N.J.

lar structures form, depending on the amount of time allowed and the rate of transformation at any particular temperature level.

Since the physical properties of the steel depend on the type of structure obtained on cooling, the metallurgist is naturally much concerned with the rate of formation of these various structures and interprets his heat treating results as time-temperature-transformation diagrams rather than in terms of the simple equilibrium diagram shown in Fig. 1.

### Transformation in Pearlitic Cast Iron

Before entering into a discussion of the corresponding transformation in cast iron, it is necessary to give some consideration to the two types of equilibrium which occur in cast iron, the stable and the metastable. Gamow<sup>1</sup> has used a very clear analogy to illustrate what is meant by these basic terms, and his description and illustration (Fig. 2) are given as follows:\*

"In order to understand this important concept, suppose, for example, that we pour a quantity of water into a deep hole in the ground (Fig. 2A). If no external work is done to extract this water from the hole,

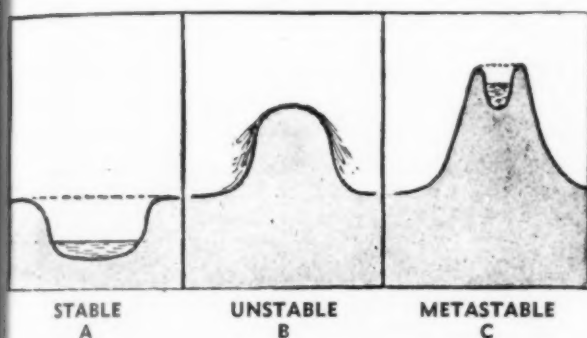


Fig. 2—Gamow's mechanical interpretation of the notion of stable, unstable, and metastable states.<sup>1</sup>

it will remain there indefinitely and will never get out by itself. We say the water in such a hole is in a *stable* state of equilibrium.

"If on the contrary, instead of pouring the water into a hole, we pour it on top of a sugar loaf mountain (Fig. 2B), it will not stay there at all and will at once come down in a stream on both sides. The position of the water on the top of such a mountain is *unstable* and can never be realized in practice.

"Let us now consider the water filling a crater of an old volcano (Fig. 2C). If left to itself, the water will stay there just as long as in the hole in the ground, although in principle much energy can be obtained from bringing it to a lower level. We need, however, to bring the water over the rim of the crater, supplying for this purpose a certain amount of energy from the outside. We say that the water in the crater is in a *metastable* state."

The stable state in all iron-carbon alloys is one of iron and graphite. The metastable state is one of iron and iron-carbide. In steel, the metastable state persists indefinitely under ordinary conditions of heating and cooling, and the metastable diagram shown in Fig. 1

is adequate to define its transformation behavior. Gray cast iron, however, because of its much higher carbon and silicon content, exhibits both the stable and metastable states under ordinary rates of heating and cooling. It is, therefore, necessary to construct two diagrams, one representing the stable condition and one the metastable.

Schematic diagrams of this kind are shown in Figs. 3 and 4. Complete equilibrium, according to either diagram, is seldom obtained under ordinary circumstances, the final result usually being a mixture of both. Using Gamow's analogy, part of the water in the volcanic crater may be considered as getting over the edge and reaching a stable condition. Some of the water, however, remains in the crater in a metastable state. This dual behavior is of fundamental importance and constitutes a basic distinction between cast iron and steel.

### Transformation According to the Metastable State

For this purpose a cast iron will be considered in which no free graphite occurs, i.e., under all conditions of heating and cooling the carbon is either in the form of iron carbide or in solid solution in the austenite. A diagram representing this metastable condition is shown in Fig. 3. The total carbon present is, say, 3 per cent, represented by the vertical dotted line.

At room temperature the metal consists of pearlite ( $\alpha + Ca$ ), together with massive particles of free carbide ( $Ca$ ), i.e., it is a white iron. Except for the presence of the massive carbide particles, the metal is exactly like the pearlitic steel previously described. Two phases occur: ferrite ( $\alpha$ ) and iron carbide ( $Ca$ ).

The metal is now heated into the temperature region marked  $\alpha + \gamma + Ca$  in Fig. 3. Some of the iron now transforms to the  $\gamma$  phase and begins to absorb the carbide of the pearlite lamellae to form a solid solution of carbon in  $\gamma$  iron (austenite), in exactly the same manner as the pearlitic steel behaved on heating. The massive free carbide remains undisturbed.

As the metal is gradually heated through the region  $\alpha + \gamma + Ca$ , the amount of austenite ( $\gamma$ ) increases and the amount of pearlite ( $\alpha + Ca$ ) decreases until, at a temperature just above the line  $CD$ , the matrix consists entirely of austenite holding in solid solution 0.70 per cent carbon (Point  $C$ ). At this stage, the massive free carbide still is intact and undisturbed.

### Carbon in Solid Solution

Suppose the temperature is raised considerably above the line  $CD$ . In this case the austenite begins to dissolve some of the massive carbide and the amount of carbon held in solid solution increases. At any temperature above the line  $CD$ , the amount of carbon in solid solution in the austenite may be determined by projecting horizontally from the dotted line to the left to intersect the line  $CE$  and then reading off the percentage of carbon on the scale at the bottom of the diagram. At 1800 F, for example, the austenite will hold in solid solution about 1.15 per cent carbon. The remainder of the 3 per cent carbon still is present as massive free carbide particles.

Upon slowly cooling the metal from 1800 F the reverse changes take place. Some of the carbon in solid solution in the austenite ( $\gamma$ ) is deposited again, either on the surfaces of the massive carbide particles or as separate particles until, at the temperature of the line

\* Used by permission of The Macmillan Company.



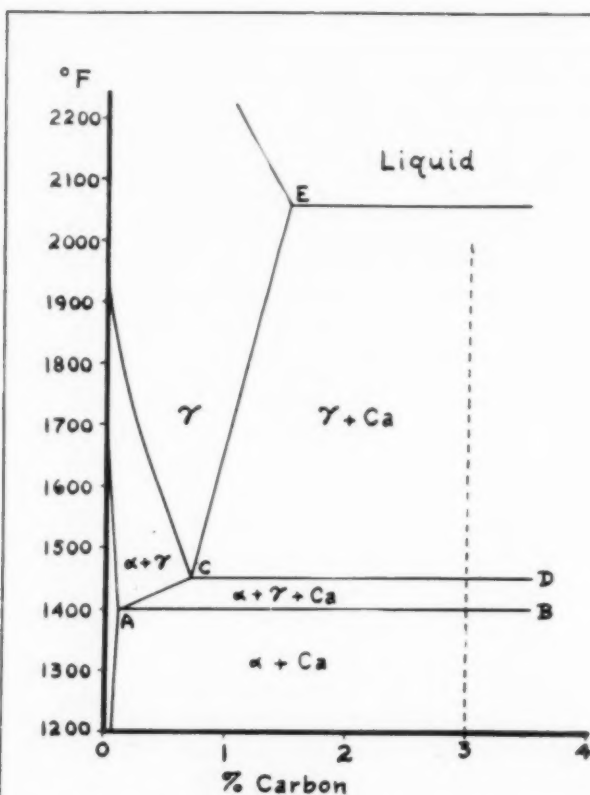


Fig. 3—Schematic diagram showing the conditions of metastable equilibrium in cast iron. Note that the lines are somewhat different from those in Fig. 1. These changes are caused by the presence of silicon in the metal. The eutectoid point "C" is placed at 0.70 per cent carbon and a new region  $\alpha + \gamma + Ca$  appears in the diagram. This diagram represents a silicon content of about 1.90 per cent.

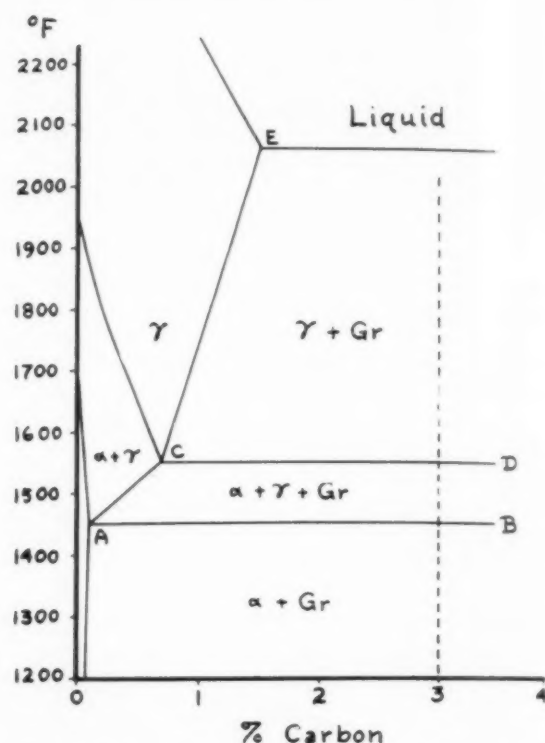


Fig. 4—Schematic diagram showing the conditions of stable equilibrium in cast iron. The eutectoid point "C" has been placed at 0.70 per cent carbon to correspond with Fig. 3. The region  $\alpha + \gamma + Gr$  is somewhat wider than the  $\alpha + \gamma + Ca$  region of Fig. 2. This diagram was drawn to represent a silicon content of about 2.20 per cent. All the lines will shift position with changes in silicon content.

CD, the austenite is again of composition C, holding 0.70 per cent carbon in solid solution.

When the metal cools further into the region  $\alpha + \gamma + Ca$ , some of this austenite ( $\gamma$ ) begins to break down into alternate layers of ferrite ( $\alpha$ ) and carbide ( $Ca$ ), forming pearlite. With falling temperature, the amount of austenite ( $\gamma$ ) decreases while the amount of pearlite ( $\alpha + Ca$ ) increases until transformation is finally complete at the line AB.

After crossing the line AB in cooling, the metal will be back again where it started and will consist of pearlite ( $\alpha + Ca$ ) plus massive particles of free carbide ( $Ca$ ). Under conditions of more rapid cooling, the matrix can be made to undergo all the structural modifications mentioned in the case of the pearlitic steel, i.e., finer and finer pearlite lamellae, or transformation into martensite or other acicular patterns, depending on the conditions of cooling. In all these cases the massive free carbide remains intact as a part of the final mixture without entering into the actual transformation in any way.

In describing the metastable transformation, it was assumed that no free graphite was formed at any stage of heating or cooling. The opposite extreme, the stable state, is a case in which no free iron carbide is present

at any temperature level. A diagram representing this stable condition is shown in Fig. 4. The similarity of this diagram to the metastable diagram (Fig. 3) is apparent, the only difference being that the notation  $Ca$  (carbide) has been changed to  $Gr$  (graphite), and that the position of the lines, with respect to temperature, has been altered somewhat.

#### Transformation According to the Stable State

To trace out the transformation according to the stable system, a start will be made with a cast iron of 3 per cent total carbon in a fully stable condition. The matrix in this case consists of nothing but ferrite ( $\alpha$ ) in which graphite flakes ( $Gr$ ) are imbedded. Upon heating into the region  $\alpha + \gamma + Gr$  (Fig. 4), part of the ferrite ( $\alpha$ ) transforms into austenite ( $\gamma$ ) and begins to dissolve carbon from the surfaces of the graphite flakes.

At successively higher temperatures, the amount of austenite ( $\gamma$ ) increases while the amount of ferrite ( $\alpha$ ) diminishes until, at the temperature of the line CD, the entire matrix consists of austenite holding in solid solution 0.70 per cent carbon (Point C in Fig. 4). The graphite flakes are now somewhat thinner than they were to begin with, having lost some carbon by solu-

tion to the matrix. At temperatures above the line *CD*, the austenite continues to dissolve carbon from the graphite flakes and will hold, at, say 1800 F, about 1.10 per cent carbon in solid solution.

Upon cooling slowly from 1800 F, the reverse action takes place. Carbon is deposited on the surfaces of the graphite flakes until the metal reaches the line *CD*, at which temperature the austenite again holds in solid solution 0.70 per cent carbon (Point *C*). The amount of carbon in solid solution at any intervening temperature may be determined by projecting to the left to intersect the line *CE* exactly as was done in the metastable diagram (Fig. 3).

At 1550 F (Line *CD*), the austenite begins to transform into ferrite (*a*) and graphite (*Gr*). This action is strictly analogous to the transformation of austenite into ferrite (*a*) and iron carbide (*Ca*) in the metastable system. The stable transformation, however, does not produce a lamellar structure of ferrite and graphite, similar to pearlite. The graphite formed during transformation is deposited on the surfaces of the graphite flakes.

As cooling continues through the region  $a + \gamma + Gr$ , the amount of austenite ( $\gamma$ ) decreases while the amount of ferrite (*a*) increases until transformation is finally complete at the line *AB* (1450 F). At any stage of transformation, the total amount of carbon remaining in solid solution may be determined by projecting to the left to intersect the line *AC*. At 1500 F, for example, about one half of the matrix will consist of austenite (holding in solid solution 0.70 per cent carbon) and the total amount of carbon in this remaining austenite will be about 0.35 per cent.

After cooling past the line *AB*, the metal will be back again where it started and will consist of two phases, a matrix of ferrite (*a*) in which the graphite flakes (*Gr*) are imbedded. The graphite flakes will now be back to their original thickness, all the carbon dissolved during heating having been restored during cooling and transformation.

#### Transformation in Cast Iron According to Both Stable and Metastable States

In the metastable transformation of cast iron, it was noted that the matrix could be made to assume all the structural variations produced in eutectoid steel by altering the conditions of cooling. This is not true of the stable system. The stable transformation produces ferrite and graphite, and the graphite so produced is invariably deposited on the surfaces of the graphite flakes present in the matrix. Coarse and fine lamellar structures or acicular structures, such as exist after metastable transformation, have no direct analogues in the stable system.

The term "stable transformation," as defined in the foregoing, must not be confused with the graphitization of pearlite. Careful studies have shown that the austenite during stable transformation breaks down into ferrite and graphite without the intervention of any metastable phase. This concept of stable transformation must be kept clearly in mind in all the discussion which follows.

With these two extreme cases of stable and metastable transformation in mind, we may proceed to the case where both types of transformation occur simultaneously in the same piece of metal, as frequently happens

in ordinary heat treatment. We will begin with a gray cast iron of 3 per cent total carbon having a fully pearlitic matrix.

At room temperature this material obviously represents both the stable and metastable states, the graphite flakes representing the stable state and the iron carbide of the pearlite lamellae representing the metastable state. Going back to Gamow's analogy (Fig. 2), part of the water in the volcanic crater may be considered as getting over the edge and reaching a stable condition. Some of the water, however, remains in the crater in a metastable state.

#### Reversion to Stable Condition

Upon heating the pearlitic gray iron to a temperature near the line *AB* in Fig. 4, the carbide of the pearlite lamellae will begin to graphitize. If time is allowed all the pearlite will decompose and the graphite so formed will be deposited on the surfaces of the graphite flakes. If a piece of cast iron is held for a considerable time at a temperature just below the line *AB* in Fig. 4 and then quenched, it will be found to consist entirely of ferrite (*a*) and graphite (*Gr*). The metal has reverted to the stable condition before any transformation to austenite has taken place.

Under ordinary rates of heating, much of the pearlite will remain intact until the metal reaches the transformation range. Transformation then begins exactly as described under the metastable system (Fig. 3). Some of the iron begins to transform into austenite ( $\gamma$ ) and absorbs the carbide of the pearlite lamellae. This action continues as the metal is heated through the region  $a + \gamma + Ca$  until, at a temperature above the line *CD*, the matrix will consist entirely of austenite ( $\gamma$ ) holding in solid solution 0.70 per cent carbon.

Upon heating to higher temperatures the austenite will dissolve carbon from the surfaces of the graphite flakes exactly as was described under the stable system (Fig. 4), and the amount held in solid solution at any temperature level may be determined as before by projecting to the left to intersect the line *CE*.

Suppose the heating process is stopped at some temperature within the region  $a + \gamma + Gr$  in Fig. 4, say at a temperature of 1500 F. If given time the metal will now revert to the stable state. The carbide of the pearlite lamellae will break down and deposit carbon on the surfaces of the graphite flakes. After holding at 1500 F for some time the matrix will then consist of about one half free ferrite (*a*) and one half austenite ( $\gamma$ ). A piece quenched in this condition will show a matrix of one half ferrite and one half martensite.

#### Quenching Temperature Effect

Any combination of free ferrite and martensite can be obtained by quenching from various temperatures within the region  $a + \gamma + Gr$  of Fig. 4. Quenching from temperatures near the lower line *AB* will produce nearly all ferrite. Quenching from temperatures near the upper line *CD* will produce nearly all martensite. To produce a fully martensitic matrix, the metal must be heated to a temperature above the region  $a + \gamma + Gr$ . It must be made fully austenitic before quenching.

When a pearlitic steel is heated to a temperature sufficient to become fully austenitic it is said to be austenized. If the surface of the steel loses carbon by chemical action during heating, it is said to be decarburized. If the surface absorbs carbon by contact with

carbonaceous matter during heating, it is said to be carburized. Applying these commonly used terms to pearlitic gray iron, its behavior on heating may be summed up in the following statements:

1. The matrix of pearlitic gray iron may be completely decarburized by heating it to temperatures just below the line *AB* in Fig. 4.

2. Metal so treated may be carburized again, up to a maximum of 0.70 per cent carbon, by heating it into the region  $\alpha + \gamma + Gr$  in Fig. 4, the degree of carburization being a function of temperature.

3. The matrix can be further carburized beyond 0.70 per cent carbon by heating it above the line *CD* in Fig. 4, the degree of carburization being again a function of temperature.

4. The matrix can be partially austenized by heating to temperatures within the region  $\alpha + \gamma + Gr$ , the amount of austenite so produced being a function of temperature. No amount of holding time within the  $\alpha + \gamma + Gr$  region will produce a fully austenitic condition.

5. The matrix can be fully austenized by heating to temperatures above the region  $\alpha + \gamma + Gr$  in Fig. 4.

The reason for these changes lies in the fact that gray iron reacts on heating according to both the stable and metastable states, and that it has its own internal carburizer provided in the form of graphite flakes scattered through the matrix.

#### Hardness and Quenching Temperatures

An example of what happens when cast iron is quenched from temperatures within the region  $\alpha + \gamma + Gr$  is shown in the work of Walls and Hartwell.<sup>2</sup> Their results are given in Fig. 5. Specimens quenched from 1100 F had about the same hardness as the original bars. Quenching from successively higher temperatures up to 1400 F caused a progressive decrease in hardness. This was caused by a gradual breakdown of pearlite into ferrite ( $\alpha$ ) and graphite (*Gr*) below the transformation range, the amount of such breakdown increasing with increasing temperature. The metal was in the region  $\alpha + Gr$  of the stable diagram (Fig. 4).

At 1450 F the hardness after quenching started to increase. Part of the ferrite ( $\alpha$ ) had now transformed into austenite ( $\gamma$ ) during heating, and this austenite formed martensite on quenching. The metal had now entered the temperature region  $\alpha + \gamma + Gr$  of Fig. 4. In the lower part of this  $\alpha + \gamma + Gr$  region only a small amount of austenite formed, and the hardness after quenching was still quite low. At 1500 F about half of the matrix was austenite ( $\gamma$ ) and the hardness after quenching reached 375 Brinell.

The metal at this temperature was in the middle of the  $\alpha + \gamma + Gr$  region. At 1550 F the matrix was nearly all austenite ( $\gamma$ ) and the ferrite ( $\alpha$ ) had almost disappeared. The metal was now at the top of the  $\alpha + \gamma + Gr$  region and developed a Brinell hardness of around 475 on quenching. Heating to 1600 F assured complete elimination of all the ferrite ( $\alpha$ ) and the matrix was fully austenized throughout. It was now in the region  $\gamma + Gr$  of Fig. 4 and gave full hardness on quenching. Specimens heated to 1650 F and 1700 F were likewise fully austenized and gave full hardness on quenching.

The point to be emphasized is that no amount of holding time in the region of  $\alpha + \gamma + Gr$  will produce a fully austenitic matrix. Specimens could have been

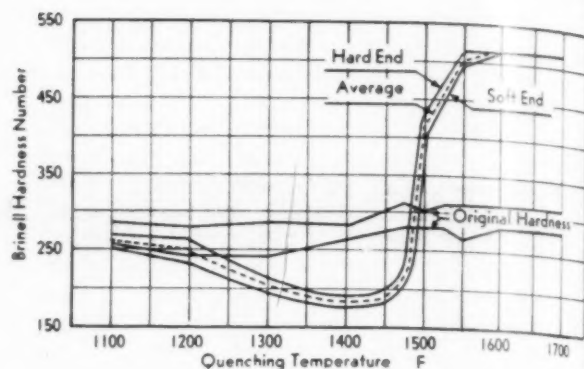


Fig. 5—Brinell hardness as affected by quenching temperature. Bars of 3-in. length by 5/8-in. diameter, water quenched. Analyses of bars follow:<sup>2</sup>

Component	Per cent	
Silicon	2.16	2.23
Total Carbon	3.39	3.28
Sulphur	0.055	0.049
Phosphorus	0.187	0.196
Manganese	0.84	0.71
Chromium	0.22	0.16
Molybdenum	0.68	0.78

held for hundreds of hours at 1500 F, for example, and would still have remained a mixture of ferrite ( $\alpha$ ), austenite ( $\gamma$ ), and graphite (*Gr*). The exact boundaries of this region of three-phase equilibrium are determined largely by the silicon content of the metal.

Figure 4 was drawn for a silicon content of about 2.20 per cent.<sup>3</sup> The precise limits of the  $\alpha + \gamma + Gr$  region for a wide variety of cast iron analyses have not been worked out experimentally. For practical purposes, it is sufficient to determine the lowest temperature from which a small specimen will give a fully martensitic matrix on quenching. This will define the upper limit (Line *CD*) of the  $\alpha + \gamma + Gr$  region for any particular analysis of iron. An iron of 1.40 per cent silicon may become fully austenitic at 1500 F, while one containing 3.50 per cent silicon may require a temperature of over 1700 F to reach a similar condition.

#### Mixed Stable and Metastable Transformation on Cooling

Both the stable and metastable transformations also occur under moderately slow conditions of cooling. In order for a cast iron to follow completely the stable transformation diagram shown in Fig. 4, it must be cooled very slowly. Ordinarily some metastable transformation to pearlite takes place, and the amount formed depends on the relative rates of the stable and metastable transformations in any particular composition.

To illustrate what is meant, we will start with a gray cast iron containing 3 per cent carbon and 2 per cent silicon, sand cast into bars one inch in diameter. The metal as-cast consists of type *A* graphite flakes in a fully pearlitic matrix. A bar is now heated to 1600 F, and held for one hour. The matrix then becomes austenite holding in solid solution about 0.80 per cent carbon, i.e., it is just above the  $\alpha + \gamma + Gr$  region in Fig. 4.

After soaking for one hour at 1600 F, the bar is cooled in air. Examination will now show that the matrix is no longer all pearlite. Some free ferrite will be found along the surfaces of the graphite flakes. Even



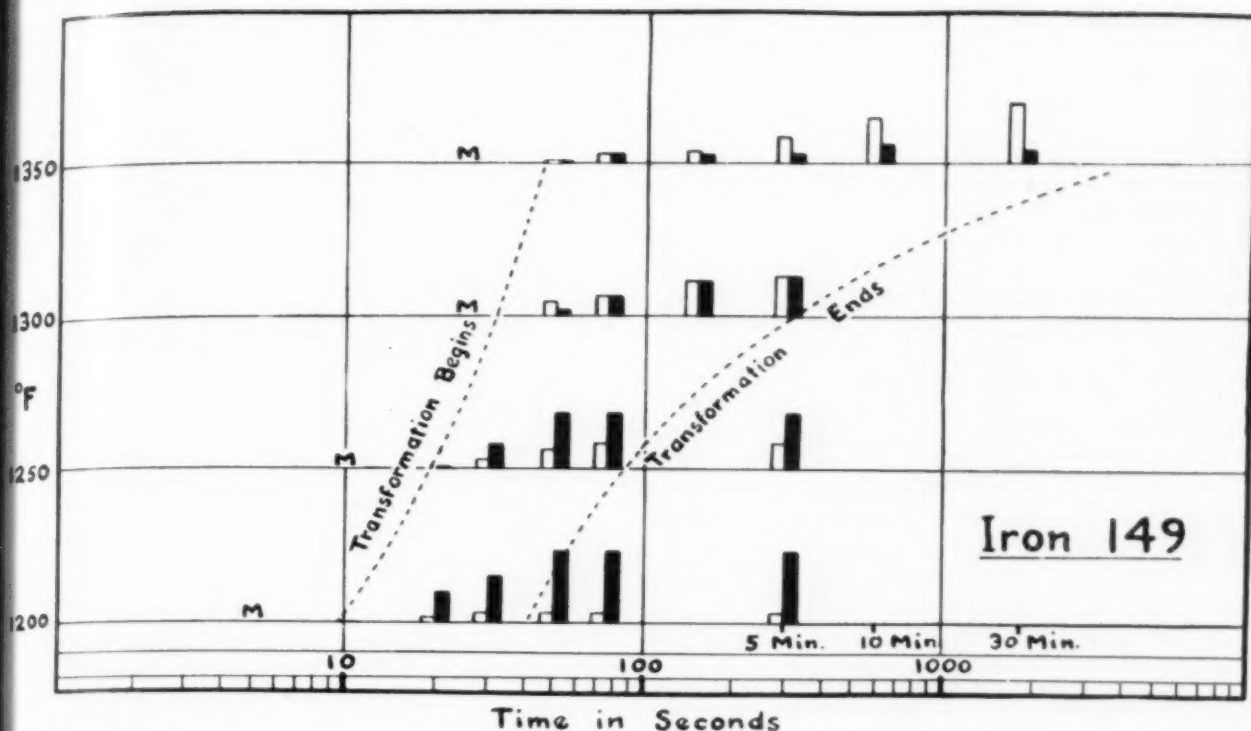


Fig. 6—Relative rates of stable and metastable transformation under conditions of isothermal treatment. The heights of the vertical bars show the relative amounts of free ferrite and pearlite. White bars represent ferrite—black bars pearlite.<sup>3</sup>

Analysis of Material, per cent

TC	Si	S	P	Mn
3.03	2.34	0.063	0.097	0.70

If the bar is cooled somewhat faster, as by blowing with a stream of compressed air, it will still show some free ferrite along the graphite flakes.

This means that a part of the austenite has transformed according to the stable system to form ferrite and graphite ( $\alpha + Gr$ ) while the remainder has transformed according to the metastable system to form pearlite ( $\alpha + Ca$ ). Since the stable transformation always involves a diffusion of carbon to the graphite flakes, it is natural that the ferrite layers should form along the surfaces of the flakes, i.e., in those parts where such carbon diffusion could occur most easily. Again it is emphasized that the ferrite layers are not a result of the graphitization of pearlite. Careful studies have shown that the ferrite forms by direct transformation, and that its formation begins before any pearlite starts to develop in the matrix.

If a thermocouple were imbedded in the bar, it would indicate that transformation did not take place at the temperatures shown in Figs. 3 and 4, but actually occurred at about 1325 F. The metal was, therefore, undercooled considerably below equilibrium conditions. The fact that the metastable transformation of austenite into pearlite is accelerated at temperatures below the equilibrium range is well known.

If the austenite in 0.80 per cent carbon steel is undercooled to a temperature of around 1200 F, for example, it will transform completely to fine pearlite in a matter of seconds. In cast iron the stable transformation of

austenite to ferrite and graphite is likewise accelerated by undercooling, but not to the same degree as the metastable transformation.

Figure 6 illustrates the relative rates of the two reactions in an ordinary cast iron of the composition given. Note that, at temperatures of 1350 and 1300 F, the stable transformation proceeds more rapidly than the metastable transformation. At temperatures of 1250 and 1200 F the reverse is the case, and the metastable transformation proceeds more rapidly than the stable transformation. These specimens were heated to 1600 F, to produce austenite containing about 0.80 per cent carbon in solid solution, and were then quenched into molten lead and held at constant temperature during transformation.

Such conditions are different from those of continuous cooling and favor the stable transformation because more time is available for the diffusion of carbon to the graphite flakes. Figure 6, therefore, indicates the "tendency" of the metal to follow the stable transformation at any given constant temperature level. What happens during continuous cooling is determined by this tendency, plus the amount of time available for carbon diffusion to the graphite flakes.

In the cast iron bar cooled in air from 1600 F, for example, it may be found that transformation occurs at 1325 F and requires about one minute to reach completion. What develops in the final structure is then determined by the following factors:

1. The tendency of the metal to follow the stable transformation under conditions of undercooling.
2. The time available for the diffusion of carbon to the graphite flakes, and the rate of such diffusion at the temperature in question.

These factors govern the amount of free ferrite which will be formed along the surfaces of the graphite flakes. The remainder of the austenite will transform according to the metastable system and produce pearlite.

Returning now to the structure of the as-cast bar, which was fully pearlitic, the question arises: If the matrix was fully pearlitic after cooling in the mold, why is not the pearlitic condition re-established by heating it above the transformation range and cooling again at the same rate that the casting cooled originally? This question has not been satisfactorily answered, but the following factors may be involved.

A casting cooling continuously in the mold is in a state of continual flux as regards the diffusion of carbon to the graphite flakes. Just after solidification the austenite holds in solid solution about 1.50 per cent carbon (determined in Fig. 4 by projecting to the left to intersect line *CE*, as previously described).

As this austenite cools and approaches the transformation range, it diffuses carbon to the graphite flakes, i.e., it tries to reach a condition of equilibrium at each successively lower temperature level. The casting, however, is cooling continuously and does not necessarily wait for the austenite to reach equilibrium throughout its entire mass. The first part of the austenite to reach eutectoid composition (0.70 per cent C) is undoubtedly that in direct contact with the surfaces of the graphite flakes.

Diffusion processes always involve a composition gradient, and in this case the lower carbon content in the austenite will be in the direction of diffusion, i.e., toward the flakes. Transformation, therefore, starts with the formation of nodules of pearlite at various points along the flakes.\*

At the time the pearlite nodules start growing, the austenite more remote from the flake surfaces still is above eutectoid composition and continues to diffuse carbon through whatever spaces are open between the growing nodules of pearlite. We must think of this mechanism as a dynamic condition in which the austenite is continually striving to reach a condition of equilibrium with falling temperature.

#### Composition and Cooling Rate

If the silicon content is correct for the cooling rate involved, the austenite will just reach a condition of metastable equilibrium at the flake surfaces and the entire matrix will transform into pearlite. If the silicon content is too high, some of the austenite at the flake surfaces will reach a condition of stable equilibrium and layers of ferrite will form along the flakes. Foundrymen have learned to balance composition against cooling rate so as to produce a fully pearlitic matrix in a given size of casting.

When a cast iron bar of balanced composition, which is fully pearlitic as-cast, is reheated again to produce austenite, the dynamic conditions mentioned in the foregoing are not necessarily re-established. Reheating to 1600 F, for example, produces a uniform homogeneous austenite holding about 0.80 per cent carbon in solid solution.

This homogeneous austenite may be quite different, as regards carbon distribution, from the austenite which existed in the original casting after it had cooled to 1600 F in the mold. When the homogeneous austenite in the reheated bar is cooled again at the same rate as the original casting, it does not necessarily transform in exactly the same manner. Conditions for

\*This process has been observed in actual castings while they were cooling in the mold.

carbon diffusion as this homogeneous austenite cools are more favorable to stable transformation, and some free ferrite is likely to form at the surfaces of the graphite flakes.

Reheating the bar to 1800 F will come nearer to the condition of the original casting, and it will be found that less free ferrite forms after cooling from 1800 F. A temperature near the melting point would probably be required to approach the actual conditions existing in the original cast iron bar as it cooled continuously in the mold.

#### Mixed Transformation Under Conditions of Isothermal Treatment

Mixed transformation, according to both stable and metastable states, may be a factor of some importance in the isothermal transformation of cast iron, i.e., cases where the metal is first heated to produce a fully austenitic matrix and then cooled suddenly to some subcritical temperature and allowed to transform at a constant temperature level. The rate at which the austenite transforms under these conditions varies considerably at different temperature levels.

Figure 7, taken from the work of Flinn, Cohen, and Chipman,<sup>4</sup> shows the time required for transformation to begin and end in a series of cast irons containing various amounts of alloying elements. Figure 8 shows the range of hardness values obtained after isothermal transformation at various temperatures. All these specimens were heated to 1600 F and held for 25 min prior to quenching. They were, therefore, fully austenitized and the austenite held in solid solution about 0.80 per cent carbon.

Taking the case of the plain cast iron (Curve 3 in Fig. 7), some of the structures obtained after isothermal treatment will now be described, using the data given by the authors. Specimens transformed at 1300 F contained free ferrite, indicating that some of the austenite had transformed according to the stable state while the remainder transformed into pearlite.

Specimens treated at 1250 F likewise showed free ferrite (compare with Fig. 6). At lower temperatures the amount of ferrite decreased and the pearlite lamellae become finer, causing the rapid rise in hardness shown on the right-hand side of Fig. 8. The fastest rate of transformation in Heat No. 3 occurred at 1200 F, and was complete in 20 sec. Below this temperature level, the rate of transformation slowed very decidedly. At 1000 F, for example, transformation required 1000 sec to reach completion (Fig. 7).

#### Acicular Type Structures

At around 900 F the lamellar transformation to fine pearlite no longer took place, and the structures changed over to the acicular type. A specimen of Heat No. 9, partially transformed at 900 F, exhibited an interesting condition which merits discussion in some detail. Transformation in this case started with the formation of needle-like particles of free ferrite, which the authors term acicular ferrite. From their explanation, this was just ordinary ferrite containing only traces of carbon.

A specimen of Heat No. 3, fully transformed at 800 F, was found to contain 90 per cent acicular ferrite, the remainder of the matrix being an intimate ferrite-carbide aggregate. Since the original austenite contained about 0.80 per cent carbon in solid solution, the

Fig. 7  
the eff  
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1300  
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question naturally arises: What became of the carbon which was rejected by the 90 per cent of acicular ferrite as it grew during transformation?

Dr. Chipman was of the opinion that this carbon was concentrated into the 10 per cent of remaining austenite, which was thereby greatly overloaded with carbon. When this greatly overloaded austenite finally transformed at 800 F, its burden of excess carbon would then be rejected as free carbide particles. X-ray diffraction studies, however, did not reveal any such excess of free carbide particles after transformation.

An alternative explanation is to consider that the excess carbon of the overloaded austenite simply diffused to the surfaces of the graphite flakes, which were near at hand and in direct contact with the austenite itself. It will be noted in Fig. 7 that the time required for the austenite to transform at 800 F was extremely long, amounting to over 10,000 sec (more than 2 hr). If the carbon had time to diffuse out of the acicular ferrite into the austenite, it would also have had time to diffuse out of this austenite and deposit on the surfaces of the graphite flakes.

This process of the diffusion of carbon to the graphite flakes with the attendant formation of free acicular ferrite is precisely equivalent to the mechanism of stable transformation previously described under the stable equilibrium diagram (Fig. 4). If such a mechanism

Fig. 7—Transformation curves for cast irons showing the effect of nickel and molybdenum on the times for the beginning and ending of austenitic transformation at constant subcritical temperatures. Base Analysis: 2.5 per cent total carbon, 2.5 per cent silicon, 0.85 per cent manganese, 0.10 per cent sulphur and phosphorus. Cooling curves for 1, 2, 4 and 6-in. diameter castings are shown with zero time taken at 1350 F. Flinn, Cohen, and Chipman.<sup>4</sup>

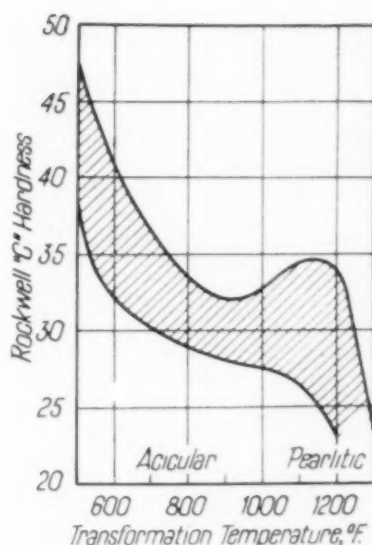
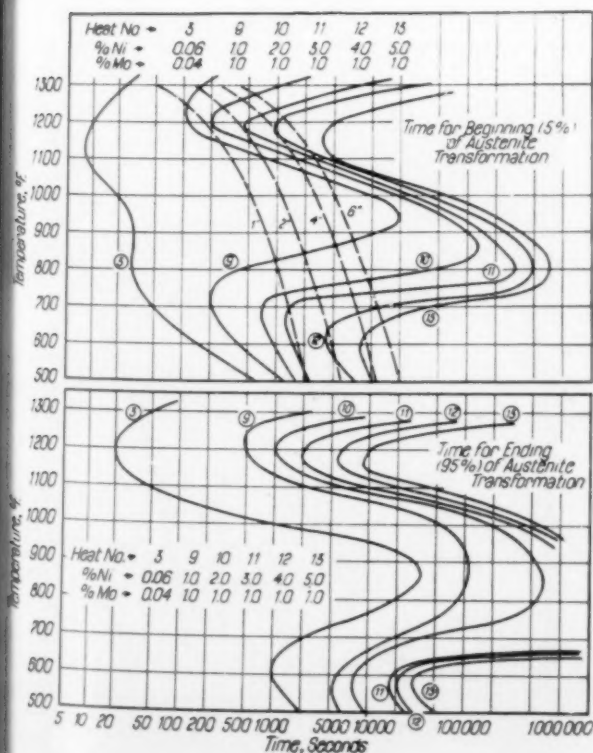


Fig. 8 — Range of hardness values found for thirteen nickel-molybdenum cast irons at the end of isothermal transformation. Flinn, Cohen, and Chipman.<sup>4</sup>

occurs isothermally at 800 F, it simply means that the metal is following a mixed transformation according to both the stable and metastable states.

Two temperature levels thus appear to favor the formation of free ferrite during isothermal transformation; one in the range from 1350 to 1200 F (see Fig. 9), in which free ferrite develops along the surfaces of the graphite flakes, and a second between 900 and 700 F in which free acicular ferrite is formed.

The formation of free acicular ferrite was responsible for the dip in the hardness values shown in the middle part of Fig. 8. At temperatures below 800 F, the amount of acicular ferrite decreased and the hardness of the matrix increased rapidly, as shown on the left-hand side of Fig. 8.

A whole series of acicular type structures was obtained at temperatures between 800 and 500 F, similar in every way to the structures obtained by the isothermal transformation of steel. This study was not carried to temperatures below 500 F, but it is known from other work that cast iron transforms into martensite at temperatures below 500 F in the same general manner as steel.

The time required for transformation to begin and end at any temperature level is easily determined by consulting Fig. 7, which illustrates the marked effect of molybdenum and nickel on the transformation behavior. The curves in Fig. 7 resemble those found for many alloy steels, and show that alloying elements influence the transformation of cast iron in a similar manner.

In general, the alloy additions serve to slow up the rate of transformation in the temperature range between 1300 and 1100 F, so that a casting may readily be cooled through this range without forming any pearlite. Having passed through the temperature range in which pearlite forms, the austenite will then transform at some lower temperature to produce a hardened matrix. It is thus possible to harden a large section all the way through without subjecting it to drastic quenching.

If sufficient alloys are added a heavy casting may be hardened all the way through by simply heating it to produce austenite and then cooling it in air. Acicular structures can also be obtained directly on casting by



suitable alloy conditions. Such castings may be considered as being heat treated in the mold.

Two other factors should be mentioned before closing the discussion. These are the effect of the graphite flake pattern and the effect of segregation. The presence of graphite flakes in cast iron is an open invitation for the metal to transform according to the stable state.

### Additional Factors Influencing Transformation

In steel the metastable transformation of austenite to pearlite ( $\gamma \rightarrow \alpha + Ca$ ) is said to be nucleated by carbide. The stable transformation ( $\gamma \rightarrow \alpha + Gr$ ) may be considered as being nucleated by graphite. The reaction therefore begins at points where a graphite surface is in direct contact with austenite, i.e., along the surfaces of the graphite flakes.

Stable transformation ( $\gamma \rightarrow \alpha + Gr$ ) does not produce a lamellar structure of ferrite and graphite, analogous to the structure of pearlite. The existence of the eutectoid point (C) in the stable diagram (Fig. 4) is no indication in itself that a lamellar structure of  $\alpha + Gr$  must form during transformation. Point C in Fig. 4 simply defines the limiting concentration of carbon remaining in solid solution in the austenite ( $\gamma$ ) after it has cooled to a temperature of 1550 F.

The anticipated lamellar graphite eutectoid structure does not develop when this austenite undergoes stable transformation. Metallurgists, to satisfy fixed habits of thought, say that the graphite eutectoid is divorced. It is simpler to say that there was no marriage to begin with.

In the metastable transformation ( $\gamma \rightarrow \alpha + Ca$ ), carbon diffuses to a lamination of carbide and layers of ferrite form adjacent to layers of carbide. This aggregate is called pearlite in steel. In the stable transformation ( $\gamma \rightarrow \alpha + Gr$ ) carbon diffuses to a graphite flake and a layer of ferrite forms adjacent to a layer of graphite. This aggregate is called primary ferrite in cast iron. The mechanism is fundamentally the same in both cases; the final aggregate is entirely different.

It would be expected that the amount of primary ferrite in the final aggregate would be influenced by the size, shape, and distribution of the graphite particles present in the austenitic matrix at the time it started to transform. The greater the graphite-austenite interface, the greater will be the opportunity for stable transformation.

For illustration purposes, imagine a 1-in. cube of iron-silicon alloy having at its center a single sphere of graphite  $\frac{1}{4}$ -in. in diameter, and assume that the cube has been heated and consists of homogeneous austenite having 0.70 per cent carbon in solid solution, ready to transform. During cooling in air, say that there was sufficient time for diffusion to produce a layer of ferrite around the sphere of graphite, this ferrite layer being 0.01 in. thick. The ferrite layer would develop by stable transformation ( $\gamma \rightarrow \alpha + Gr$ ).

The remainder of the cube would form pearlite by metastable transformation ( $\gamma \rightarrow \alpha + Ca$ ). If the cube were then sampled for analysis, by reducing the whole thing to chips and mixing thoroughly, the "combined carbon" result would show almost 0.70 per cent, or practically the same amount held in solid solution by the original austenite.

Suppose now that the same  $\frac{1}{4}$ -in. diameter sphere of graphite is reduced to many small spheres, and that

these are scattered through the same 1-in. cube of iron-silicon alloy so that the distance between the small spheres is 0.01 in. The cube is now heated and cooled exactly as before. A layer of ferrite 0.01 in. thick will again form around each small sphere of graphite.

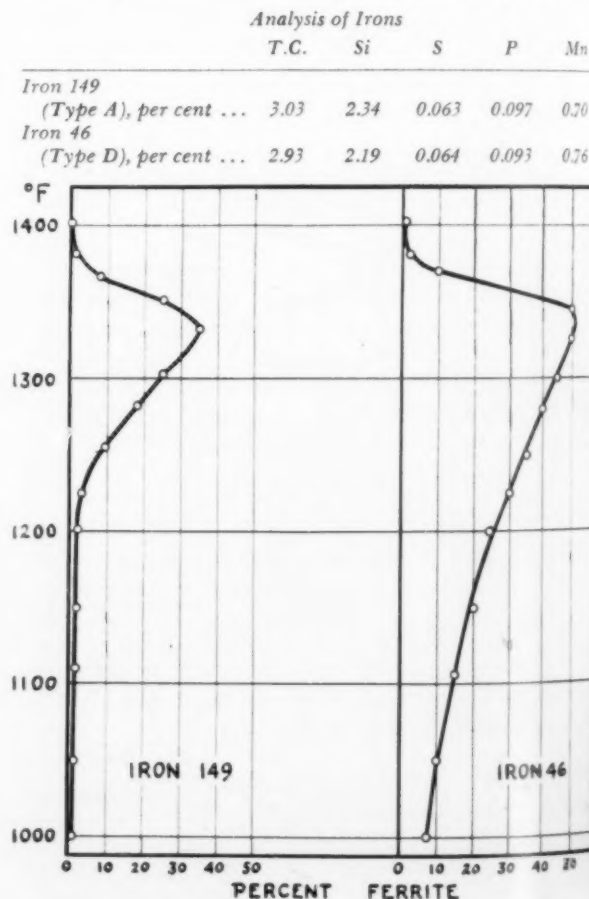
Since the distance between spheres is only 0.01 in., it is evident that the ferrite layers will merge completely during transformation to produce a fully ferritic matrix and no combined carbon would be found by analysis. Any intermediate amount of combined carbon could obviously be obtained by adjusting the size and spacing of the graphite spheres in the iron cube.

The effects of this mechanism are readily seen in malleable iron, which forms a "bull's eye" of ferrite around each nodule of graphite during transformation. Similar effects are found in the so-called graphitic steels, which contain substantial amounts of silicon. In gray iron, the ferrite forms as layers along the surfaces of the flakes, but the same principles apply.

Irons having type D graphite are liable to give much more primary ferrite than irons having type A graphite (see Figs. 9, 10, 11). In cases of mixed structure, where type D graphite occurs at the centers of the eutectic cell pattern, primary ferrite will form as patches in these locations.

For this reason, irons having type A graphite are

Fig. 9—Amount of free ferrite formed in 5 min under conditions of isothermal treatment. Specimens were austenitized at 1600. Iron 149 had type A graphite (see Fig. 10). Iron 46 had type D graphite (see Fig. 11). The small amount of ferrite formed at 1400 F simply means that transformation had not gotten underway in 5 min.<sup>3</sup>



usually preferred for heat treatment. It is quite easy, however, to quench both types of iron to produce a fully martensitic matrix without any trace of free ferrite. Rapid cooling does not allow sufficient time for stable transformation to occur, and the austenite undercools to temperatures below 500 F and then transforms into martensite. In heavy sections, where the cooling rate at the center is insufficient to produce a fully martensitic condition, the foundryman resorts to the use of alloys, following the principles mentioned in the section on isothermal treatment.

Segregation in steel has been found to influence its transformation behavior on heating and cooling. If a steel bar is banded, due to segregation in the original ingot from which it was rolled, the rate of transformation will be different in different portions of the bar, as a result of differences in chemical composition. Similar effects of segregation are found in gray cast iron.

The segregation pattern in cast iron is different from that of steel because the mechanism of freezing is different. In cast iron, the major segregation effect occurs during the freezing of the eutectic and takes the form of a network which has been termed the cell structure. The boundaries of these cells represent the last portion of the eutectic to freeze and, consequently, contain the highest concentration of segregated elements. Segregation of phosphorus into the cell boundaries is the most obvious thing noted in the structure. Molybdenum also tends to segregate along with the phosphorus.<sup>5</sup> In irons containing chromium, the chromium carbides tend to appear at the cell boundaries.

Effects of cell boundary segregation on transforma-

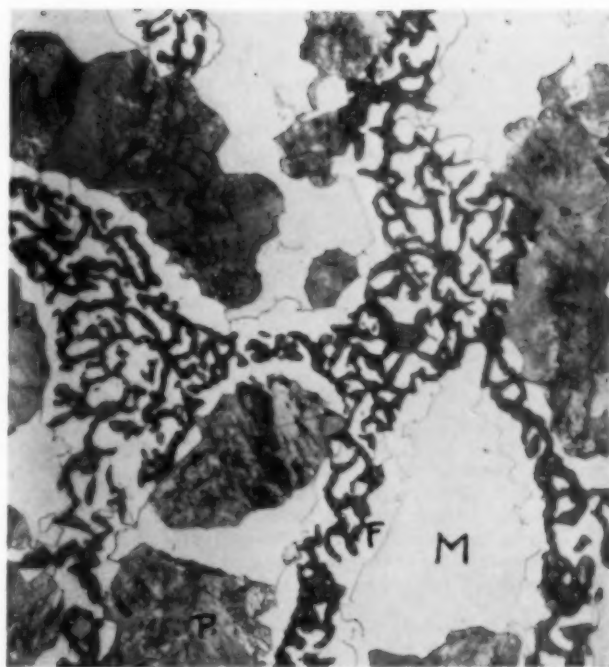


Fig. 11—Specimen of Iron 46 treated in exactly the same manner as that shown in Fig. 10. Note that the free ferrite is further developed than in Fig. 10. The small ferrite grains have consolidated into layers. There is some metastable transformation to pearlite, however, at points in direct contact with the surfaces of the graphite flakes. Etched.  $\times 750$ .

Fig. 10—Specimen of Iron 149, heated to 1600 F, quenched into a lead bath standing at 1250 F, held 30 sec. at 1250 F and finally quenched in water. "M" represents martensite, "F" is free ferrite, "P" is pearlite. Both stable and metastable transformations are going on at the same time. Free ferrite forms as small grains along surfaces of the graphite flakes. Etched.  $\times 750$ .



tion behavior have been noted by several investigators.<sup>6, 7, 8</sup> With moderately slow rates of cooling, the cell boundaries tend to follow the metastable transformation ( $\gamma \rightarrow \alpha + Ca$ ), while the stable transformation ( $\gamma \rightarrow \alpha + Gr$ ) is favored at the centers of the cells. Isothermal transformation in the temperature range from 1350 to 1200 F produces similar effects.<sup>8</sup>

At lower temperatures of isothermal treatment, the transformation is found to be more sluggish in the cell boundaries than elsewhere.<sup>6</sup> Under conditions of extremely fast heating and cooling, as by induction hardening, some untransformed austenite may be retained at the cell boundaries.<sup>8</sup>

All of these effects are natural consequences of differences in chemical composition at different points in the matrix. The cell structure must not be confused with the austenite grain size obtained on reheating above the transformation range. The cell structure is a segregation pattern formed during freezing and cannot be altered by recrystallization in the solid state. Control of the pattern goes back to the original casting and is carried out by inoculation of the molten iron before pouring.

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## DISCUSSION

Chairman: J. S. VANICK, International Nickel Co., New York

Co-Chairman: T. E. EAGAN, Cooper-Bessemer Corp., Grove City, Pa.

J. E. REHDER:<sup>1</sup> Why should there be a break in the curve in Fig. 5 at 1500 F instead of a smooth climb?

MR. BOYLES: The sudden break at 1500 F, I think, is more a matter of plotting than anything else. If you had enough observations it would curve just like it does at the bottom.

CHAIRMAN VANICK: One of the most fascinating features derived from the presentation Mr. Boyles made is in comparing the difference between the steel diagrams and the iron diagrams in the 1500-deg band, where in the stable system cast iron goes completely to graphite and ferrite. In the steel system it will go to globular pearlite and ferrite. This distinction where the structures do go to ferrite and graphite, if they have time to complete that action, may become an important procedure in processing cast irons at some future time. It should be possible to stop cooling at that level and wait for the reaction to move along in the stable state to make a new product.

MR. BOYLES: There is a possibility of that from a time standpoint. I had never thought of it in just that light, but it might have possibilities.

CHAIRMAN VANICK: It is similar to malleable annealing.

J. H. SCHAUUM:<sup>2</sup> Does Mr. Boyles have data on rates of transformation of the unstable alpha plus pearlite system to the stable alpha plus graphite system in the area below A-B? It must be a function of time and temperature in that zone.

MR. BOYLES: I do not have quantitative data on that. It does go much more rapidly as you approach that line and much slower at lower temperatures.

W. J. DIEDERICKS:<sup>3</sup> Some years ago there was considerable

work done on white cast iron at the University of Michigan and elsewhere. I believe if you have some of the stable system present, in other words, if you have the structure in the form of pearlite with stable graphite already present at around 1200 or 1300 F it takes about 20 hr to complete the break-down of the pearlite. I am quite sure of my memory as to the 20 hr, but I am not so sure as to how long it will take to break down pearlite if some of the graphite is not present also. I would expect that to take much longer.

R. SCHNEIDEWIND:<sup>4</sup> The paper by S. C. Massari, "Heat Treatment—Its Effect on the Combined Carbon and Physical Properties of Wheel Iron," *TRANSACTIONS, A.F.S.*, vol. 47, pp. 1-13 (1939) gives a great deal of data regarding decomposition of cementite. It may be seen from his data that the time for disappearance of combined carbon is influenced by the temperature of annealing. The decomposition of carbides is slow below the critical temperature, say 900 or 1000 F, but if a temperature is used which is near the critical (Line A-B, Figs. 3 or 4) the action is quite rapid.

MR. SCHAUUM: This discussion emphasizes the point that in describing these diagrams it is important to mention that the pearlite to graphite transition is a time-temperature function.

MR. BOYLES: That rate of transition is conditioned by the silicon content. In the case of malleable iron the silicon is quite low. If you had 2½ per cent silicon you would find a much higher rate at any given temperature.

These diagrams are not binary diagrams. They are sections through a ternary system and they are more complicated than I indicated. I tried to simplify the principles as much as possible.

MR. DIEDERICKS: I agree with Mr. Boyles. That is the point that I was about to make. The figure of 20 hr is for complete break-down of the pearlite, and as the silicon increases the rate of break-down would be considerably faster.

R. W. LINDSAY:<sup>5</sup> There is another factor that could be introduced into this discussion of break-down of the combined carbon and that is the type of graphite that is present. I have noticed that when we have dendritic graphite structures the combined carbon break-down takes place quite rapidly whereas with normal graphite our break-down is slower even with the same analysis.

MR. REHDER: Relative to the rate at which pearlitic carbide decomposes, Dr. Schneidewind indicated some time ago that the time-temperature relationship is a semilogarithmic one. In other words, as the temperature increases the time decreases in a logarithmic fashion, so it is not proportional as time-temperature goes.

<sup>1</sup> University of Michigan, Ann Arbor, Mich.

<sup>2</sup> Pennsylvania State College, State College, Pa.

<sup>3</sup> Bureau of Mines, Ottawa, Canada.

<sup>4</sup> National Bureau of Standards, Washington, D.C.

<sup>5</sup> The Autocar Co., Ardmore, Pa.



# GRINDING STANDARDS HELP ELIMINATE CLEANING ROOM BOTTLENECKS

By

Dean Van Order \*

AN EVEN FLOW OF CASTINGS through the cleaning room has long been recognized by foundrymen as an important factor in establishing sound and economical casting costs. During the war, when many plants were being pushed beyond capacity, the bottleneck tendencies of the cleaning room were more than ever apparent. Not only did the cleaning room have to correct errors of the foundry, but also had to devise new ways of handling this increased production at a minimum cost. Foundries which previously had no incentive plans in the cleaning room found it necessary to install them; foundries with established plans had to bring them up to date if they intended to handle this increased burden.

Progressive foundries desiring to meet present-day competition soon realized that they had to maintain a better control on cleaning room costs. Grinding, being one of the larger operations in the cleaning cycle and also one of the easier cleaning tasks on which to establish piece rates, was quickly selected by time and methods engineers for standardization and methods improvement. The value of a good incentive plan on stationary or swing grinding can be readily recognized by cost-minded supervision. The plan presented in the paper is in operation at the present time; it is one that is simple to maintain and readily adapts itself to any method changes.

## Management and Supervision Must Cooperate

Too often, time study engineers make the mistake of not laying the groundwork for installing an incentive plan before they actually take studies. Management and supervision must be thoroughly sold on any plan; they must be presented with at least a rough estimate of the cost of installation and the benefits and savings to be derived. Management and supervision must be completely in accord with the plan.

It is definitely the problem of the time study engineer to sell his ideas to management and supervision. The time study department should be the source of all costs and, therefore, an essential part of the business organization. A department as important as this cannot be turned over to just any individual. Almost any man with an average education and common sense will

produce accurate time studies, but an ability of far greater magnitude is needed to sell and keep sold the mechanics of time study and methods work. The human problems encountered by the time study man must be worked out with the help of management and supervision.

The time study department is strictly a service to supervision. Foremen must feel free to ask the time study man to assist them in their production problems. Quite often supervision is kept in the dark as to time study and methods procedures, resulting in poor co-operation. It is the job of the time study engineer to convince the foreman that it is to his advantage to report all delay time, method changes, and any irregularities in the production schedule of his department.

Since the cleaning room contains the most varied operations in the modern foundry, it is essential that the time study engineer have the complete confidence of its foreman. In return he must recognize all the problems encountered in the cleaning room, realizing that they will not be corrected overnight.

## Improve Methods

Before any new incentive plan may be installed, or the old one revised, present methods should be thoroughly scrutinized. What can be done to improve present grinding operations? The following factors should be investigated:

1. Sequence in which castings flow through the cleaning room, with particular attention to backtracking.
2. Possibility of improving the pattern match between cope and drag in order that the parting line or amount of fin to be ground may be decreased or even eliminated.
3. Improvement of core boxes and core prints to eliminate some of the potential core fin.
4. Excessive use of nails and vents. These are costly items in snag grinding, the ratio of grinding wheel cost to the amount of metal removed being quite high. On heavy work, removal of nails and vents by chipping may reduce costs.
5. Use of flogging or necked down cores to reduce the contact and grinding area.
6. Changing the location of heads or gates to make them more accessible to the large grinding wheels. Metal can be removed at much lower cost on a wheel of large surface area than on a wheel of smaller area.

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7. Use of the proper grade of wheel for the average type of castings ground. The grit size, hardness, and bond of the grinding wheel have a direct bearing upon the speed with which the metal can be removed and the life of the wheel itself.

8. Pressure bars may be used to advantage on stationary grinding.

9. Methods of holding castings rigid for swing grinding may be improved. Vibration will tend to break down grain structure in the grinding wheel and also retard its cutting action.

10. Castings should be properly burned with only a minimum of metal left for grinding.

11. Grinding operators should be properly instructed in the best grinding techniques.

No doubt, alert foremen and time study engineers will be able to add other method improvements to this check list. Methods which are improved before studies are taken will reduce the future work of the time study men. Method changes made after standards have been set necessitate a change in standard tables, standard data and a selling job done on the employee; however, any method changes that have to be made after standards have been established should be corrected immediately. If neglected, they will result in either poor employee relations or loose standards.

#### Adequate Time Studies

Various sizes, shapes and weights of castings produced in a miscellaneous jobbing foundry necessitate that a sufficient number of studies be taken on grinding operations. In the opinion of the author, the number of time studies taken should vary somewhere between 50 and 100 on each type of grinding (swing or stationary), depending on the variation of work.

The time study engineer must use his own judgment on the type of castings to study; however, he should try to cover: 1. A variation in weight; 2. numerous head and gate sizes; 3. various locations of heads and gates; 4. amount and location of fin; 5. number and location of nails and vents.

Since most present-day foundries have had to produce numerous types of metal to meet competition, another problem confronts the time study man. Allowances must be made over and above the regular types of steel produced to take care of the additional grinding time needed on certain high-alloy and high-carbon metals. The time study engineer must collect sufficient studies on this type of work to enable him to determine these additional allowances. If the standard table is based on studies made on the simpler and more common types of metal, then one or two additional percentage allowances probably will take care of the remaining types of metal.

The time study man must know something about grinding operations before he can take any studies; he must be familiar with the types of metal, and the inspection requirement on each. Grinding is one operation where the engineer may be misled on pace rating. In determining whether the operator is using effort to remove metal, he must watch the amount and color of the sparks. An ammeter can be installed on each grinding wheel to help determine how much pressure is being exerted by the operator.

An analysis of the elements in the grinding operations should be made before taking studies. If the

variable and constant elements and their unit of measure are recognized beforehand, the time study observer will be able to do a much better job of breaking up the elements on the time study sheet. The selection of the proper unit of measure for each element is the secret of establishing a good standard table. The time for grinding heads will vary with the size or area of the head, not the weight of the casting; likewise, handling the casting will vary with the size or weight of the casting and not the area of the heads.

At the present time, three swing frame grinders located side by side are in use in this plant. All machines are equipped with a fully enclosed 15 hp motor. The belt is a double "V" of sufficient length to reach from wheel pulley to motor pulley. The stands on which work is placed are built up of welded sections of flask filled with molding sand. This type of bench is ideal for swing grinding as it absorbs any vibration built up by the action of the wheel on the casting.

Six stand grinders are used for grinding smaller castings. All machines are 15-hp single motor with double grinding wheels; belt drive is used with four "V" belts on each wheel.

Extensive tests were made on grinding wheels to determine the best grade of wheel for the average type of castings ground. The wheel selected was the one giving the best life with the greatest amount of metal removed in the shortest time. Wheel dimensions were 24-in. outside diameter and 3-in. thickness. Tests were conducted on grinding wheels by weighing both the wheels and the casting to be ground before and after grinding to determine the amount of metal removed for each pound of wheel loss. Ammeters were installed on each grinding machine that was used for the test to determine if the operators were applying a uniform amount of pressure.

Since the wheels selected were high speed and used both on the stationary and swing grinders, a surface speed of approximately 9,250 fpm was recommended on all grinding machines. New wheels are placed on the swing grinder and used down to 18 in., from whence they go to certain stand grinding machines where they can be used down to the safety rim. By changing the wheels from machine to machine as they are reduced in diameter, it is possible to maintain approximately the same surface speed at all times by increasing the spindle speeds on the various grinding machines.

#### Establish Standard Data

All studies were taken by the snap-back method, two time study watches being used; one watch clocked the time for each element, and the other clocked the overall time for the entire cycle. The snap-back method was used because of the smaller amount of clerical work entailed in figuring studies, and the ease with which the time study man could follow the operator if he changed the sequence of the elements during the study, as will occasionally happen when timing miscellaneous operations.

Complaints arising from the loss of a certain amount of time or leaving out elements or delays when using the snap-back method was eliminated by the use of the second watch to check the sum of the readings for each element in every complete cycle. The overall time for each cycle must agree within a few one-hundredths of a minute with the sum of the readings for all the ele-

- TIME STUDY SHEET -										Pattern No. Y-567				
Customer - <u>John Doe</u>	Dept.	Floor	Clock No.	Machine No.	Type Mch.	Fixture No.	Material	Part No.						
Study No. <u># 6</u>														
Operation No. <u>Grinding</u>	<u>G</u>		<u>#20</u>	<u># 2</u>	<u>Stand</u>		<u>#1 Steel</u>	<u>Part Name Front Wheel</u>						
Observer <u>W. L.</u>							<u>135-160 Br.</u>	<u>Dwg. No. Support Adapter</u>						
Date <u>6-5-40</u> Elapsed Time <u>10:01</u> To <u>10:15</u> Speed and Feed O.K. -														
Description of Operation - <u>Grind all fin and all heads &amp; gates except one, on 3" wheel - Stationary grinder</u>										STANDARD <u>1.91 Minutes</u>				
Elements of Operation	1	2	3	4	5	6	7	8	9	10	Total	Actual	Code Relax.	Stand.
<u>1 - Complete</u>	<u>1.22</u>	<u>1.35</u>	<u>1.50</u>	<u>1.56</u>	<u>1.69</u>	<u>1.15</u>	<u>1.44</u>	<u>1.40</u>	<u>1.63</u>	<u>1.64</u>				
<u>Pick up &amp; position</u> <u>23.5" wt</u>	<u>.12</u>	<u>.08</u>	<u>.11</u>	<u>.08</u>	<u>.12</u>	<u>.11</u>	<u>.09</u>	<u>.10</u>	<u>.09</u>	<u>.11</u>	<u>1.01</u>	<u>.101</u>	<u>65</u>	<u>.13</u>
<u>Grind Fin</u> <u>* total of 41"</u>	<u>.10</u>	<u>.11</u>	<u>.12</u>	<u>.16</u>	<u>.11</u>	<u>.09</u>	<u>.12</u>	<u>.11</u>	<u>.17</u>	<u>.10</u>	<u>1.19</u>	<u>.119</u>	<u>70</u>	<u>.17</u>
<u>Grind Gate</u> <u>2 x 1/2 regular</u>	<u>.20</u>	<u>.17</u>	<u>.23</u>	<u>.18</u>	<u>.32</u>	<u>.19</u>	<u>.18</u>	<u>.22</u>	<u>.19</u>	<u>.31</u>	<u>2.19</u>	<u>.219</u>	<u>50</u>	<u>.23</u>
<u>Grind Fin</u> <u>* see note above</u>	<u>.09</u>	<u>.11</u>	<u>.11</u>	<u>.16</u>	<u>.11</u>	<u>.08</u>	<u>.12</u>	<u>.10</u>	<u>.17</u>	<u>.10</u>	<u>1.15</u>	<u>.115</u>	<u>70</u>	<u>.17</u>
<u>Grind Gate</u> <u>1 3/4 x 1/2 regular</u>	<u>.13</u>	<u>.11</u>	<u>.17</u>	<u>.12</u>	<u>.16</u>	<u>.12</u>	<u>.12</u>	<u>.16</u>	<u>.13</u>	<u>.15</u>	<u>1.37</u>	<u>.137</u>	<u>75</u>	<u>.21</u>
<u>Grind Head</u> <u>2 x 1 3/4 - (1/2 irregular)</u>	<u>.51</u>	<u>.71</u>	<u>.68</u>	<u>.81</u>	<u>.82</u>	<u>.50</u>	<u>.72</u>	<u>.67</u>	<u>.82</u>	<u>.81</u>	<u>7.05</u>	<u>.705</u>	<u>65</u>	<u>.96</u>
<u>Casting aside</u> <u>23.5" wt</u>	<u>.03</u>	<u>.03</u>	<u>.03</u>	<u>.02</u>	<u>.03</u>	<u>.02</u>	<u>.04</u>	<u>.02</u>	<u>.03</u>	<u>.02</u>	<u>.27</u>	<u>.027</u>	<u>75</u>	<u>.04</u>
	<u>1.18</u>	<u>1.32</u>	<u>1.45</u>	<u>1.53</u>	<u>1.67</u>	<u>1.11</u>	<u>1.39</u>	<u>1.38</u>	<u>1.60</u>	<u>1.60</u>	<u>14.23</u>	<u>1.423</u>		<u>1.91</u>

Notice - Fin condition very good.  
2 x 1 3/4 head must be ground to the contour of the casting.

Fig. 1—Time study sheet for stand grinding on 3-in. wheel.

ments in that cycle (Fig. 1). If they do not agree, the time study man knows he has left an element out or missed some delay time.

The studies are broken down into elements, and each element is pace rated and quoted for fatigue independently. By pace rating each element, the time study man will derive a much more uniform curve when he plots the elements upon graph paper. Pace

rating is expressed in terms of 60 units per hour being normal; this would correspond to 100 per cent as normal under a percentage system of rating.

Fatigue or rest allowances vary from 15 per cent for light castings handled on the stand grinder, to 40 per cent for heavy castings handled on the swing grinder. All castings which had to be handled by hoist were given a flat 20 per cent fatigue allowance for handling.



Study number	#1	#2	#3	#4	#5	#6
Part. number	M 2617	28672	16384	N-1	386	Y-567
Customer						John Doe
Operator No.	#20	#16	#15	#16	#20	#20
Date taken	4-10-40	4-12-40	5-18-40	5-25-40	5-26-40	6-5-40
Casting Weight	6.5	22.	9.	3.1	16.	23.5
Grind Hds or Gate - Size	3X1 - .97	4X1 - .72	3X2 - 1.33	1X1 - .23	3X1 $\frac{3}{4}$ - 1.24	2X $\frac{1}{2}$ - .23
" " " " "	2X $\frac{3}{4}$ - .43	5X3 - 2.29	2X1 $\frac{3}{4}$ - .45		2X $\frac{1}{2}$ - .21	1 $\frac{1}{4}$ X $\frac{1}{2}$ - .21
" " " " "			1"D - .30			2X1 $\frac{3}{4}$ - .96
" " " " "						$\frac{1}{2}$ " irregular
Grind Fin - Regular - Lgth.	40" - .31	62" - .51	17" - .14	12" - .22	27" - .22	41" - .34
Grind Fin - Irregular - Lgth.			14" - .23			
Grind Vents - Size - No.	3/8"-4 - .31			1/8X1/8 - 08		
Grind Nails - Size - No.		3/8"-4 - .40			3/8"-2 - .15	
Handle - (Pick up & away)	.12	.16	.13	.07	.12	.19
Total	2.14	4.06	2.58	.60	1.94	1.91

Fig. 2—Comparison sheet for stand grinding on 3-in. wheel, showing levelled times for each element.

Actual grinding was allowed 25 per cent fatigue on both stand and swing grinding. All pace rating and fatigue allowances are in increments of 5 per cent; that is, 15, 20, 25, 30, 35 per cent, etc. An attempt by a time study engineer to break down allowances any finer than this is purely a guess and will entail more clerical work.

The time study sheet must be a record of all conditions prevailing at the time the study was taken. This is very important, as the time study man must refer back to the time study sheet when recording all of his accumulated data on a comparison sheet. The important items that must be listed (Fig. 1) can be summed up as follows:

1. Date the study was taken with the name or initials of the observer.
2. Pattern number and customer for whom the casting studied is being made.
3. Number of the operator studied and overall time of the study.
4. Type of machine and size of grinding wheel upon which the casting was ground.
5. Type of metal and Brinell hardness at the time the casting was ground.
6. Brief description of the operation.
7. Part number or part name if one is supplied by the customer.
8. Brief description of each element breakdown with area of head or gate ground, length of fin, number and size of nail buttons and weight of casting.
9. Any unusual condition about the casting that might affect the time for grinding should be recorded; such as heavy fins, heads burned too high or too low,

and the irregular location of head or gate which might require excessive grinding operations for removal.

10. A rough sketch of the casting with the location of the heads and gates is helpful to the time study man when he has to refer back to the study. The making of a sketch may depend, however, upon the time and ability that the time study engineer has at his disposal.

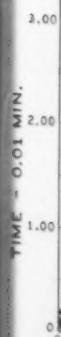
After a sufficient number of studies have been taken and computed for pace rating and fatigue, the levelled standard for each element is recorded on a comparison sheet. A comparison sheet is nothing more than a large sheet or sheets upon which all the levelled times for every element in each study are recorded side by side so that they can be compared (Fig. 2).

Each element is listed against the unit of measure selected for it; the standard for removing the heads or gates is listed against the area of the head or gate, the standard for removing fin against the length of fin, the standard for removing nails or vents against the number and size of nails or vents, and the standard for handling the casting against the weight of the casting.

After all elements have been recorded on the comparison sheet, the constant elements (those which do not vary with any unit of measure) can be totalled across the sheet and the average used as standard. The variable elements such as, grinding heads or gates on stand grinding (Fig. 3), are plotted on a graph and a curve drawn through the average of the points. The points on this curve make up the standard table for this particular element.

It is a good practice for the time study man after he has taken 15 or 20 studies to record them on the comparison sheet and plot the variable elements on a graph. By doing this, before too many studies have been taken, he can see certain elements or the range within elements where more studies should be taken, and avoid the waste of collecting unnecessary studies.

Fig. 3—3-in.



As previously mentioned, stand and swing grinding are among the simpler operations in the cleaning room upon which to establish standards. In both stand grinding and swing grinding, the area of the head or gate in square inches determines the standard for grinding. The time for removing metal on the swing grinder will be greater than that on the stand grinder as the heads generally are larger and the wheel must be moved to the work, which takes longer than moving the work to the grinding wheel. The standard for handling castings varies directly with the weight on both stand and swing grinding.

Castings ground on the swing grinder are paid an additional value, called resetting, for handling (Fig. 4).

This value is paid each time the swing grinder must set his machine aside and move the casting in order to reach other grinding areas with his wheel. Resetting time is based on the weight of the casting, but does not take as long as handling which covers picking up the casting, placing on the grinding block, and finally setting the casting aside. Nails, vents and fins are not ground with the swing grinder as they can be removed by the chippers at lower cost.

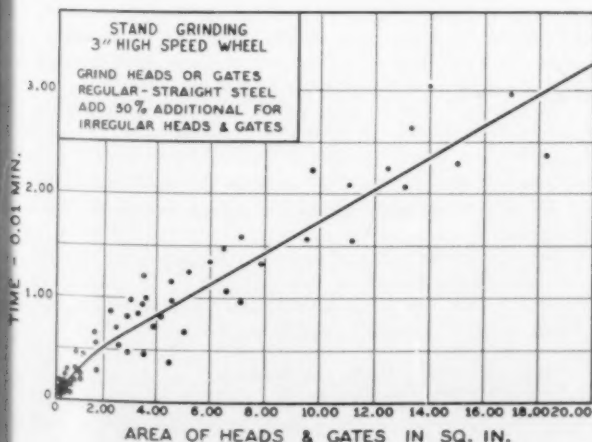
After the curves have been drawn for each variable element, the time for that element may be expressed by the formula for the curve or recorded in tabular form.

The use of a table to set standards is preferred as it is quicker and more accurate than applying a formula, in the opinion of the author. When deriving the tables from the curves, thought should be given to keeping them as simple as possible.

Elements which are classed by identical units of measure can be combined in the same table. Every effort should be made to eliminate any necessity on the part of the rate setter to interpolate values. Since standards are only average values, the bracketing of standards over a certain range will not change this average; for example, the removal of any amount of fin between 60 and 70 in. would carry the value for 65 in. of fin.

All information necessary to determine stand or swing grinding standards is recorded on a job specification sheet (Fig. 4). This information is recorded right from the pattern so that the piece rate can be figured before the casting is ground. The determina-

Fig. 3—Standard curve for grinding heads or gates on 3-in. width wheel on high speed stand grinder.



SWING GRINDING			
Size of Cuts	Reg. —	No.	Std.
5 X 5	Reg. —	4	15.48
5 X 1 3/8	Reg. —	2	3.16
	Reg. — Irreg.		
	Reg. — Irreg.		
	Reg. — Irreg.		
	Reg. — Irreg.		
	Reg. — Irreg.		
	Reg. — Irreg.		
	Reg. — Irreg.		
	Reg. — Irreg.		
	Reg. — Irreg.		
	Reg. — Irreg.		
Reset		4	7.96
Handling Wt 4 1/4 #			1.99
Gauge			
Surface			
Total			28.59

STAND GRINDING			
Size of Head or Gate	No.	Std.	
4 X 2 3/4	1	2.03	
1 1/4 X 3/8	2	.82	
2 X 2 (irregular)	1	1.25	
Size — Vents		No.	
3/4 X 3/8		1	.09
Parting		Length	
Reg.		4 1/2	.31
Irreg.			
Nails — No. (touch up) 3/8" — 3			
Miscellaneous			
Wt. 17.5 #			
Handle			
Total			4.88

Fig. 4—Work sheets listing all items ground by stand or swing grinders.

tion of irregular heads, gates or fin is left to the judgment of the time study man or rate setter. Standards for items not covered by the table, such as gauging or grinding special surface areas, are determined by check studies unless there is sufficient quantity upon which to build a table. The specification sheet should be kept as simple and neat as possible.

#### Good Checking Control

After the standard table has been compiled, a pre-check period of at least 2 weeks should be taken with the new standards. During this period, the time study man can check the table against actual performance to determine if any corrections are to be made before the standards go into effect. Also, while this precheck period is in effect, the time study man can instruct his checkers, or whoever will be responsible for keeping track of the operator's performance, on proper checking methods.

Checking may sound like a simple duty, and too often will be passed over lightly as far as instruction is concerned. Many time study engineers have had a good standard table appear tight or loose due to incompetent checking. Standards are only as accurate as their checking and maintenance of methods improvement. The employee selected to do the checking must be intelligent and honest; he must count accurately and report method changes to the standards department.

A savings in maintaining the piecework system can be made by combining the checking for piecework and inspection under one person. The employee doing the checking should have the authority to reject castings if the operator has not satisfactorily completed all of the elements included in his standard. The operator would not receive credit on his work sheet for castings ground until they met inspection requirements. Occasionally, certain irregularities will arise that are not covered by the standard table. The checker must record accurately the time and number of pieces ground under these conditions in order that the operator may be properly compensated.

In the opinion of the author, any foundry with a sound, intelligent organization can establish its own standards department and install accurate standards in the cleaning department, as well as in all other departments of the plant.

## DISCUSSION

Chairman: R. J. FISHER, The Falk Corp., Milwaukee.

Co-Chairman: M. E. ANNICH, American Brake Shoe Co., Mahwah, N.J.

A. E. VISK<sup>1</sup>: Does anyone know of a practical way of taking a photo of a casting rather than developing a blue print from a sketch, which I think would be extremely helpful particularly in predetermining costs or rate-setting? I have been thinking of taking a picture and transferring it onto a data sheet, but I have not found the answer.

MR. VAN ORDER: We thought of that idea also.

MEMBER: We have been grinding castings from standards for probably 10 years. Previous to our grinding operation we were sorting castings by pattern number. Is it possible to set up grinding operations on a tonnage basis rather than on a per-piece basis?

MR. VAN ORDER: If you have a miscellaneous shop you cannot do it successfully. Perhaps in a production shop you can where the heads and gates are uniform. Then I think everything could be ground on a pound or tonnage basis, but in a miscellaneous shop I do not think you can do it.

MEMBER: When you classify those castings you tend to put them in a particular group. Do you try to take each casting and figure out the operation as you have the size of the gate and so forth. Do you try to break those down in a particular group and have so many standards?

MR. VAN ORDER: We go through that same procedure on every casting, 1 piece to 1000, whatever it is. It does not take long. The clerk or rate setter sets the casting specification, measuring the size of the head even for one piece. We have full coverage on everything that goes through, whether it is production work or miscellaneous. We do that for every order, every pattern number.

C. O. HIMMELMAN<sup>2</sup>: How do you pay your men during the time you are taking your standards and doing your checking work to be sure that they are putting forth an equitable effort?

MR. VAN ORDER: If the man had been on day work previously, he would receive his base rate for grinding. If you were changing your table and had some old piecework method in and you wanted to revise it you would probably have to guarantee the man the old standards while retaking your standards. In some cases we have guaranteed men what we call a 75-unit hour, 125 per cent. We still do that when some conditions come up that we cannot measure. Otherwise they would receive daywork if they had never been on piecework.

C. J. PRUETT<sup>3</sup>: How does the author arrive at the correct percentage for fatigue on the various elements for handling grinding, etc.?

MR. VAN ORDER: There is no set rule for fatigue allowance. Our fatigue allowances vary from 5 to 50 per cent throughout our shop. We have a small machine shop which is included. Parts on which we are machining long cuts, we allow 5 per cent fatigue. The fatigue allowance on loading a bag of material into a car may be as high as 50 per cent. We set up schedules for that. Occasionally we have even added over 50 per cent. When you start doing that, however, you should look for some other method to improve this condition. You are asking a lot of a man where you have to add more than 50 per cent factor for relaxation and fatigue. I cannot give you any rule for setting those values. The important thing is to be uniform. If you use, say 20 per cent for handling with a hoist in one department, use it throughout your shop. We did not attempt to evaluate or set fatigue allowance in less than 5 per cent increments. There are some plants where they use 1 per cent and 1/2 per cent increments. I do not think you can be that precise.

CHAIRMAN FISHER: A few years ago an article in a university publication told of results of a study by their medical department on the subject of fatigue. They found that fatigue seemed to vary with certain chemicals produced in the kidneys and liver and secreted from the body. These chemicals increased with the amount of fatigue. Until they develop that further we will have to work up our own tables. Possibly some day we will have a more scientific method of determining what fatigue is.

MR. VISK: I think there are some well developed rest, delay

and fatigue curves in a rather old book, Merrick's *Timestudy and Rate Setting*.

CHAIRMAN FISHER: It might be a good basis to start on and adjust upward or downward according to your requirements.

MEMBER: If you pay a swing grinder who requires no training for the job, \$14 for an 8-hr day and a machinist who had to come up via an apprentice course the same pay, how are you going to prevent discontentment on the part of the machinist?

CHAIRMAN FISHER: Generally in any job evaluation plan you have three major items, skill, responsibility, and working conditions to consider. The swing grinder would rate rather low in the skill but probably high in working conditions, while the machinist would rate high in the skill and probably very low in working conditions. The two would balance fairly close in total number of points.

L. H. NABERHAUS<sup>4</sup>: I think the \$14.00 you are comparing here is an overall rate, base rate plus bonus. I think in job evaluation you would use the standards of comparison without bonus. I do not believe you are making the proper comparison for your analysis.

MR. VAN ORDER: We have a job evaluation plan in which we evaluated everybody in the plant, including the machine shops. On that basis it was rather amazing to discover how much lower the machine shop men were than what they had been getting and what the rate was for strict machine shops in our locality.

MEMBER: Once a job has been timestudied how do you go about retiming it? In our particular case our castings are not cleaned for two or three days after they are made, and sometimes we may have 500 or 800 castings ahead. They may all have excessive grinding on them, and the question is do you reset your original studies or just make an off-standard allowance?

CHAIRMAN FISHER: Figure 3 is the index and the paid unit hour which takes care of those variations in the author's plant. In our plant and in many other plants they make an expense allowance for that extra grinding. The standard stays the same.

R. R. MCCLINTIC<sup>5</sup>: Your standard for grinding heads, to my mind, presupposes a uniform height. Were you reasonably successful in having your heads come to the grinders with uniform height?

MR. VAN ORDER: Yes, I think we were. Before you start you should see that they are getting the castings burned properly before you even start to set standards in the jobs you are going to study. We occasionally get high heads. As a whole we have good conditions under which we study our jobs.

MR. MCCLINTIC: Do you establish your standards for molding, coremaking and cleaning from the pattern?

MR. VAN ORDER: In 90 per cent of the cases we do. We may run into certain jobs where the rate setter is not able to visualize the finished casting. You can do it yourself and sometimes you cannot visualize just how some of the cores are going to look in the finished casting. We may hold up the specification and then recheck in the cleaning room.

MEMBER: Do your men tend to relax after they have reached their task point?

MR. VAN ORDER: Perhaps to some extent. Whether men would tend to work just as hard physically for short periods and relax or work at a rather even pace over the whole day is a good question. I feel that when men are working at an excessive pace they are making a poorer product. We found from experience they have stretched it over the day more evenly, which is the way we might want to see it. We are not trying to take any pay from our men. We are trying to have them work a little slower and get a little better product at the top point.

R. J. LOGES<sup>6</sup>: Is that expense allowance you have made for the day workers' down time, or do they have checkouts for their down time?

MR. VAN ORDER: We check out for the down time.

MR. LOGES: In other words, the expense is just for the casting condition?

MR. VAN ORDER: We try to eliminate the down time if we can. We may have a minute or two here and there, which we cannot record, which the curve would help. Any item that they catch is recorded and is paid as day work.

<sup>1</sup> National Foundry Co. of N.Y., Inc., Brooklyn.

<sup>2</sup> Oklahoma Steel Castings Co., Tulsa, Okla.

<sup>3</sup> McWain Cast Iron Pipe Co., Birmingham.

<sup>4</sup> Cincinnati Milling Machine Co., Cincinnati.

<sup>5</sup> Oil Well Supply Co., Oil City, Pa.

<sup>6</sup> Kramer Bros. Foundry, Dayton, Ohio.



# A NEW PERMEABLE METAL CASTING PLASTER

By

K. A. Miericke\* and E. S. Johnson\*\*

## ABSTRACT

A new type of highly permeable gypsum metal casting plaster has been developed, embodying the following advantages:

1. Any desired degree of permeability may be obtained with the plaster.

2. Complete dehydration of the molds is no longer necessary. The free water need only be removed, leaving the chemically combined water, which is liberated as steam and diffuses readily through the permeable mold structure when the molten metal is poured into the mold. This serves as a chill to the metal and results in better metal characteristics.

3. Lower "burnout" temperatures are possible, which practically eliminates any dimensional changes in the mold. Thus is eliminated the mold shrinkage of 1.0 to 2.0 per cent common to the previous plaster formulas often resulted in dimensional inaccuracies and warped, or cracked molds.

4. Good mold strength, both in the wet and "burnout" stage, will result in less mold breakage and permit the pouring of larger castings.

5. Economies of time and fuel are possible by the use of this new plaster, since "burnout" can be accomplished in less time with less heat.

6. A far greater degree of operational flexibility is possible since the permeability and mold strength can be varied to fit any specific condition.

In spite of flexibility of this process plaster, plant control methods are relatively simple, and are based almost entirely on the wet density of the mold. Once this is established, mixing procedure can be standardized.

USE OF GYPSUM PLASTERS with various additives as a molding compound into which non-ferrous metals are cast is not a new development. For many years some brass, bronze and aluminum has been commercially cast in plaster molds. The advantage accruing from the use of plaster molds over sand foundry practice are as follows:

1. *Accurate reproduction of fine pattern surface detail.* Plaster molds are superior to the sand molds in producing smooth and finely detailed surfaces.

2. *Production of dimensionally accurate castings.* Savings in machine finishing operations required are frequently more than sufficient to compensate for the increased cost of mold materials and processing of plaster molds as compared with sand molds.

3. *Large scale production of small detailed castings,* resulting in efficient production line operations, using

multiple cavity molds. When employed for this purpose, the plaster is usually compounded with various types of fibrous and refractory aggregates such as fibrous talc and finely ground silica.

A composition widely used for the preparation of such molds comprises 70 to 80 per cent plaster of paris and 30 to 20 per cent fibrous talc. In use, compositions of this type are generally mixed with a relatively high proportion of water, in the range of 120 to 180 parts of water to 100 parts of plaster by weight.

The plaster slurry or mix so formed is poured over a pattern and allowed to set. The set mold is dehydrated completely to liberate the free and chemically combined water. This dehydration is accomplished by heating the mold at temperatures ranging from 400 to 1400 F for varying periods depending upon the size and shape of the mold. This is necessary to insure complete removal of all free and chemically combined water, since the presence of even minute amounts of water in the mold at the time of casting would result in production of imperfect castings. The heat of the molten metal would convert any water present into steam, which cannot escape readily because of low permeability of this type of mold, and would result in rough surfaces and blow-holes in the casting.

The high temperatures employed in the dehydration, or "burnout," process are accompanied by mold shrinkages ranging from 1.0 to 2.0 per cent depending on the temperature used. The higher the "burnout" temperature, the higher the degree of mold shrinkage. In addition, the strength of the mold as cast which is about 100 psi is reduced approximately 50 per cent in the "burnout."

## Advantages of Process

The chief disadvantages of this process are as follows:

1. *Size of the casting is limited* due to the low strength and low conductivity of the mold.

2. *Inferior metal characteristics* because of the slow rate of cooling of the metal.

3. *Excessive mold shrinkage* on "burnout" which often results in cracked or warped molds.

\* Metallurgical Development Engineer and \*\* Research Engineer, U. S. Gypsum Co., Chicago.

During the war years, the advantages of precision metal casting in plaster molds became known, indirectly, to many people. This led to ideas, sometimes exaggerated, as to what could be and what could not be produced by this process. Precision metal casting is a comparatively undeveloped industrial art. However, it has reached the stage where overall costs are reduced and quality of the castings is improved. In many instances, metal casting in plaster molds can be substituted for other casting methods in order to eliminate a long series of expensive machining operations of a given part. Loss, due to rejections, can be greatly reduced because rejected parts involve the loss of only the plaster molding and metal casting operations. In ordinary sand casting, porosity, or other common sand casting defects are often only discovered after a series of machining operations and the loss incurred due to rejection at this point involves not only the cost of the molding and casting operations but also the cost accumulated in the machining operations.

It was because of the rather wide-spread interest in the use of gypsum plaster for metal casting and because of the inherent shortcomings of the present processes that an intensive research program was undertaken to produce an improved metal casting plaster, which would be better adapted to the casting of non-ferrous metals and alloys. Numerous attempts have been made to produce plaster molds having a higher degree of permeability than is obtainable by the use of the above described plaster compositions and processes. The best known method for producing permeable plaster molds is the patented Antioch process. By this process molds may be produced having a variable range of permeability, but the process is lengthy, highly technical, and the equipment expensive.

#### Object of This Study

Our aim was to develop a plaster for use with non-ferrous metals which would:

1. Have a wide permeability range.
2. Produce as smooth a mold surface as is obtainable with the present plasters.
3. Have greater mold strength, both as cast and after "burnout."
4. Give better metal characteristics to any metal cast in it.
5. Possess no undesirable characteristics not inherent in the present plasters.

This research program has been largely successful in developing plaster that meets these requirements.

Molds made by this new process may be used in highly mechanized casting systems, or in small foundries. Brass, bronze, aluminum alloys and practically all types of non-ferrous metals can be accurately and smoothly cast providing one is willing to follow directions carefully until experience has been gained in using the new metal casting plaster.

This product as developed is completely formulated to include sufficient refractories to reduce thermal shock, fibrous binders to add strength and a degree of flexibility, and ingredients to produce the necessary porous structure.

Some important factors in the successful use of this newly developed permeable plaster casting medium are:

1. Mixing equipment
2. Mixing procedure
3. Pattern equipment
4. Parting or separating compounds
5. Mold "burnout"
6. Gating and risering
7. Metal temperature.

1. *Mixing Equipment*—A mechanical mixer is necessary to obtain a satisfactory mold. In order to produce efficiently and rapidly the uniform cellular mold structure, upon which the mold permeability is dependent, a special type of mixer had to be designed.

2. *Mixing Procedure*—Under this heading are the three factors:

(1) Water plaster ratio,

(2) Mixing time, and

(3) Temperature of mix by which the physical properties such as permeability, mold density, and mold strength are controlled. The effect on mold permeability of varying the water plaster ratio is shown in Table 1. In general increasing the water-plaster ratio increases the permeability, if all other factors are held constant.

The effect on mold permeability of mixing time and temperature of the mix is shown graphically in Fig. 1. For example, it can be seen from Fig. 1 that a perme-

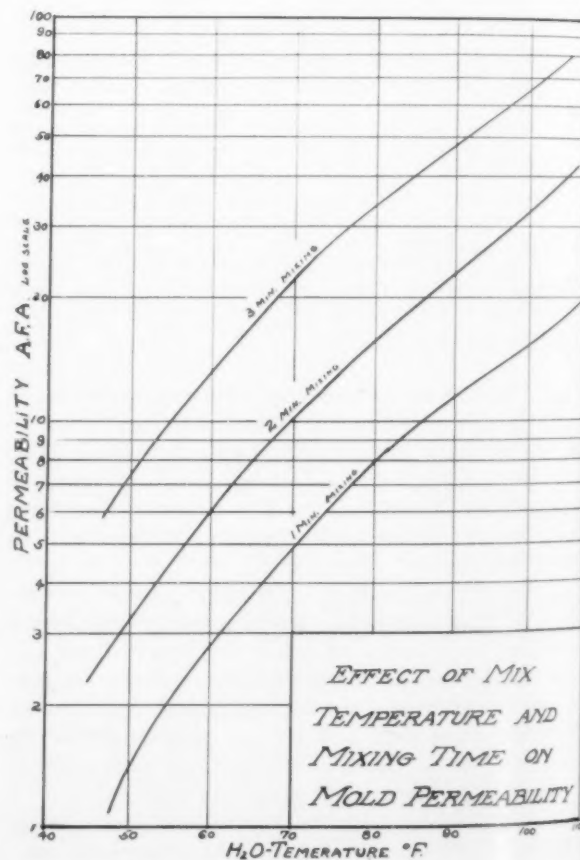


Fig. 1—Effect of mix temperature and mixing time on mold permeability.

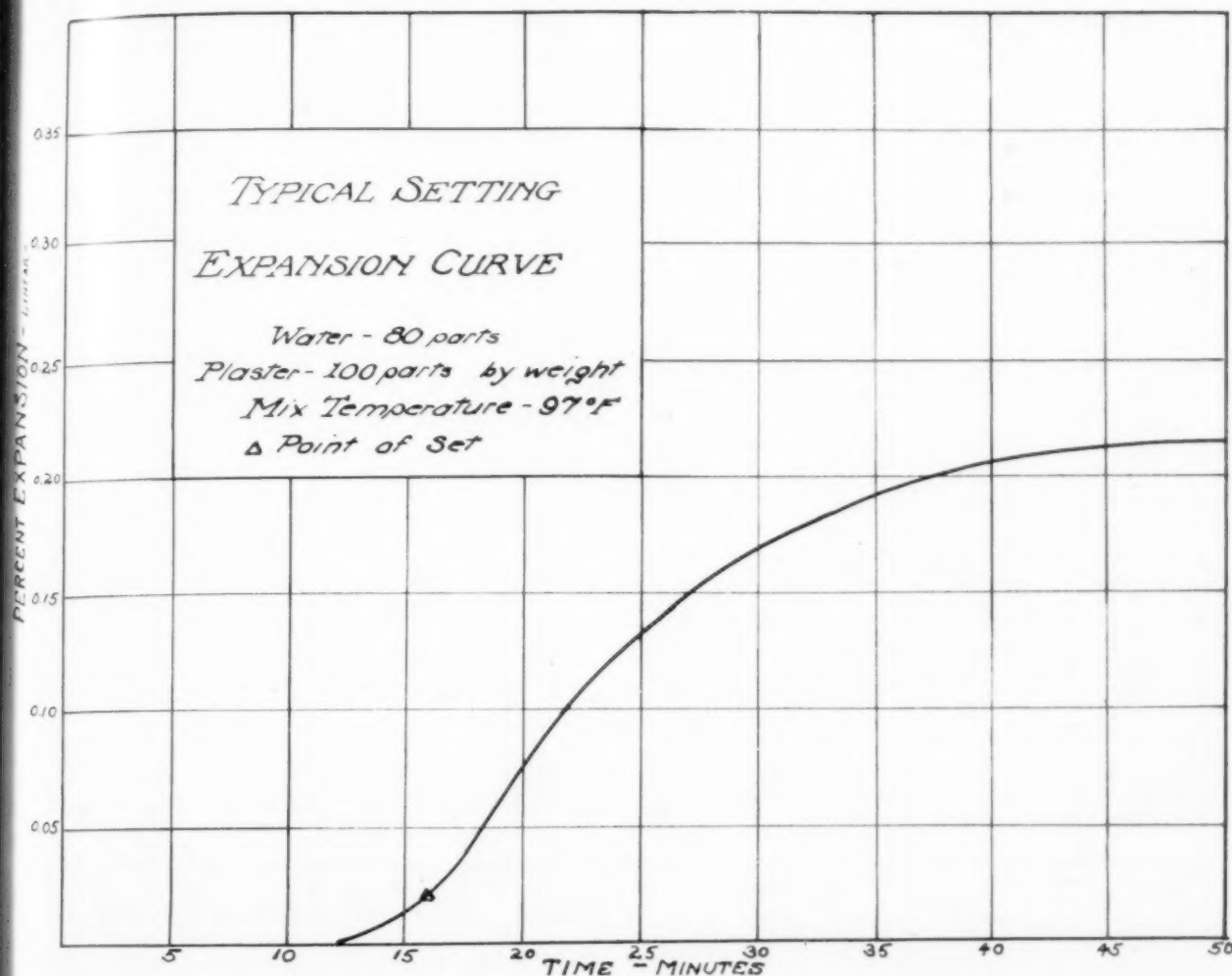


Fig. 2—Typical setting expansion curve.

ability of 11 (A.F.A. units) is obtained with a mixing time of 2 min. and a mix temperature of 70 F.

By increasing the mixing time to 3 min. and holding the mix temperature constant, the mold permeability is increased to 23 (A.F.A. units). Also, it may be noted that increasing the mix temperature from 70 to 90 F (21 to 32 C) holding the mixing time constant at 2 min., the mold permeability is increased from 10 to 23 (A.F.A. units).

Mixing recommendations call for a water plaster ratio of 80 parts of water to 100 parts of plaster by weight. Mixing time may be varied from 1 to 3 min. depending on the degree of mold permeability desired. Mix temperature should be at least 70 F (21.1 C) and

in certain cases, it is advantageous to employ slurries at elevated temperatures, in the range of 125 to 150 F (51.6 to 65.5 C). Higher mix temperature make for a more efficient mixing operation and decrease the setting expansion of the mix. The latter factor is important at times in getting mold release from patterns.

Figure 2 shows the typical expansion that occurs during the setting, or hydration, of the mix.

There is a definite relationship that exists between the wet mold density and the permeability, which provides an easy and simple method for control of permeability in the foundry. Figure 3 shows this relationship graphically.

By determining the wet density of a mix at periodic intervals during the mixing operation, it is possible by reference to the wet density-permeability graph (Fig. 3) to determine accurately the mixing time required to produce a mold of any desired permeability.

This determination is made by simply filling a container of known volume with the wet mix, weighing it, and then converting the density figure obtained to lb per cu ft. In this development work, a 2-in. x 4-in. cylinder was used and the density in lb per cu ft was obtained by multiplying the weight of the cast cylinder in grams by the factor 0.303.

Mold strength decreases as permeability increases. The relationship between these two factors is shown

TABLE 1—MIXING TIME VS. PERMEABILITY OF THE DRY PLASTER MOLD USING MIXING WATER AT 70 F

Mixing Time, Min.	A.F.A. Permeability		
	At 75 Consistency	At 80 Consistency	At 85 Consistency
1	3	3.9	5.8
2	5.6	7	10.8
3	9.8	12.3	18



Fig. 3—Wet density of mold vs. permeability.

in Fig. 4. Mold strength can also be predicted in advance from the wet density. Figure 5 shows the relationship between wet density and mold strength.

The setting time of this plaster is currently held to about 15 min., but by the use of accelerators, or retarders, it may be varied to suit any desired condition.

3. *Pattern Equipment*—Cellular compositions as described above may be formed over many types of patterns. Patterns suitable for sand practice are not generally suited for use with plaster since their surfaces are not always perfectly smooth and unless they are, difficulty in getting good mold release from the pattern will be experienced. It is imperative that the pattern equipment be excellent. Flexible patterns may be advantageous for some intricate castings, where mold release would be difficult.

The customary metal shrinkage allowance must be made on the pattern. Since much lower "burnout" temperatures are possible with new plaster no additional allowance for mold shrinkage need be made.

An outstanding characteristic of this new plaster is that it produces a smooth surface wherever it is in

Fig. 4—Mold permeability vs. wet and dry compressive strength.

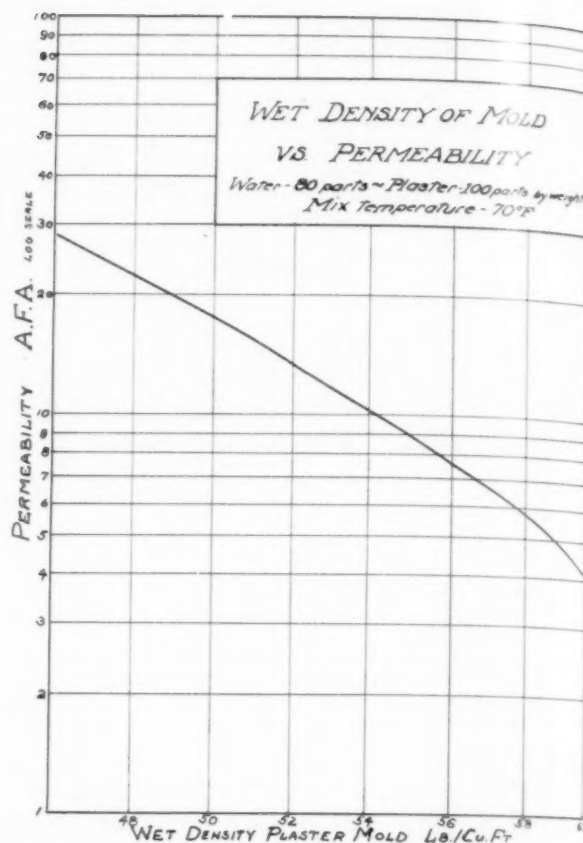
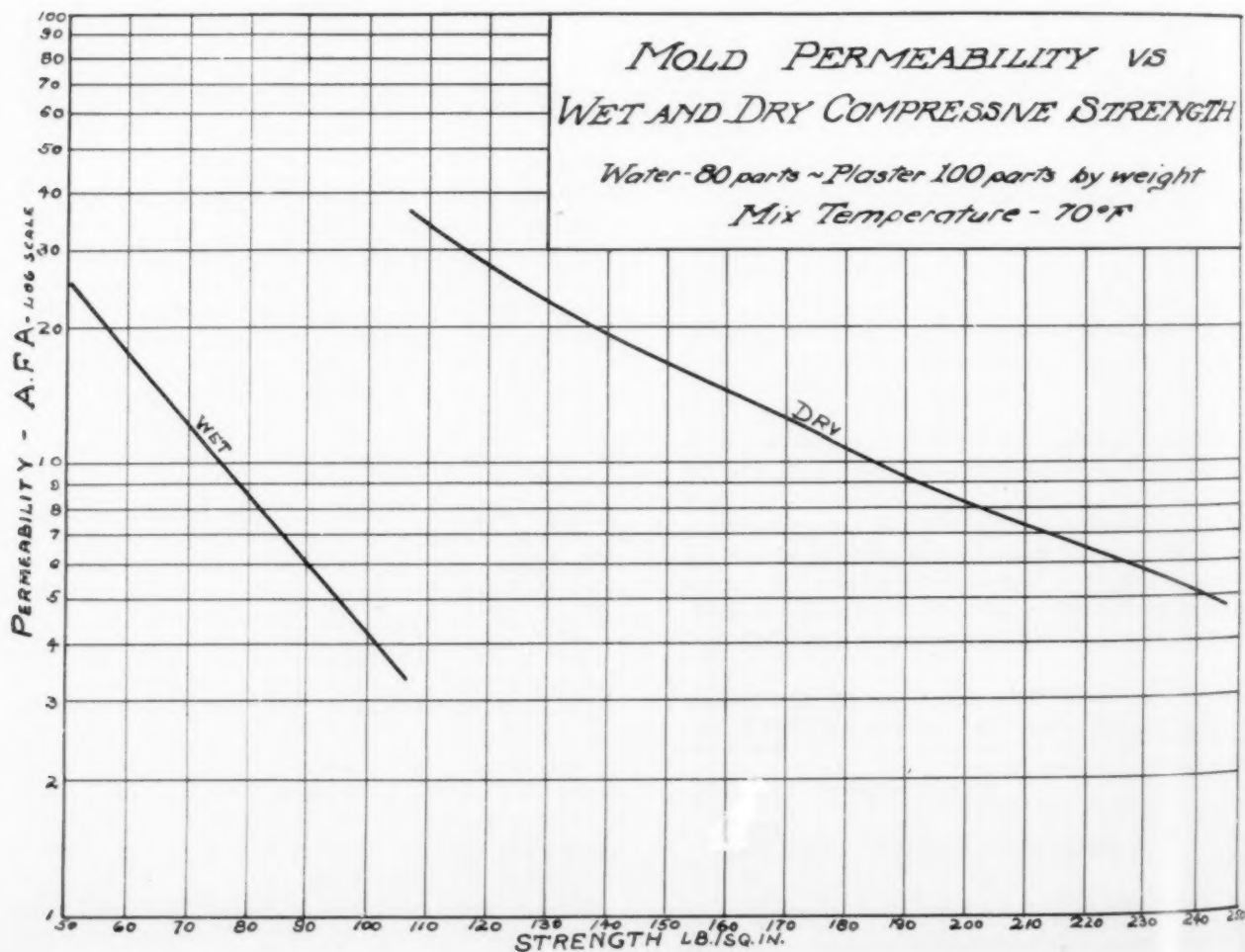


Fig. 5

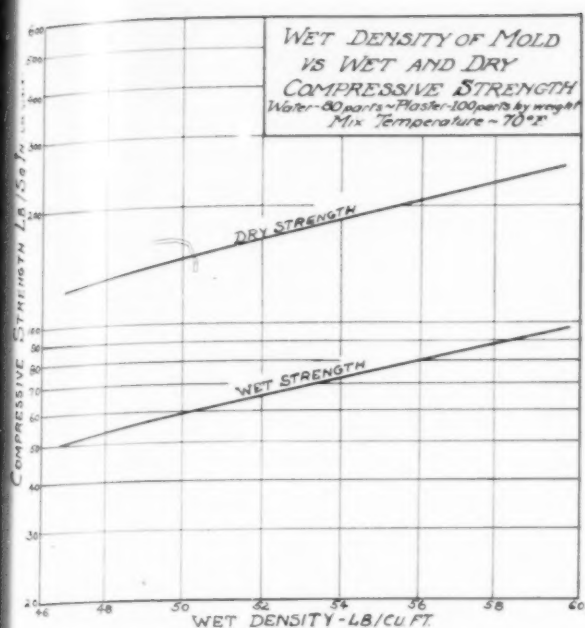


Fig. 5—Wet density of mold vs. wet and dry compressive strength.

contact with the pattern, but immediately beneath this smooth surface, the structure consists of an infinite number of tiny air cells with interconnecting porosity. This is clearly shown in Fig. 6, 7 and 8.

4. *Parting or Separating Compounds*—Commonly used separating or parting compounds to effect mold release from the patterns include paraffin wax suspensions, or solutions, and oil-wax emulsions. Anti-foaming agents may be incorporated in the separating medium to increase slightly the thickness of the microporous, non-cellular plaster film that forms adjacent to the pattern surface. This aids in preventing surface film rupture in pattern removal and also during the metal casting operation.

5. *Mold "Burnout"*—The length of time required for the proper degree of "burnout" or drying, before casting the metal depends upon several factors.

1. Type and degree of heat employed,
2. Extent of circulation of heat in the oven,
3. Size of the mold, and
4. Ratio of surface area to volume of the mold.

Consequently, no accurate predictions can be made by "burnout" time. The proper time cycle for any given mold may be determined by periodic weighing of the mold during "burnout," or by following the internal mold temperature by means of an imbedded thermocouple.

After the mold has been removed from the pattern, it should be uniformly supported face up on a flat perforated metal plate, or on a rigid metal grill and placed in the drying oven.

When the plaster is mixed with water, the dry plaster combines chemically with 14 per cent of its dry weight of water during the setting action, or hydration. The water in excess of this 14 per cent must be evaporated before casting of metal. This excess water amounts to about 34 per cent of the wet mold weight

if a water plaster ratio of 75 lb of water to 10 lb of plaster was used and about 37 per cent if a water plaster ratio of 80 lb of water per 100 lb of plaster was used. It is therefore necessary to dry the mold until it has lost the 34 or 37 per cent of its wet weight.

#### Calcining the Mold

If molds of lower permeability (3 to 7 A.F.A. units) are used, it may be necessary to calcine them by drying them until they lose up to 46 per cent of their wet weight by removal of a part or all of their chemically combined water. The initial drying temperature is not too critical and may be as high as 600 F (315.5 C) since the actual mold temperature will not rise much above 212 F (100 C) until all the free moisture is evaporated. When the bulk of the free water has been removed, it is best to lower the oven temperature to about 250 F (121 C) until the molds are completely dry.

If unvented plaster cores are to be used in the molds, they must be carried to partial, or complete calcination at approximately 400 F (204 C) to eliminate danger from release of steam during casting.

Due to the low "burnout" temperatures possible with this new plaster and the resultant low mold shrinkage it is possible to use metal flask equipment and permit the flask to remain on the mold until the castings have been poured. This method of handling will permit foundries to handle much larger and

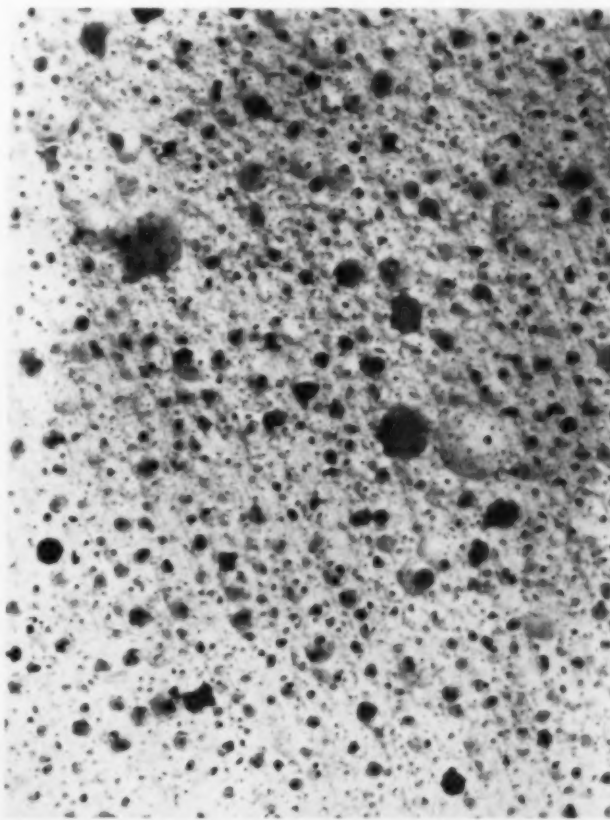


Fig. 8—Macroscopic view at 20X of the permeable mold structure shown in Fig. 7.

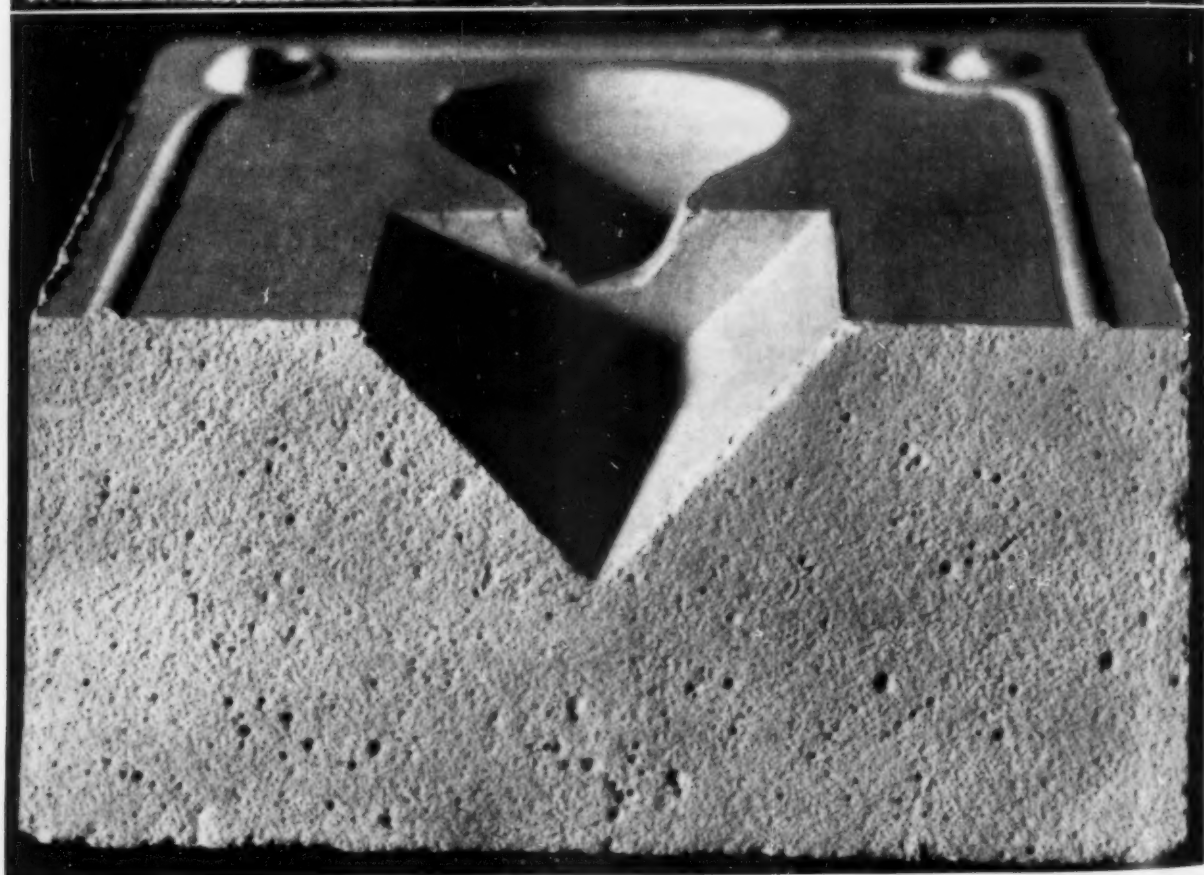
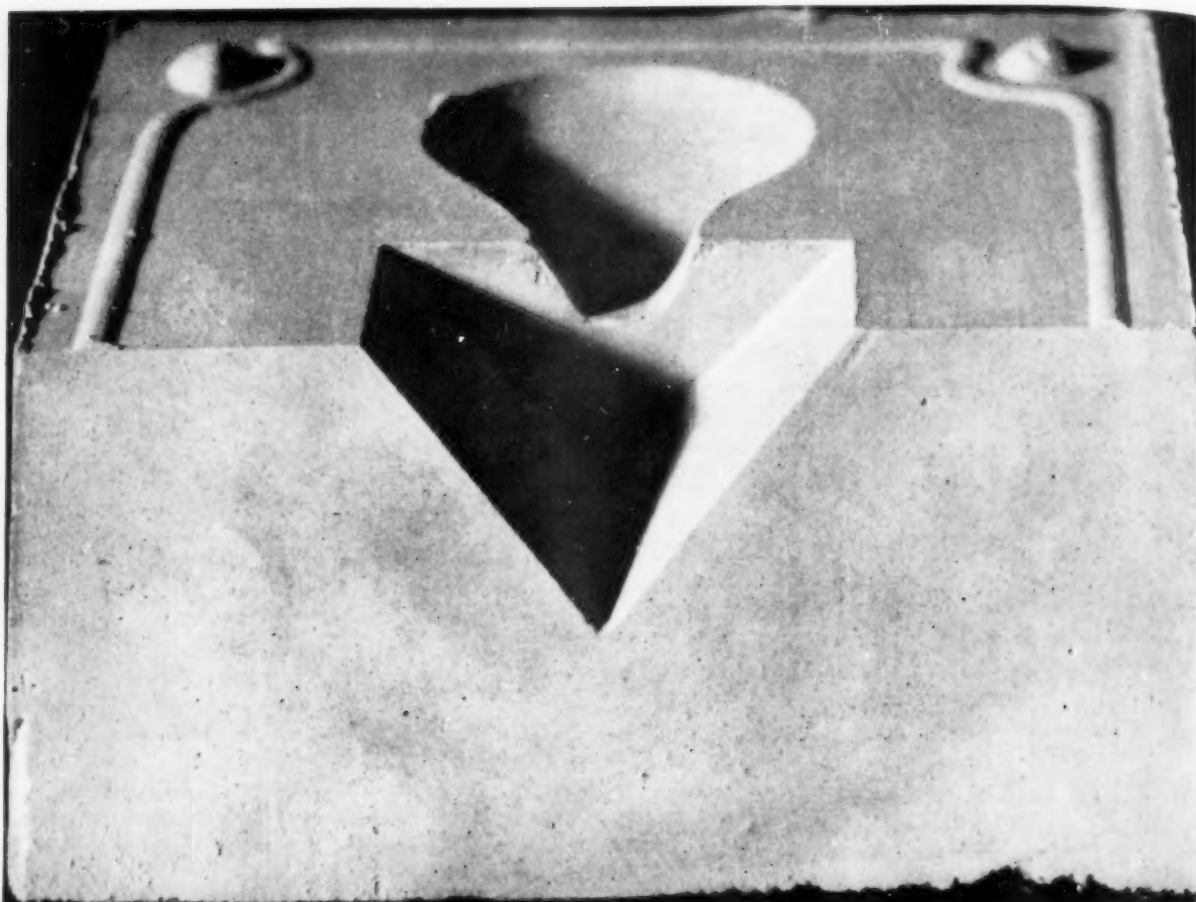


Fig. 6 (top)—Mold, 80-20 plaster, 160 consistency.

Fig. 7 (bottom)—Structure of a permeable mold.

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more complicated molds than were hitherto possible, since mold registering methods can be retained.

It is also possible to regulate mold strength so as to preclude any possibility of "hot tear." This is done by increasing the permeability, or by calcining the mold surface slightly, which gives a result comparable to the soft ramming of sand. Usually, during the pouring of the metal, this latter is automatically accomplished since the hot metal in contact with the plaster actually calcines it and reduces the strength of mold surface by as much as 50 per cent. Mold porosity permits internal crushing during metal freeze.

6. *Gating and Riser*ing—Because of the insulative nature of the molds made from this plaster, gates may be made smaller than in sand, resulting in a less turbulent flow of the molten metal. On experimental molds, sprues and gates may be carved in the wet molds with ease. However, the carved surfaces should be smoothed carefully to prevent loose plaster particles from washing into the cavity during the metal pour. In standard production, all gating should be part of the pattern. As is customary in other type molds, it is good practice to vent high portions of the mold to aid in gas release.

7. *Metal temperatures*—Since this new plaster is less conductive than other molding materials, metal may be cast at lower temperatures than is common for sand molds. It is advisable to keep casting temperatures of heavy castings below 2000 F (1093.3 C), or some chemical decomposition of the gypsum may occur with subsequent metal attack. Small thin-sectioned castings may be cast at temperatures up to 2300 F (1260 C). The metal should be well skimmed and poured without much delay.

Field trials of this new plaster have been made in the matchplate industry, precision metal casting field, and as cores for permanent molds and sand casting and the results to date are highly favorable. Additional research is being conducted for the improvement of this product, correlation of field data, and further adaptation of this product to the metal casting industry in general.

## DISCUSSION

Chairman: R. W. PARSONS, Ohio Brass Co., Mansfield, Ohio

Co-Chairman: A. K. HIGGINS, Allis-Chalmers Mfg. Co., Milwaukee

A. K. HIGGINS (*Written Discussion*): In the past, the application of plaster molds to the casting of the random parts that are found in even a small foundry has necessitated a refinement of control, and a degree of skill in the development of process, that was beyond the reach of most of us. Where single designs were ordered in large enough quantities to justify the needed development, plaster casting has been, and remains attractive. It has not been applied successfully (to the best of my knowledge) without a rigidity of control and process that would horrify the average foundryman. (It is interesting to speculate on the probable changes in casting quality that would result if the same conditions were applied to the sand foundry).

Those of you who have not tried plaster casting will find hard to believe, the amount of warping and distortion that may occur in a plaster mold when drying is not uniform, or even under good conditions when mold symmetry is bad. The new type of plaster proposed by the authors appear to offer some release from these problems. The new plaster would appear to offer possibilities for use of modern types of equipment to secure maximum production rates in minimum space.

As an example, the excess water can be removed rapidly and uniformly by the use of di-electric heating. The cope and drag of a 12-in. x 14-in. mold (4 in. overall) should dry in 10 min or less with power from a 20 kw generator. If calcination of the casting surface is required, the use of infra-red would seem logical. In any case, the total time from pattern to casting could become a matter of minutes rather than hours. Certainly the high permeability of the plaster removes the danger of exploding the mold by too rapid heat input. We have dried samples of this plaster by the above method but have not as yet cast metal into them.

I would like to thank the authors for a valuable contribution to the non-ferrous casting industry.

MR. MIERICKE: Commenting on Mr. Higgin's method of di-electric drying, it is one method we did not try, but I can see no reason why it would not work. It would certainly be a fast method of drying. We have done some work with infra-red ray drying and found it successful.

In most instances we have used a gas or electrically-heated core oven. The drying problem will have to be worked out in your individual plant. We know how much water has to be taken out; how it is removed will depend on the facilities available. But one can see from the amount of water that has to be evaporated and the temperatures required that it is considerably less than was previously necessary, and should be less expensive. Drying at 1500 F in a radiant oven was rather costly and caused a great deal of warping and shrinkage.

W. O. WETMORE:<sup>1</sup> I have three questions about this material. We have been unable to obtain satisfactory results with information and materials supplied us in the past.

1. Is this new casting plaster available, and are the data given in the paper applicable to its use?

2. Can cores of this material be removed by any method except hand chipping?

3. How thick a section of aluminum can be cast around a 1½-in. core of this material? We are using a green sand mold and desire a very smooth finish on inner surfaces which cannot be machined.

MR. MIERICKE: The material is available to a limited extent. We are not trying to put out a product that is not fully developed. We can make good castings and we have had failures. Hand chipping is not required to remove the cores.

I cannot give you a definite answer to the third part of your question. This practice should be quite similar to sand practice because your core crushing strength can be regulated by calcination and by permeability.

HIRAM BROWN:<sup>2</sup> Have you tried this investment casting with the lost pattern process? Would you briefly go over the baking cycle, temperature and time on that new permeable mixture?

MR. MIERICKE: Frankly, this material may not be suitable for the lost pattern process, especially for wax and polystyrenes, since the porous structure may cause wax or plastic absorption. We have other materials available for the lost wax. Some work was done in using a non-permeable material as a precoat, then filling the rest of the mold with this permeable material as the two will adhere. The temperature of your mold must be raised sufficiently to remove all wax and plastic. The temperature required to remove wax and other residue (800 F approx.) will calcine the mold completely.

In answer to the second part of Mr. Brown's question, I have dried some highly permeable molds for aluminum by leaving them in a high velocity oven at about 150 F overnight and made sound castings with good surfaces. On other molds we started with an oven temperature of 500 F for one-half hour, then lowered it to 250 F and had our drying cycle completed in 3½ hr. The drying cycle will vary with each installation and the number of molds to be dried. If the oven is loaded, there is more water to be eliminated.

MEMBER: Does this new plaster have greater tolerance for the lead brasses and bronzes?

MR. MIERICKE: The 85-5-5-5 alloy still gives trouble. You can get somewhat better results by going to partial calcination. Even with the permeability available any alloy containing over 1½ per cent lead will give trouble. However lower melting lead alloys may be cast successfully.

<sup>1</sup> U. S. Naval Ordnance Test Station, Inyokern, Calif.

<sup>2</sup> Solar Aircraft Co., Des Moines, Iowa

# ELECTROCHEMICAL CLEANING OF A LARGE STEEL CASTING

## AN EXPERIMENT

By

John A. Wettergreen \*

### ABSTRACT

*An experiment was conducted to determine the possibilities of removing fused sand and iron oxides from the interior of a large cast steel turbine shell with the use of an electrolytic molten caustic bath. The major problems of handling the casting and auxiliary equipment were satisfactorily overcome. The portion of the casting immersed in the bath was successfully cleaned of fused sand and oxides. The process offers possibilities for development in step with increasingly severe service requirements.*

DURING THE RECENT WAR a process was developed at one of the company's plants for electrolytic molten caustic cleaning of small parts. This process was successfully used over a period of several years for the cleaning of small castings and forgings. Sand, scale, grit, dirt, oils and greases were satisfactorily removed prior to tinning, plating and soldering operations. The results obtained by this method were far superior to any other method available, both from the standpoint of time required and effectiveness of results. This early work was the basis for the cleaning of a large casting as reported upon in this paper.

### Experimental Installation

As shown in Fig. 1, the experimental installation consisted of a motor generator set rated at 5000 amperes at 6 volts, a fabricated steel tank with immersion type gas burners, and the necessary fixtures for handling the casting. The tank, which was 8x10x4 ft, was sunk below floor level in an abandoned molding pit.

Since it was planned to use this installation for only one cleaning cycle, it was necessary to provide means for disposing of the molten caustic upon completion of the experiment. The tank was provided with a drain near the bottom controlled by a gate valve. The valve was manually operated. The handle for operating this valve may be seen in Fig. 1. This drain discharged into another pit adjacent to the tank pit. A crane-suspended ladle was filled with molten caustic and the caustic poured into 55-gal steel drums. After solidification the caustic was diverted to other uses.

The complete lifting device and electrode holder is illustrated in Fig. 2. This apparatus was assembled to the casting prior to preheating of the casting and remained in place until completion of the cleaning cycle.

\* Formerly with General Electric Co., Schenectady, N.Y., now with Monighan Div. Bucyrus-Erie Co., Chicago.

The electrode supporting rod was constructed of 1½ in. diameter copper above the bath level and 2 in. diameter steel below the level of the bath, with four bus bars each ¼ x 4 in.

Approximately 19,000 lb of flake caustic soda was charged into the preheated tank. The remaining 11,000 lb was added in smaller batches at intervals during the first 24 hr of melting. A total time of 38 hr was required to completely melt the caustic. During the melting operation it was determined that the melting temperature was 300 C.

No particular difficulties were encountered in melting the caustic, aside from the time consumed. This time is not considered excessive in view of the large amount involved in the experiment. The temperature of the molten caustic was easily maintained between 400-450 C. The bath temperature after melting was quite uniform, with a maximum variation of approximately 10 C between different points in the bath at any time. All bath temperatures were taken with an immersion type recording pyrometer.

### Preheating the Casting

The complete assembly of casting, lifting device, and electrode holder with electrodes in place, as shown in Fig. 2, was preheated in an electric car bottom furnace to 400 C. The casting temperature, as measured by a

Fig. 1—Developmental installation for experimental cleaning of steam chest portion of cast steel turbine shell by electrolytic molten caustic method.



contact pyrometer during the transfer of the preheated casting from the furnace to the tank, was 300 to 325 C. These readings were taken approximately 5 min after the car bottom was pulled, and when the casting was resting on the foundry floor at some distance from the furnace. The purpose of preheating was to eliminate all moisture and prevent chilling the bath of molten caustic.

### Introducing Casting Into Molten Caustic

An overhead travelling crane was used to slowly lower the casting into the caustic bath. Great care was exercised in this operation because it was realized that there was danger of air trapped in the cored openings expanding when heated, and ejecting molten caustic from the casting cavity. Although considerable agitation occurred within the casting, there was no splashing of caustic beyond the confines of the tank and no undue hazard to operators. The greatest amount of agitation seemed to come from gases generated by the reaction between the caustic and silica.

After the casting was in place on the supports provided, the crane was disconnected and electrical connections made. One side of the bus was connected to the electrode support bracket, and the other to the lifting device. The connection consisted of four  $\frac{1}{4} \times 4$  in. copper bars for each side of the circuit. It was necessary to clean the oxide from the copper bars before bolting up the connection. This required about 30 min, and it was necessary for the electricians to wear protective clothing and work behind asbestos shields because of the reflected heat.

### Oxidizing Cycle

With the casting positive, power was applied for approximately 3 min at 2200 amperes. This resulted in an oxidizing condition which would have removed any organic material present, such as grease or oil. Actually, this cycle was unnecessary because of the preheating to 400 C.

The polarity was reversed and cleaning was continued at approximately 4000 amperes and 6 to 7 volts for approximately 2 hr. During this cycle, considerable agitation occurred in the interior bores of the casting, particularly around the electrode ends. There was some evolution of gas and caustic vapor from the tank. This vapor was objectionable only when operators placed their heads directly over the tank.

At the end of this 2-hr cleaning cycle, the power was shut off, the leads disconnected and the casting removed from the bath for inspection. It was observed that a considerable amount of cleaning had been accomplished. In some areas the cleaning was incomplete and, in the particular area where fused sand had been observed prior to cleaning, an appraisal of the results was impossible because of inability to observe this portion while the casting was suspended from the crane. Observation of the casting required approximately 30 min.

### Further Cleaning

The casting was returned to the bath, electrical connections restored, and cleaning resumed. At the end of the  $2\frac{1}{2}$ -hr period of additional cleaning, the casting was removed and the caustic allowed to drain off. The temperature of the casting upon removal from the bath, as measured by contact pyrometer, was 360 C on

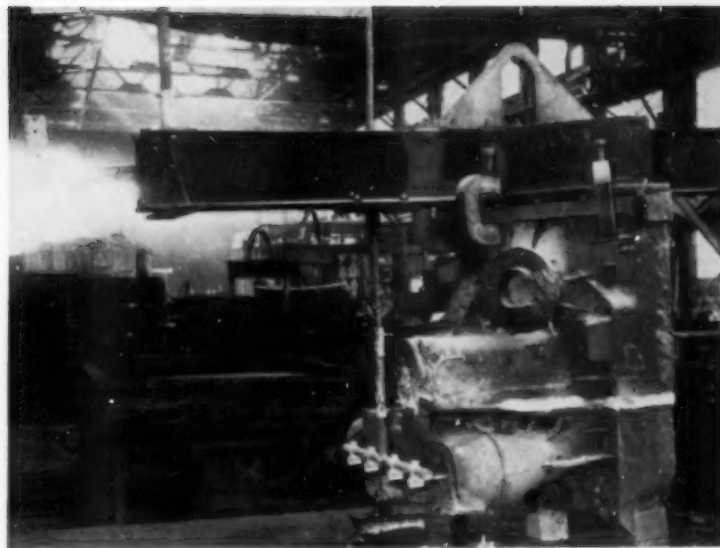


Fig. 2—Steam turbine shell casting with lifting device and electrodes in place, after steam chest portion of casting was experimentally cleaned by electrolytic molten caustic method. Note surface conditions above and below line marking immersion depth.

the cleaned surface and 260 C on the portion of the casting which had not been immersed in the caustic.

As may be seen in Fig. 2, only the steam chest portion of the casting was cleaned during the experiment. The reasons for confining the cleaning to this portion of the casting were threefold:

1. Most severe conditions of fusion and penetration are encountered in the steam chest portion of these castings. Any results accomplished on this portion of the casting would be easily duplicated on other parts.
2. The steam chest portion of the casting is the part which is subjected to the high pressures and temperatures in service, and it is therefore the most critical.
3. By confining the cleaning to the steam chest end of the casting, it was possible to keep the installation within practical size and cost limits.

### Washing the Casting

After cooling to room temperature the casting was removed to an isolated yard area and washed with a fire

Fig. 3—Caustic being washed from cast steel turbine shell, steam chest portion of which was experimentally cleaned by electrolytic molten caustic method.







hose. It required approximately 6 hr for this operation (Fig. 3). Checks for remaining caustic were made by the use of litmus paper and phenolphthalein.

The casting used for this experiment is known as a "turbine shell lower," with an analysis of carbon, 0.20, manganese, 0.76, silicon, 0.37, and molybdenum, 1.10 per cent. Pouring weight was 43,300, and shipping weight 18,000 lb. This is a representative example of the larger, more complicated, and critical castings made in the Schenectady Works steel foundry of the company for steam turbine use.

Because of the intricate coring and a design involving extremely heavy sections, this type of casting is

*Fig. 4—Cast steel turbine shell, steam chest portion of which was cleaned by electrolytic molten caustic method. Close-up view showing typical fused sand conditions on exterior fillets not cleaned.*

subject to the occurrence of considerable fused sand and metal penetration in the outside fillets and interior cored openings (Figs. 4 and 5). By close control of raw materials and foundry practices, these conditions have been minimized to the extent that they present less of a problem than was the case in the past. However, even under present improved conditions cleaning represents a major element of cost.

The particular casting chosen for this experiment contained an unusual amount of fused sand and penetrated metal. It was conservatively classified as "worse than average" before caustic cleaning.

#### Penetration and Fusion

Mechanism of metal penetration into sand has been the subject of considerable investigation and theorizing. Its removal from steel castings has always been a major cleaning problem, particularly in foundries producing heavy-section castings. As normally encountered it consists of a network of metal particles of minute cross section mixed with fused sand (Fig. 4). In the removal of this material, shotblast is quite ineffective and chisels are readily dulled. If conditions permit, a chisel can sometimes be inserted between the casting and this conglomerate, and pieces can be removed by prying. Sometimes it can be pulverized by

*Fig. 5—Steel turbine shell, steam chest portion of which was cleaned by electrochemical process. Compare cleaned and uncleaned portions of casting. Steam chest only was cleaned. Note condition of fillets and absence of burning slag on horizontal joint at cleaned end.*



lengthy pounding with a pneumatic hammer and chisel. In inaccessible cored openings this is quite time consuming.

In the case of the subject casting the electrolytic molten caustic cleaning process completely removed all traces of silica particles except in the two center pockets of the port chamber. In this area fully 80 per cent of the fused sand was removed, and it is certain that a brief extension of the cleaning cycle would have completed the removal. The small amount of fused sand remaining, due to too short a cleaning cycle, was quite fragile, and the author was able to pulverize it with a cold chisel and machinist's hammer (Fig. 6).

Several samples of penetrated metal, after removal of sand in the caustic bath, were combined and analyzed for composition. These results are compared with casting composition as follows:

Component	Casting	Penetrated Metal
Carbon, per cent .....	0.20	not analyzed
Manganese, per cent .....	0.76	not analyzed
Silicon, per cent .....	0.37	0.22
Molybdenum, per cent ....	1.10	1.38

These results prove conclusively that the penetrated metal is not ferrous silicate nor iron oxide, as has been claimed by other investigators. The material is definitely the same as the casting. Still unexplained is the mechanism which will permit steel tapped at 2850 F and poured from a 20-ton bottom pour ladle to "run needles."

Following molten caustic cleaning the casting was heat treated at 1050 C. After heat treatment the metal penetration was found to be quite fragile so that almost all of it could be rubbed off with the finger tips. Apparently, it had become completely oxidized (Fig. 7).

Figure 8 illustrates the progressive removal of fused sand from this casting. In the case of (A) the material is shown as normally encountered; (B) some removal of fused sand has taken place; (C) illustrates complete removal of fused sand with the penetrated material remaining. Laboratory indications confirm these observations as follows:

	A	B	C
SiO <sub>2</sub> , per cent .....	43	46	Trace
Metallic Iron (Fe), per cent .	56	31	97
Iron Oxide, per cent .....	Small Amount	21*	Small Amount

\* Probably due to anneal.



Fig. 7—Cast steel turbine shell, steam chest portion of which was cleaned by electrochemical process. Close-up view shows easily removed searched or penetrated metal remaining in exterior fillets after heat treatment at 1050 C and after removal of fused sand.

The casting which was the subject of this experiment had not been annealed or heat treated. As a result no evaluation of the efficiency of this process in removing annealing scale is possible in connection with this particular experiment. However, in the previous work done with small parts in another plant of the company, it was conclusively demonstrated that this process readily removed annealing scale.

Also, in the case of the present experiment, oxide resulting from prior flame cutting operations was readily removed. This is evident in Fig. 5, where the portion of the heavy flange which was immersed in the caustic is entirely clean of flame-cutting oxide and the remainder of the flange has a heavy coating. If this process were applied to production operations it would be desirable to perform the cleaning after annealing and heat treating operations. This procedure was impossible in the case of the subject experiment because of production requirements.

#### Experimental Installation

Cost of the experimental installation for conducting this cleaning process was considerable when related to its value in the cleaning of one casting. However, analyses of this installation cost and the operating cost for the experiment indicate that they are not so high as to preclude the possibility that a production installation could be economically justified.

Analyses of foundry cleaning costs on the subject casting indicates that, as a result of the electrolytic cleaning process, cleaning direct labor costs were reduced by one third. These costs would have been further reduced if the casting had been previously heat treated so that the electrolytic caustic cleaning method would have reduced or eliminated a shotblast operation.

From an engineering standpoint, there seems to be considerable merit in the further investigation of this

Fig. 6—Steam turbine shell casting. Steam chest portion was cleaned by electrolytic molten caustic method. Close-up view shows condition of cored openings in port chamber after cleaning. Some fused sand, which is quite fragile, remains after too short cleaning cycle.



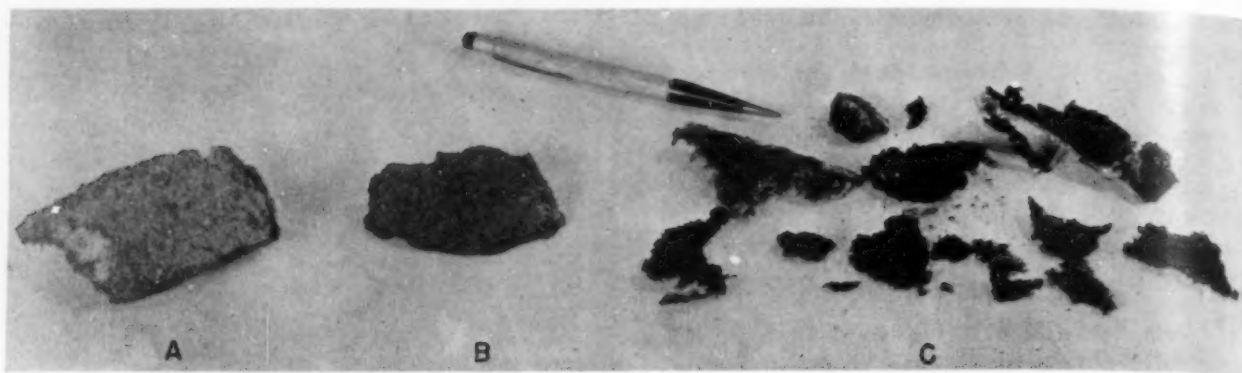


Fig. 8—Fused sand from steam chest portion of cast steel turbine shell cleaned by electrolytic molten caustic cleaning method. A—Piece of fused sand and metal penetration removed from portion of casting not cleaned by this method. B—Piece of fused sand and metal penetration removed from one of center pockets of port chamber. Note change in color. See report for chemical analysis. C—Pieces of metal penetration after complete removal of fused sand in molten caustic bath.

process since the requirements for steam turbine castings are becoming more severe. Operating temperatures and pressures are steadily increasing. The production of better castings to meet these increasingly more rigid requirements must involve the development of better foundry methods and controls.

Conventional hand methods for the removal of fused sand from cored passages in complicated steam turbine castings are slow and in some cases uncertain. Despite close supervision, qualified operators, and the best of equipment, it is conceivable that 100 per cent removal of sand particles from steam passages may not always result. Chemical dissolution of sand deposits is a certain method of assuring absolute freedom from embedded and fused sand.

Hand methods of removing sand from castings are not subject to good control. Controversy sometimes results from the effort to equitably compensate operators for this type of work. A chemical method of performing this operation would seem to lend itself to better control and elimination of controversial factors.

### Safety Problems

Conduction of this experiment involved protection against a number of hazards. Operators were protected by asbestos clothing and rubber clothing as required. Respirators and splash-proof goggles were provided. Guard rails were installed and provision was made to operate heating burners from outside the tank.

Every possible precaution was observed in the handling of the flake caustic and molten caustic to prevent the possibility of this material coming in contact with the operator's skin. All operators were carefully instructed regarding the hazards of handling the caustic. As a result of the precautions taken no accidents were encountered in connection with this experiment. Upon completion of the experiment, it was generally agreed by all concerned that molten caustic in a large quantity, such as was involved, could be successfully and safely handled without encountering any undue hazard.

Experience gained from this experiment indicates that foundries producing complicated steel castings for high temperature and high pressure service might profitably investigate the possibilities of utilizing electrochemical methods for cleaning certain types of castings. Indications are that demands for higher quality steel castings may involve the development of methods which will insure the production of castings entirely free from embedded fused sand. This method offers possibilities from the standpoint of obtaining better control of labor input for cleaning operations.

### DISCUSSION

Chairman: FRANK KIPER, Ohio Steel Foundry, Springfield, Ohio

Co-Chairman: V. E. ZANG, Unitcast Corporation, Toledo, Ohio

C. E. SIMS: <sup>1</sup> I do not believe you said which was the anode and which the cathode in that cleaning operation.

MR. WETTERGREEN: The power was applied in one direction for a brief period of time. It was reversed and remained in the other direction for the greater part of the experimental cleaning cycle, the idea being that in one direction the operation was oxidizing and in the other direction, reducing.

MR. SIMS: Can you tell us which was more effective, or which way it was on the longer time?

MR. WETTERGREEN: It first started out oxidizing and then the greater part of the time it was reducing.

A. W. GREGG: <sup>2</sup> The author stated that he reduced cleaning costs 33 per cent. Did that include power and amortizing the equipment?

MR. WETTERGREEN: The total foundry cleaning operations on this casting were reduced by one-third.

MR. GREGG: That meant labor only?

MR. WETTERGREEN: That is only labor.

G. C. DICKEY: <sup>3</sup> Suppose we are about to clean a small casting and are trying to determine the amount of current in amperes to use in doing the job. Should we try to estimate on metal section or area?

MR. WETTERGREEN: The earlier work that preceded this experimental job, which was carried out on smaller parts, both castings and forgings, indicated that approximately 200 amperes per sq ft was required to do a good cleaning job.

CHAIRMAN KIPER: What was the weight of that steel casting used in this experiment?

MR. WETTERGREEN: The weight of this steel casting was approximately 15 tons.

R. E. KERR: <sup>4</sup> I was curious about one point. The casting was cleaned while it was in the green. Is that correct?

MR. WETTERGREEN: That is right.

MR. KERR: There was a statement made that the penetrated metal remaining after caustic cleaning was highly oxidized during heat treatment and was therefore easily removed. There was also a statement made that the cleaning could be done, or

<sup>1</sup> Battelle Memorial Institute, Columbus, Ohio

<sup>2</sup> Whiting Corp., Harvey, Ill.

<sup>3</sup> Harrison Steel Castings Co., Attica, Ind.

<sup>4</sup> Pettibone Mulliken Corporation, Chicago



maybe even improved, if the heat treatment were performed before the caustic immersion. I was wondering if this is true. If heat treatment were performed first, you would not experience the beneficial oxidation of the penetrated metal to as great a degree and any of it remaining on the casting after caustic cleaning might not be so easily removed.

MR. WETTERGREEN: That statement was based on the thought that there is a shot-blasting operation involved in removing the annealing or heat treating scale and that if this operation were performed after the heat treatment, it would remove that scale.

MR. KERR: In the paper the author states that the casting is a green casting. He cleaned it electrochemically and then heat treated it. Prior to heat treatment, there was metal penetration, or the effects of metal penetration, still remaining on the casting. Heat treatment oxidized that metal to the extent that it was easily removable.

MR. WETTERGREEN: That lacy network of metal penetration which remained on the casting after the cleaning cycle was somewhat ductile before the heat treatment. After heat treatment it was brittle, but in either case, ductile or brittle, it was an easy matter to remove it with a shot blast.

MEMBER: Do I understand that you could anneal a casting with penetrated sand and thereby loosen up the penetrated sand so it would come off easily? If you have penetrated sand on the casting and you anneal it long enough to loosen the sand you will not have any casting left.

MR. WETTERGREEN: There is a slight benefit in the removal of this material by the annealing operation, but it is rather minor.

CHAIRMAN KIPER: We have had some experience in our foundry with castings having burned-in sand and know that it is quite a problem. Heat treating of the casting has little effect on the subsequent cleaning of the burned-in sand.

JOHN HOWE HALL: Prolonged heat treatment of the casting will have no effect on the subsequent cleaning of the burned-in condition.

MR. WETTERGREEN: Does anybody know how to eliminate fused sand on castings?

C. W. BRIGGS: To prevent penetration my suggestion is to not use sand. The Germans have been making the same kind of castings for some time and they do not have the amount of penetrated material on internal surfaces that we do. They use chamotte.

CHAIRMAN KIPER: To what extent has chamotte been used in this country?

MR. BRIGGS: On large castings it has not been used at all.

There is a material comparable to chamotte that has been produced in this country in the last few years. It changed its trade name several times but at present I believe it is manufactured by the Harbison-Walker Co. I think they named it Calamo. It is a calcined clay. It is broken down to various grinds. I believe the usual method is to mix a couple of different grinds together. The ground material is bonded with a clay which is of the original material prior to calcining. It is aluminous clay, and that gives a rather stable material, but one which is quite open. A fairly thick coat of wash which consists for the most part of the clay and some of the fine calcined material, and in some cases graphite, mixed together is then applied to the mold which is dried by torches.

This material has been used not only as molding sand but also in cores. A large number of cores are made from it in Germany and, in the larger steel plants in Germany, it is almost entirely used on castings and cores. I saw some rather large and complex castings, much like the one which is shown in the paper, in all the various stages of manufacture and cleaning, and I was quite surprised at the small amount of burned-in material in their passageways. The Germans claimed that it was largely due to this use of the chamotte type of material.

CO-CHAIRMAN ZANG: Is that material used in the Fischer process, or connected with the Fischer process?

MR. BRIGGS: Yes, that is a standard method of molding in Germany and always has been, apparently. In fact, practically 60 per cent of all molding is done in chamotte on the continent.

CO-CHAIRMAN ZANG: In drying, is that material highly burned, then?

MR. BRIGGS: They have used various temperatures. In some

plants, they went up as high as a dull red, but I understand that the practice in this country is to only go up to 400 to 500 F.

MEMBER: How would you proceed to make a core out of that material?

MR. BRIGGS: It is rather readily moldable by standard methods.

MEMBER: Is chamotte just a clay?

MR. BRIGGS: It is the mixture of ground calcined clay and uncalcined clay that I described previously. There are German companies that prepare and sell it just like we have companies that sell sand. There are companies that prepare this material and sell it to the foundries who only have to add water to the chamotte to use it for molds.

MEMBER: How would you use the American counterpart for chamotte?

MR. BRIGGS: You proceed by mixing two different grain sizes of this material together, add the raw clay and water, and then make the mold.

MEMBER: Then how do you dry the mold?

MR. BRIGGS: In this country the practice is to dry the mold at 400 to 500 F. The foundry then applies a wash and torch-dries the mold. That is essentially the way in which it is done. I imagine there are a lot of variations to the process.

MR. HALL: I would like to ask Mr. Briggs if he is perfectly sure that we can make a 1½-in. diameter core out of chamotte, pour it in an 8-in. section of cast steel, and have no penetration of the steel into the core.

MR. BRIGGS: I simply stated that on comparable types of castings, I think the Germans have less cleaning jobs than we do. I did not say there was no cleaning job.

MR. WETTERGREEN: I would like to clear up a point that may be in your minds because of this discussion. In the manufacture of castings such as portrayed in this paper, we sometimes have good luck and produce one of these castings without any of this burned-in material. In other words, the cores blow out with an air hose and we have nice clean steam passages in the foundry and perhaps spend \$80 to \$100 cleaning that casting. Then, because of the engineering requirements for this job, several hundred hours are spent cleaning those steam passages after rough machining. That is where the saving would come from a process like this, in the subsequent cleaning after the foundry, which is called for because of the severe engineering requirements. So that we are not primarily interested here in the removal of this material to save foundry labor. We are interested in the removal of minute embedded sand grains in the surface of the steam passages, the removal of which is a very costly process because of the complicated coring. This, of course, is important. The other is many times more important to us.

CO-CHAIRMAN ZANG: That is another angle on the question of cleaning these castings. Any small adhering sand in those passages could get into the mechanism and ruin it, so that your process undoubtedly, in your mind, assures the removal of those small particles and so protect your customer.

MR. WETTERGREEN (Author's Closure): All the evidence acquired by the author indicates that the American steel foundryman has not been able to consistently produce heavy sectioned, complicated steel castings without encountering fused sand and penetrated metal in amounts which are costly to remove, and potentially harmful in applications where severe engineering requirements are involved. Reports of successful materials and methods in European practice are not confirmed by experience with the same materials and methods when used here in a number of shops over a period of years.

Even under the best of controlled conditions results leave considerable room for improvement from the standpoint of optimum surface conditions in complicated cored passages surrounded by heavy sections, such as are normally encountered in turbine castings. Costly hand grinding operations are required to produce a type of surface, free from embedded sand grains, and other surface conditions such as lumps, fusion, fused sand, and penetrated metal.

An electrochemical process, such as was the subject of this experiment, seems to offer possibilities of development in conjunction with good foundry practice. The combination of the two should result in a casting of lower cost, improved quality, and greater value, from an engineering standpoint for the severe service applications which must be supplied in the future.

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\*Steel Founders' Society of America, Cleveland

# A THEORETICAL APPROACH TO THE PROBLEM OF DIMENSIONING RISERS

By

J. B. Caine \*

## ABSTRACT

*The dimensioning of risers has been an art of founding that, in the past, has resisted scientific approach. A mathematical approach that has been checked experimentally for steel is advanced. Such an approach, based on heat evolved, heat dissipated and volume contraction on solidification has possibilities for scientific riser dimensioning of any metal and should result in an appreciable increase in casting yields.*

ONE OF THE last remaining phases of the art of founding is that of risering, a phase of foundry operation that has, in the past, resisted scientific study. Even though many papers have been published on risering they have all been a description of the art, describing in one way or another risering procedures for given castings, procedures that unfortunately will not apply to even similar castings differing only slightly in one or two dimensions.

Another significant proof that risering as described in the literature and as used in present practice is still an art is that with one exception<sup>1</sup> all past thought has not considered risering in quantitative terms. The aim at all times has been to produce sound castings, the exact riser size required to produce solidity being relegated to a subordinate position if considered at all.

In this day of increased competition and increased costs this matter of riser size and maximum yield becomes increasingly important. Consequently any knowledge as to the minimum riser size required to obtain solidity in the casting or section to be fed will pay large dividends. It is surprising what difference in yield a variation of only one-half inch in the diameter, or one inch in the height of a riser will make, a difference easily amounting to 10 per cent in yield. If some method of scientifically proportioning the riser can be established—a method accurate enough to enable the foundryman to proportion his risers to the minimum required for solidity—the increased yield possible can result in a saving in the order of 10 per cent of the cost of the casting.

One mental hazard that has impeded scientific approach to quantitative risering is the infinite variety of castings produced. This hazard is illusionary, for

any casting no matter how complex is composed of a number of units as far as risering is concerned, each unit solidifying as an individual casting. These units are simple shapes, combinations of plates, cylinders, squares and rectangles and an approach to risering even the most complex casting can be made by studying the solidification characteristics of such simple shapes. If the solidification characteristics of these simple shapes are known, the only problem in the approach to risering the most complex casting is that of determining just how to divide the casting into simple shapes, each requiring a riser of definite proportions to feed it.

If the casting is composed of thick and thin sections the problem of breaking down the casting into the proper parts, each to be fed by one riser, is not too difficult. The thin sections will solidify rapidly enough so that each heavy section will require an individual riser and each of the heavy sections with the thinner sections attached to it will constitute an individual casting. If a casting section requiring more than one riser is uniform, or approximately uniform the question of how to break down the casting into a number of individual castings, each requiring one riser is more difficult. It is necessary that the distance each riser will feed be known before the number of risers required, and consequently the amount of metal to be fed by each riser, can be established.

## Increased Yield Results

Nevertheless, there are enough castings that can either be fed with one riser, or can with our present knowledge be safely broken down into simpler parts for risering that a theoretical approach to dimensioning risers for such simple shapes and castings that can be fed with one riser will pay immediate dividends in increased yield.

In the past every foundryman has had his own rules as to how to proportion the riser. The trial and error method of risering has been based on section thickness, depth of section and rarely on volumes or weight of the casting or section to be fed. It was generally agreed, at least in this country, for steel, that the riser section should be larger than the casting section, but

\* Metallurgist, Sawbrook Steel Castings Co., Lockland, Ohio.

there was no agreement as to how much larger it should be. Another fairly general rule was that the riser height should exceed the depth of the casting or section to be fed, but this rule was broken many times on very deep sections. Another general rule was that the height of the riser be one and one-half times its section. None of these rules with the possible exception of the first had any experimental foundation, nor were they observed at all times, even by their strongest exponents. Other than general rules such as these, riser dimensioning was an art based solely on experience, plus a lot of plain guessing.

### Shrinkage Characteristic of Metals

However, the shrinkage characteristics of metals during solidification are very precise and fortunately do not vary appreciably with composition within the range of composition encountered for that particular metal. This should make the scientific approach relatively simple and amenable to precise mathematical study. All that need be taken into consideration is the rate of heat evolution by the metal until it is completely solidified, and the rate of heat transfer across the sand-metal interface, as the two controlling factors in the relative rate to complete solidification of the casting and riser.

From the standpoint of scientific risering for maximum riser efficiency the problem can be simplified. What is necessary is a method of determining the smallest riser that will deliver the necessary volume of metal to the casting as it is undergoing solidification contraction and one that will remain liquid until the casting or section to be fed has completely solidified. It is axiomatic that no feeding can occur after solidification is complete.

Another simplification can be made. The problem of solidification shrinkage must be divided into two parts, for each part, although dependent on the same physical laws, must be studied and controlled as a separate entity. One part, that under discussion in this paper, concerns itself with shrinkage at, or immediately adjacent to the riser. These shrinkage defects are due to an improperly proportioned riser that is either deficient in volume or solidifying at a rate faster than the casting or section it is to feed.

Shrinkage defects, whether definite voids or metal of low density in areas relatively remote from the riser are an entirely different phase of this problem. There is a limit to the distance a riser will feed, this distance depending on the section shape as well as section size for any given metal. Increasing the size of the riser has very little if any effect on this type of shrinkage, away from the riser. Control of this type of shrinkage must be based on directional solidification, chilling, or the use of more risers, not in the size or shape of riser relatively remote from the areas showing shrinkage defects.

Therefore, in the study of riser proportions and riser efficiency, shrinkage defects in areas remote from the riser can be neglected and if the junction of the riser and casting is sound, the casting can be considered sound as far as that riser is concerned. This phase of the problem then resolves itself into one

concerned with dimensioning the riser to insure solidity within the feeding area of the riser and to accomplish this with the minimum riser volume.

Only two factors are required for the study of this phase of the problem of solidification shrinkage: volume, and a factor denoting the relative rate of solidification of the casting and riser. Some limits are also known. If the solidification rate of the riser is infinitely small in relation to that of the casting, the riser volume required would be equivalent to the contraction in volume during solidification. In other words, if an electric arc be used to keep the riser liquid indefinitely, a riser whose volume is equal to the amount of contraction during solidification will drain out and fill all the voids opening up within the feeding range of the riser, as the casting or section freezes. In steel this riser would be equal to one 3.0 per cent of the casting volume.

This value is much smaller than required in production and the reason is simply that in production the riser is solidifying at the same time the casting is solidifying. For efficient scientific risering the problem is to determine the smallest riser that will solidify at a slower rate than the casting it is to feed, delivering sufficient feed metal to fill all the voids opening up within the feeding range of the riser until the casting is completely solidified.

Another limit has been established. It has been proven for steel<sup>2</sup> that a riser whose section is equal to the casting section will not feed efficiently, in fact will not feed at all, regardless of its height, unless some type of insulation is placed around the riser to slow up its freezing rate, or some addition be made to the riser to accomplish the same purpose. This limit should also apply to any metal whose solidification shrinkage is of such a nature as to form definite voids on freezing. As two equal sections must solidify at the same rate, it is concluded if the riser freezes at the same rate as the casting, the riser volume required is infinity.

### Freezing Rate of Riser

Both limits are now established. If the riser freezes infinitely slowly in relation to the casting, the riser volume required is equal to the volume contraction on solidification. If the riser freezes at the same rate as the casting, the riser volume required is infinity. The question now is to establish the shape of the curve designating this relation between these two extremes. The freezing times of the casting and riser between these two extremes must be in relation to the amount of heat evolved by the metal as it cools from the pouring temperature through the solidification range, and the amount of heat dissipated into the sand in contact with the casting and riser.

Unfortunately the constants required to establish these values, specific heat, latent heat of fusion, and especially the quantitative rate of heat flow across the sand-metal interface, are not known and a direct solution of this problem is not possible at present. Fortunately, from the standpoint of riser dimensioning and riser efficiency, direct numerical values are not necessary. If the riser and casting are in contact with



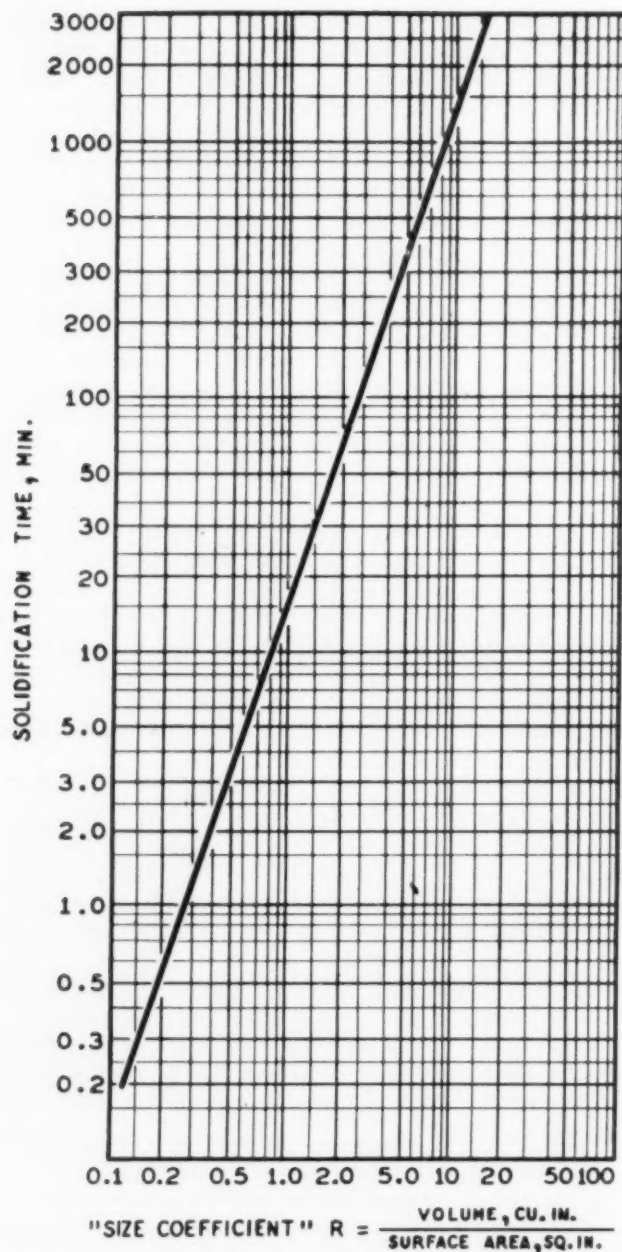


Fig. 1—Logarithmic curve giving solidification time of castings of various sizes (after Chworinoff).

the same sand, the relative freezing rate of both casting and riser to complete solidification is all that is necessary.

Chworinoff<sup>3</sup> has shown that the time for complete solidification of steel in contact with sand molds is proportional to a relation between the volume and surface area of the casting. If the solidification time of castings weighing from a few pounds to 65 tons is plotted logarithmically against a value he calls "Size Coefficient" a straight line can be drawn through the points representing these castings. This curve gives the solidification time of any casting regardless of size and shape provided that the metal is in contact with sand of sufficient thickness that the sand will not be saturated with heat until solidification is complete. Chworinoff's "Size Coefficient" is simply a relation

between volume, the factor determining the amount of heat dissipation into the sand in contact with the casting, as follows:

$$\text{"Size Coefficient" } R = \frac{\text{Volume mm}^3}{\text{Surface Area mm}^2}$$

Chworinoff's results after converting to English units are shown in Fig. 1. Metals other than steel will result in a line with a different position than that for steel shown in Fig. 1, but there is no reason to doubt that the same relation between solidification time and a relation between surface area and volume would hold true for any metal.

As a riser is a casting differing only in shape from the casting to be fed and is in contact with the same sand, the relative solidification rate of the casting and riser is dependent on the relation between the surface area and volume of the casting and riser. From the standpoint of scientific riser dimensioning, the riser should be proportioned for maximum riser efficiency so that the minimum size riser freezes at a slower rate than the casting or section it is to feed. At the same time it should contain sufficient volume of metal to satisfy the feed requirements of the casting.

It is therefore possible to neglect entirely the unknown constants, specific heat, latent heat of fusion and thermal diffusivity and solve the problems of scientific riser dimensioning with relations. Only two relations are required, one for relative volume, one for relative freezing time to complete solidification, as follows:

$$\text{Volume} = \frac{\text{Volume of Riser as poured}}{\text{Volume of Casting}} \dots\dots\dots(1)$$

$$\frac{\text{Relative Freezing Time to Complete Solidification}}{\text{Volume of Riser}} = \frac{\text{Surface Area of Casting}}{\text{Volume of Casting}} \dots\dots\dots(2)$$

Again, as discussed previously, Eq. (1) for volume approaches infinity as Eq. (2) for relative freezing time approaches unity. Equation (1) approaches the amount of contraction on solidification when Eq. (2) approaches infinity. Such a relation indicates a hyperbolic function most simply expressed by the equation:

$$x = \frac{a}{y - b} + c \dots\dots\dots(3)$$

whose asymptotes for steel are unity relative freezing time and 0.03 relative volume. Therefore Eq. (3) can be written for steel as:

$$x = \frac{0.1}{y - 0.03} + 1.0 \dots\dots\dots(4)$$

where  $y$  = volume expressed as a fraction such as those resulting from the solution of Eq. (1),  $x$  = relative freezing time in units resulting from Eq. (2). Constant  $a$  has been assigned the value 0.1 for steel so that the results of Eq. (4) will be in the same order of magnitude as those resulting from Eq. (1) and (2). Constant  $b$  is the relative contraction of steel on solidification. Constant  $c$  is the measure of any change in the relative freezing rate of the casting and riser. If both casting and riser are in contact with sand and are dissipating heat into the sand at the same rate constant  $c$  is 1.0. Mathematically the inclusion of

this constant merely shifts one asymptote from  $x = 0$  to  $x = 1.0$ .

The solution of Eq. (4) for steel results in a curve represented by the solid line in Fig. 2. As an increase

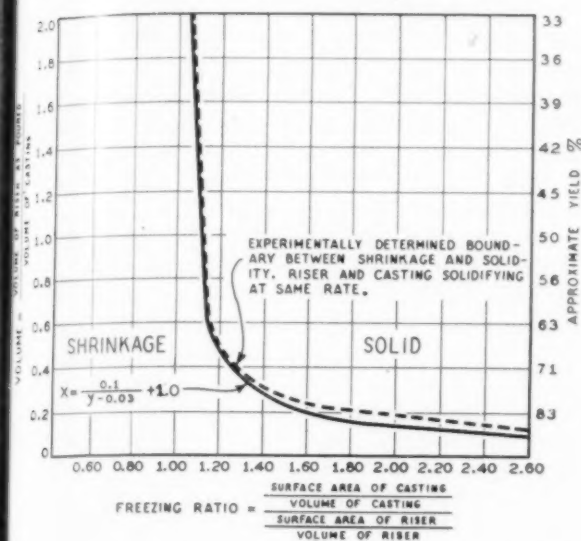


Fig. 2—See text.

in riser size results in an increase in both the  $x$  (relative freezing time) and  $y$  (volume) values, those risers corresponding to points above and to the right of the curve should have sufficient volume and freeze at a rate slow enough to result in solidity of the casting or section they are feeding. Those risers corresponding to points below and to the left of the line should feed improperly due to either insufficient volume or too rapid solidification.

In Fig. 2 the volume ( $y$  ordinate) is plotted vertically as the fraction resulting from Eq. (1). Relative Freezing Rate to Complete Solidification ( $x$  ordinate), hereinafter designated as Freezing Ratio, is plotted horizontally as a proportion resulting from Eq. (2). If a riser is of the same shape and the same size as the casting the volume is 1.0, the freezing ratio 1.0 and there are a number of risers of different shape for each casting that will result in the same proportions. As the size of the riser is increased, both the volume and freezing ratio values increase, but the change does not follow either the  $x$  or  $y$  ordinates, but at an angle to them, the angle changing from almost parallel to the  $y$  ordinate with low freezing ratios to an angle almost parallel to the  $x$  ordinate at high freezing ratios.

To the right of the chart are given the approximate yields when the weight of the gate equals that weight of metal equivalent to the volume of the shrinkage cavity in the solidified riser. In production the actual yield will vary according to the weight of the gate, being lower if a long weighty gate is used, higher if an exceptionally small or no gate is used.

In production risering for maximum riser efficiency the problem is to dimension the riser for any given casting or section so that the point representing the volume and freezing ratio for that particular casting and riser falls just above and to the right of the

boundary line between shrinkage and solidity for the metal to be cast. That point represents the most efficient riser and maximum riser efficiency. The values for volume and freezing ratio given in Fig. 2 cover the range for production castings. It is interesting to note the yields possible even for a metal of high solidification shrinkage such as steel, as shown at the right of Fig. 2.

It is also interesting to note that the dotted line in Fig. 2, the boundary line between shrinkage adjacent to the riser and solidity determined empirically by a number of steel foundries,<sup>4</sup> when no addition is made to the riser, checks the theoretical curve within limits of experimental error. The castings used to determine the empirical curve were relatively simple shapes and no additions were made to the risers to retard the rate of solidification in relation to that of the castings. The implications of the curves shown in Fig. 2 are far reaching. No longer need it be said that riser dimensioning is an art and must remain an art, for here is proof that riser dimensioning of at least steel castings follows relatively simple mathematical laws.

As would be expected in practice the problem is not as simple as the equation indicates, not because of any ambiguity in the equation, but due to ambiguity in the values used for  $x$  in the equation. The  $y$  (volume) values are simple and straightforward, simply the relative volumes of the casting and riser. All difficulties arise in the determination of the  $x$  (Freezing Ratio) values. The surface area component in these values, the component measuring the relative rate of heat extraction can be influenced by many independent variables.

The curve used to check the theoretical curve in Fig. 2 is actually the simplest curve possible, the one denoting the boundary between shrinkage and solidity for relatively simple shapes fed by one blind riser where the total surface area of the casting and riser is in contact with a sand wall of infinite depth, capable of absorbing all the heat generated by the casting and riser until they solidify completely.

#### Restricted Heat Flow

Many production castings are so designed that much of their surface area is not in contact with a sand wall of infinite depth and these areas do not absorb heat at the same rate as the outer surfaces. Cores surrounded on more than one side by metal, re-entrant angles, changes in section all restrict the flow of heat and the surface of the casting at these locations cannot be included in the surface area of the casting at their full value. The correction necessary for these areas of restricted heat flow will have to be determined experimentally and will probably vary with each type of metal cast. This work should not be too difficult, much has been done for steel, and will be repaid many times, not only in increased yield, but also in sounder castings.

If all the riser surface is not in contact with sand and dissipating heat into the sand at the same rate as the casting, this variable must be taken into consideration. This variable can be compensated for in

the basic equation by varying the value assigned to constant  $c$ . This manipulation only shifts the curve in relation to the  $y$  (volume) ordinate. If, for example, a pipe eliminator is used that restricts the freezing rate of the riser, the relatively slower freezing rate of the riser as compared with the same riser in contact with sand would shift the position of the curve to the left and decrease the value of constant  $c$ . Conversely if the riser be exposed to air, dissipates heat at a higher rate than into sand the riser would freeze faster than normal and the value for  $c$ .

The dotted line in Fig. 3 represents the curve experimentally established for steel by another group of

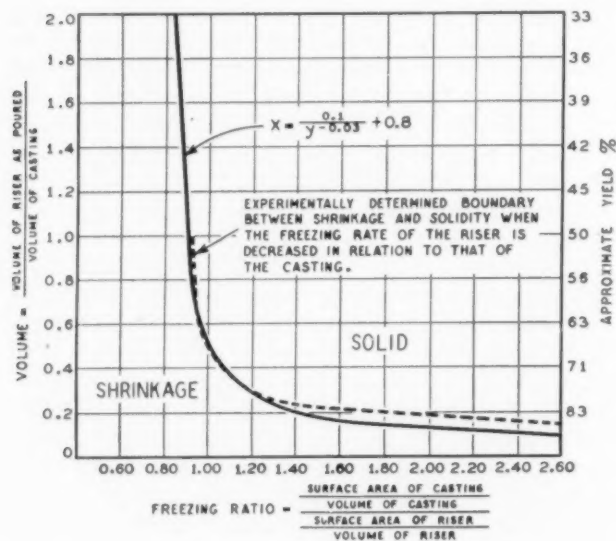


Fig. 3—See text.

steel foundries<sup>4</sup> for risers whose freezing rate is decreased by the use of commercial pipe eliminators. The solid line in Fig. 3 represents the curve resulting from the equation:

$$x = \frac{0.1}{y - 0.03} + 0.8 \dots \dots \dots (5)$$

Again the two curves check within the limits of experimental error. The effect of a material decreasing the relative freezing rate of a riser in relation to the casting can be determined by comparing Fig. 2 and 3. It is interesting to note that these materials are quite useful in increasing riser efficiency for steel with castings of low surface area, resulting in Freezing Ratios below about 1.40. If the Freezing Ratio of the casting riser combination is below 1.10, the use of an efficient pipe eliminator on the riser is imperative.

If it is agreed that the basic risering relation as illustrated in Eq. (4) and (5), Fig. 2 and 3 for steel is valid, it is interesting to speculate at least on the relations for other metals. As this relation is relative, the differences in specific heat, latent heat of fusion and in thermal diffusivity all cancel out and as far as this relation is concerned the only difference that need be taken into consideration for different metals is the difference in the volume contraction on solidification, and possibly the solidification characteristics between the liquidus and solidus temperatures.

As all common cast metals, with the possible exception of the manganese and aluminum bronzes show less solidification shrinkage than steel,\* constant  $b$  in the fundamental equation would be less than the value 0.03 for steel. If this is the only change made the shape and position of the curve does not change materially, the only difference being that the curve approaches the  $x$  (Freezing Ratio) ordinate at a smaller value than that for steel. If these curves be used to proportion risers for other metals, ridiculously large risers result, in most cases, risers the same size as those required for steel. Too many millions of castings have been made sound in metals other than steel with much smaller risers. Therefore constant  $a$  in the numerator of the basic equation must change and this constant may be the measure of the different solidification characteristics of the different metals.

As the author has not been able to obtain any information as to the volume contraction on solidification as defined in this paper of any metal other than steel (constant  $b$ ), much less the proper values for constant  $a$ , two arbitrary solidification shrinkages have been selected, 1.0 and 0.5 per cent. In two instances constant  $a$  is decreased in the same proportion as constant  $b$ , as examples of equations for metals with a lower solidification shrinkage than steel and with solidification characteristics that allow smaller risers than needed for steel. In the third example constant  $b$  is decreased, constant  $a$  increased as compared with the constants for steel, as a hypothetical example of a metal with a relatively low solidification shrinkage but with perhaps a much wider solidification range. The basic equations are as follows.

$$x = \frac{0.033}{y - 0.01} + 1.0 \dots \dots \dots (6)$$

$$x = \frac{0.0167}{y - 0.005} + 1.0 \dots \dots \dots (7)$$

$$x = \frac{0.2}{y - 0.01} + 1.0 \dots \dots \dots (8)$$

The curves resulting from these equations are plotted in Fig. 4 in comparison with the curve for steel taken from Fig. 2, a curve that has been checked experimentally.

The values assigned to constant  $b$ , the volume contraction on solidification as defined in this paper for any metal are simple and straightforward and can be determined easily by experiment. The values for constant  $a$  are unknown at present for any metal other

\* Solidification shrinkage in this paper is that amount of contraction in volume resulting in definite voids of a size large enough to allow the flow of liquid metal into them. The 30 per cent contraction for steel used in this paper is almost all this type of shrinkage. Many of the non-ferrous metals, although showing a greater overall volume contraction during solidification, do not open up definite voids. In foundry parlance they "freeze flat." This type of shrinkage is encountered in steel in areas remote from the riser and is known as "centerline shrinkage." Therefore the values used for solidification shrinkage in the following equations may or may not be the total volume contraction on solidification, but only that shrinkage capable of taking liquid metal from the riser. These values can easily be determined for any metal in any foundry by pouring a simple shape of known volume and measuring the volume of the shrinkage cavity that is formed.



than steel. They can be established experimentally by rising simple castings and determining the boundary line between shrinkage and solidity, especially in the region of the knee of the curve. If the position of the knee of the curve is established, one can feel fairly assured that the position of the rest of the curve is correct. This procedure involving a dozen or so castings is much more efficient than determining the shape as well as the position empirically, as the latter procedure involves casting hundreds of castings.

It may be found experimentally that constant  $a$  for some metals will be larger than that for steel, perhaps due to a longer freezing range, even though these metals show a smaller volume contraction on solidification. Nevertheless it is interesting to speculate at least on the curves. If constants  $a$  and  $b$  are both relatively small as shown by the curves for Eq. (6) and (7) in Fig. 4, the riser size for solidity within the feeding range of the riser is much smaller than that required for steel, about a third of that required for steel at a Freezing Ratio of 1.20. On the other hand, a metal with a low solidification shrinkage, but with such freezing characteristics as to necessitate increasing constant  $a$  will require larger risers than required for steel, unless the surface area of the casting is quite large insuring a high  $x$  (Freezing Ratio) value. A hypothetical example is shown by the curve for Eq. (8) in Fig. 4.

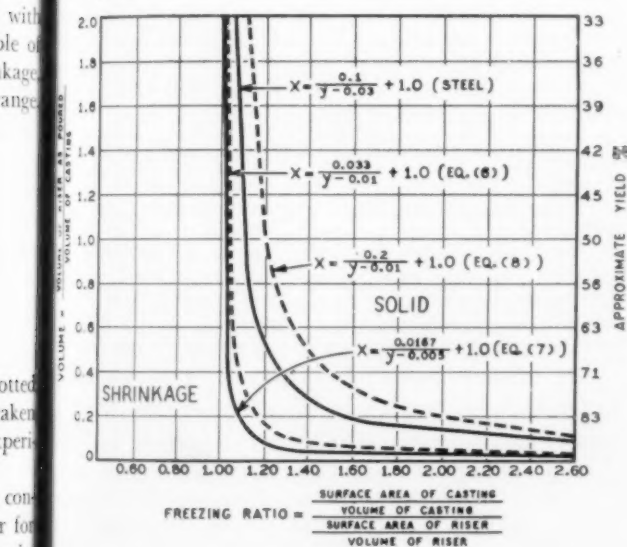


Fig. 4—See text.

Another interesting speculation is on the relative positions of the lines in Fig. 4. As will be noted, the curve established for steel is quite far removed from both ordinates and therefore is relatively insensitive to changes in constant  $c$ . The curves for metals of less solidification shrinkage falling closer to the ordinates, especially the  $y$  ordinate will be more sensitive to changes in  $c$ . As constant  $c$  shifts the position of the curve in relation to the  $y$  (volume) ordinate and as a measure of any difference in the relative freezing rates of the casting and riser, slight variations in constant  $c$ , such as caused by temperature differentials in the mold will be pronounced with metals of low solidification shrinkage and relatively inappreciable

with metals of high solidification shrinkage. Therefore slight variations in temperature differentials such as caused by changes in gating will make an appreciable difference in the feeding efficiency of a riser with a metal of low solidification shrinkage. The same change in gating practice in steel would result in an inappreciable difference in feeding efficiency. This may be a clue to the solution of a number of feeding problems in the nonferrous metals.

### Summary

It has been established both theoretically and experimentally that scientific riser dimensioning depends primarily on two relative relations, volume and relative freezing rate, to complete solidification (Freezing Ratio) as follows:

$$\text{Volume} = \frac{\text{Volume of Riser as Poured}}{\text{Volume of Casting}}$$

$$\frac{\text{Relative Freezing Rate to Complete Solidification (Freezing Ratio)}}{\text{Surface Area of Casting}} = \frac{\text{Volume of Casting}}{\text{Surface Area of Riser}}$$

$$\text{Volume of Riser}$$

These relations can be expressed mathematically as a simple hyperbolic equation:

$$x = \frac{a}{y - b} + c$$

where  $x$  is the Freezing Ratio,  $y$  volume, constants  $a$  and  $b$  are in relation to the volume contraction on solidification of the metal, constant  $c$  is a measure of the difference in relative freezing rate of the riser in relation to the casting due to independent variables. For steel the equation in its simplest form becomes:

$$x = \frac{0.1}{y - 0.03} + 1.0$$

For maximum riser efficiency and yield the riser should be so proportioned that the point representing the volume and freezing ratio values falls just above and to the right of the curve between those areas representing shrinkage and solidity derived from the equations.

Speculations are advanced as to the extension of these relations to metals other than steel.

### Acknowledgments

The writer thanks the Steel Founders Society of America for permission to quote the experimental verification of these hypotheses for steel. He must also acknowledge the suggestions and help of Dr. H. A. Schwartz and Dr. A. J. Smith as well as the help and assistance of the members of his Company.

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## DISCUSSION

Chairman: W. W. MOORE, Burnside Steel Foundry Co., Chicago

Co-Chairman: H. F. TAYLOR, Massachusetts Institute of Technology, Cambridge, Mass.

DR. H. A. SCHWARTZ (written discussion)<sup>1</sup>: J. B. Caine's conclusions are expressed in a graph (Fig. 4) indicating the relation between the ratio of volume of feeder and casting and the ratio of freezing rate of casting and riser which just suffice to produce complete feeding.

The form in which Mr. Caine gives his data has both advantage and disadvantage. The advantage is that, from the dimensions of a given casting and feeder, regardless of any limitation of shape, the adequacy of the feeder can be determined. The disadvantage is that, given only the description of the casting, the calculation of a feeder is a matter of trial and error and there is no guide to the selection of the most efficient feeder form.

It is the purpose of this discussion to supply means for calculating the most efficient feeder if only the description of the casting is given. The calculation assumes the validity of Mr. Caine's conclusions and is applicable only so far as the conclusions are applicable. The work would not have been possible without his demonstration of principles.

The most efficient feeder, of given volume, is that which freezes most slowly, i.e., that of minimum surface area. Elementary geometry teaches that this is a sphere and that the ratio of volume to area is  $\frac{V}{A} = \frac{r}{3}$ , where  $r$  is the radius of the sphere. The sphere could be used where a small (negligible) feeder mouth is possible, especially for side feeding, but has certain practical difficulties.

A cylinder has more practical advantages, especially for top feeding, and, again assuming a negligible feeder mouth, it can be calculated that the cylinder having a minimum surface has a height of  $2r$  and a ratio of volume to area  $\frac{V}{A} = \frac{r}{3}$  as before.

Note incidentally that the constant ratio of  $\frac{V}{A}$  is not a contradiction of the statement that  $\frac{V}{A}$  is greatest for a sphere, for the value of  $r$  for a given volume is greater for the sphere than for the cylinder.

If the cylinder is to be attached to the casting by an entire end, the most efficient form has a height of  $r$  and the same  $\frac{V}{A}$  ratio,  $\frac{r}{3}$ .

Should we wish to have a feeder consisting of a cylinder and hemisphere, as for example the common side feeder with a hemispherical bottom, then for a negligibly small feeder mouth the best form has the cylindrical portion of height  $r$  and the familiar  $\frac{V}{A} = \frac{r}{3}$  ratio. If, however, such a feeder is to be attached to the casting by its entire bottom, the height of the cylindrical portion should be zero and a hemisphere,  $\frac{V}{A} = \frac{r}{3}$  is again indicated.

Presently we will use the volume and areas of these shapes, so these constants may be tabulated below:

TABLE A

Form	Mouth	Exposed Area	Volume
Sphere	Zero	$4\pi r^2$	$\frac{4}{3}\pi r^3$
Cylinder	Zero	$6\pi r^2$	$2\pi r^3$ (Height = $2r$ )
Cylinder	$\pi r^2$	$3\pi r^2$	$\pi r^3$ (Height = $r$ )
Hemisphere	$\pi r^2$	$2\pi r^2$	$\frac{2}{3}\pi r^3$
Hemisphere and Cylinder	Zero	$5\pi r^2$	$\frac{5}{3}\pi r^3$ (Height = $r$ )

These forms are believed to suffice all needs and do represent

<sup>1</sup> Manager of Research, National Malleable and Steel Castings Co., Cleveland.

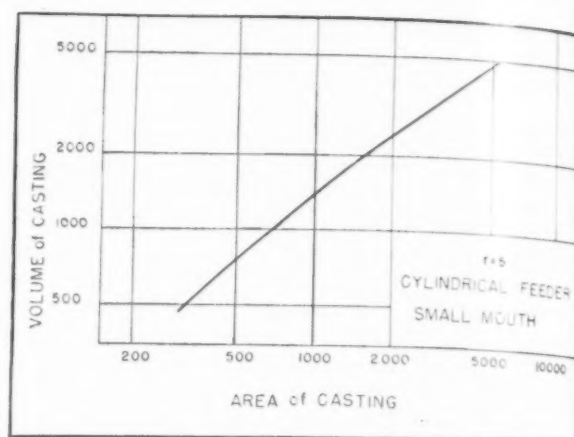


Fig. A—See text for discussion.

the slowest freezing feeds if Chvorinov's principles are acceptable. Caine uses them without any apparent reason developing that they may be inadequate. It must, however, be recognized that there are valid reasons for questioning Chvorinov's conclusions.

Now there are certain other limitations. No matter how slowly the feeder cools or how rapidly the casting cools, the volume of the feeder must be equal to the contraction of the metal in the casting. That is if  $V =$  casting volume and  $V_r =$  riser volume then:

$$V_r \geq 0.03 V_c \dots \dots \dots (1)$$

The minimum value of  $r$  for each of the feeders described in Table A (corresponding to enormously large casting areas  $A_c$ ) would be given by equating the numbers in the last column of the table to  $0.03 V_c$ . It is not implied that in practice the freezing of the casting could be so far accelerated as to realize the conditions for such small feeders. The limits are computed merely as a guide to the scope of later calculations. These limits are:

TABLE B

Form	Mouth	Minimum Value of $r$
Sphere	Zero	$0.193 V_c^{1/3}$
Cylinder	Zero	$0.168 V_c^{1/3}$
Cylinder	$\pi r^2$	$0.212 V_c^{1/3}$
Hemisphere	$\pi r^2$	$0.243 V_c^{1/3}$
Cylinder and Hemisphere	Zero	$0.179 V_c^{1/3}$

The principle of the sphere's relation of  $\frac{V}{A}$  sets a lower limit to the value than  $A_c$ , the exposed area of the casting, can have for a given volume. Were the casting a sphere

$$V_c = \frac{4}{3} \pi r^3$$

$$r^3 = \frac{3 V_c}{4 \pi} = 0.239 V_c$$

$$r = 0.62 V_c^{1/3}$$

$$A_c = 4 \pi r^2$$

$$= 4.85 V_c^{2/3} \dots \dots \dots (2)$$

In any future calculation it is then not necessary to consider any value of  $A_c$  less than  $4.85 V_c^{2/3}$ , a fact which may later have significance.

Further it is plain that for the feeder to operate it must freeze through in a longer time than the casting, hence

$$\frac{V_f}{A_f} < \frac{V_c}{A_c}$$

where  $V$  represents volume,  $A$  exposed surface and the subscripts  $c$  and  $f$  represent the casting and riser respectively.

We have seen from Table A that for all the efficient shapes

$$\frac{V_f}{A_f} = \frac{r}{3}, \text{ hence}$$

$$\frac{V_c}{A_c} < \frac{r}{3} \dots \dots \dots (3)$$

where  $r$  is the radius of the feeder. Here we have at once a minimum value of  $r$  for a value of  $\frac{V_c}{A_c}$  calculated from the castings design. This is not to say that the "r" so calculated is the correct one, only that the acceptable "r" is not less than this value.

The smallest total area to which a circle of radius  $r$  can be attached is  $2\pi r^2$ , i.e., the total area of an indefinitely thin disk whose diameter equals that of the feeder; hence this value is the minimum for a completely attached feeder of radius  $r$ . For fully attached feeders, therefore, the least value of  $A_c$  is

$$A_c = 2\pi r^2 \quad (4)$$

To obtain definite values of "r" we turn to Caine's observations. So far our conclusions are rigidly accurate results based on simple geometry and the calculus of maxima and minima. The only assumption is that freezing time is measured by  $\frac{V}{A}$ , an assumption also made in Caine's interpretation, so this discussion has set no limits to the application of his data.

Note that  $\frac{V}{A}$  measures freezing time but freezing time is not proportional to  $\frac{V}{A}$  but to its square. If there should finally appear limits to his principles, the present discussion has not overcome them and what follows is only applicable so far as Caine's curves are applicable. The writer has no doubt that such limits can be set by temperature gradients between feeder and casting, by the distance through which a feeder can be effective, by the size of feeder mouths and by complexities of form of castings. Within the scope of his experiments, Caine has been able to deal with these vagaries.

Caine suggests an equation as follows:

$$\left( \frac{\frac{A_c}{V_c} - 1}{\frac{A_f}{V_f}} \right) \left( 100 \frac{V_f}{V_c} - 3 \right) = 10 \quad (5)$$

Here  $A$  and  $V$  are the surface area and volume of casting and feeder designated by the subscripts  $c$  and  $f$  respectively.

The reason for using this equation is that Caine found it to fit the facts. It is not, however, improbable for it assumes that a blind feeder (presumably of the same temperature as the casting) will, if it freezes in the same time as the casting, be unable to feed the latter. Also that a feeder must contain at least three per cent of the volume of the casting to feed the latter, no matter how slowly the feeder solidifies, and lastly and perhaps least obviously, that proportion of the feeder volume available for feeding bears some simple inverse relation to freezing time.

On the other hand it may be that some of the constants require further correction and it is certain that the constant 1 in the first parentheses of Eq. (5) will depend upon the relative cooling conditions of feeder and casting.

A little rearranging converts Eq. (5) into

$$\frac{100 A_c V_f r^2}{A_f V_c^2} - \frac{100 V_f}{V_c} - \frac{3 A_c V_f}{A_f V_c} = 7 \quad (6)$$

Remember that  $\frac{V_f}{A_f}$  for all forms of feeder considered is  $\frac{1}{3}$

$$\frac{100r A_c V_f}{3V_c^2} - \frac{100 V_f}{V_c} - \frac{r A_c}{V_c} = 7 \quad (7)$$

Having chosen a feeder form, the appropriate value of  $V_f$  in terms of  $r$  can be inserted in the first and second term and for known values of  $A_c$  and  $V_c$  (exposed surface and volume of casting) we have an expression of the fourth degree in  $r$ , i.e., involving powers including  $r^4$ . Such an expression has four roots, real or imaginary, i.e., four values of  $r$  which satisfy it. Its analytical solution is, however, impracticable. Only positive real roots complying with the restrictions of Eq. (2), (3) and (4) have physical meaning.

A graphic solution is possible for a given value of  $r$  by selecting a series of values of  $V_c$  and solving Eq. (7) for a given value of  $r$ , to obtain  $A_c$  and plotting the result. A single such curve

is shown in Fig. A and a whole family for various values of  $r$  could be computed.

From Eq. (7) it follows that  $A_c$  becomes infinite when, for a cylindrical feeder,

$$V_c = 209.45 r^3 \quad (8)$$

For other forms of feeder the coefficient of  $r^3$  will have other values. It follows that such graphs as that in Fig. A each approach a limiting value of  $V_c$  above which the feeder will not feed the casting no matter how great the area. This writer does not know whether this arises out of the inability of a feeder to feed a great distance or whether it represents a mathematical reason for criticising the form of relationship.

The point is not of great importance for in any event Caine's curves should not be extrapolated far beyond his experimental range.

If complete attachment of the feeder is contemplated, the  $A_c$  of Eq. (7) refers to the exposed surface of casting. To convert the statement so that  $A_c$  shall retain its meaning of actual casting surface  $A_c - \pi r^2$  must be substituted for  $A_c$  and we have

$$\frac{100 r V_f (A_c - \pi r^2)}{3 V_c^2} - \frac{100 V_f}{V_c} - \frac{r (A_c - \pi r^2)}{V_c} = 7 \quad (9)$$

The equation is now of the sixth degree in "r" but can still be solved graphically, as before.

Be it remembered that Caine's reasoning does not include a consideration of heat transfer across the boundary, between feeder and casting. This must exist to some extent for the casting must cool faster than the feeder for the latter to function.

Mr. Caine was influenced in his original thinking by Chvorinov's publications. However, he has set down on paper a curve relating certain measurements of feeders and of castings. It makes no difference if Chvorinov is right or wrong, or indeed if he ever was heard of, Mr. Caine's results expressed relationships which, as found experimentally and which if so verified, are not affected by what one thinks of Chvorinov's conclusions. I assumed that the curve drawn by Mr. Caine represented a fact, no matter by what method that fact was substantiated. His curve must, of necessity, be drawn through an assemblage of points which do not fit in exactly.

We are then confronted when writing an equation for the curve with the question of curve fitting. We are looking for the equation which will reproduce a curve which best separates the good castings from the bad on Mr. Caine's chart. Since there is some vagueness on where the curve ought to be, there will be some vagueness in the equation of that curve, which may or may not be cleared up by theoretical consideration of what volume changes we are talking about. One of the difficulties is that our choice of constants will be largely influenced by how important it is to get a suitable fit far out at the right end of the curve. The observations are mostly near where the curve bends sharply and we have not many at the outer right-hand side. If we want to know more about this with accuracy, the only answer is additional experimentation in the region where we desire more precise knowledge. If this region happens to be one not often attained in practice it may not be worth while to worry about it merely with a laudable ambition to have the curve mean something besides an empirical observation.

Let us remember that we are using mathematical formula to represent observed fact. It is the fact which concerns us and the formula which describes the fact is usable and desirable, no matter whether we have correctly understood why the curve should have that particular shape or not.

In this, as in all other similar problems, we cannot hope ever to get absolute and complete accuracy. It will always be necessary to be content with something that works a large proportion of the time. Certainly we wish to keep on until the times when our predictions are not justified become smaller and smaller, and we would like them to vanish. It is not, however, a criticism of a method in the present circumstance that it sometimes fails. Mr. Caine has made a major contribution when he has devised a method which works in a very large percentage of the time, and which, with a suitable safety factor, can be made to operate as nearly all of the time as we can repeat observations in the foundry.

J. B. CAINE (Written Reply to Dr. Schwartz): Dr. Schwartz has filled in a most glaring blank spot in the problem of



dimensioning risers. The charts as given in the paper cannot be used to determine the dimensions of the risers directly, but can be used only to check the correctness of an arbitrarily selected riser. It is usual, when using this method that two or three tries are necessary before the casting-riser setup results in a volume-freezing ratio relationship giving the most efficient riser.

Although Dr. Schwartz's mathematics may seem highly theoretical, his equations result in curves such as shown in Fig. A, that are very useful in production. With a set of these curves for the various riser sizes and shapes it is possible to determine the most efficient riser directly, knowing only the surface area and volume of the casting. The writer has used these curves in production and has found that the dimensions so determined are accurate to 1 in. If the riser need be dimensioned to accuracies of less than 1 in., one additional calculation is all that is required to bring the point representing the volume-freezing ratio as close to the lines in Figs. 2 and 3 as desired.

However, with our present knowledge, risers accurate to 1 in. are sufficient. Samples must be made if greater accuracy is required, until we learn a great deal more as to the effective surface area of the casting. It should be remembered that allowances in the surface area as determined geometrically must be made for areas surrounded on more than one side by metal. changes in section, any area that is saturated with heat before the casting section freezes.

Since this paper was written, Dr. Schwartz and the writer have attempted to increase the accuracy of the constants used in Eqs. 4 and 5. Dr. Schwartz has determined theoretically that the constant  $a$  should be 0.12 for steel. This change alone gives a better fit with low freezing ratio values, but increases the discrepancy between the theoretical and experimental lines for high freezing ratio values. However, if consideration is given to liquid shrinkage as well as solidification shrinkage and liquid shrinkage be included in constant  $b$ , a better fit at higher freezing ratios can be had than shown in Figs. 2 and 3.

The inclusion of liquid shrinkage into the equations seems logical and may explain some variations in shrinkage due to pouring temperature. Constant  $b$  will then change according to pouring temperature increasing from 0.03 to 0.07 for very high pouring temperatures. With an average pouring temperature constant  $b$  will be about 0.05 and the equation,

$$x = \frac{0.12}{y - 0.05} + 1$$

results in as close agreement with the experimentally determined curve as is possible until more accurate experimental data are available.

It should again be emphasized that the difference between the equation just given and Eq. 4 is so slight that either can be used in production risering. Errors in determining the effective surface area overshadow the difference between the two equations.

CHARLES LOCKE:<sup>2</sup> Many of us who have had access to the Steel Founders Society research report referred to by the author have attempted to make use of the method for determining riser size that it advocates. Assuredly, no substitute is advanced for experience, which heretofore has been the basis for determining the correct riser for a particular casting. A guess still must be made as to what riser dimensions should be used to derive a location on the coordinate paper for comparison with the boundary curve, as explained by the author. Because of this, the method is open to justifiable criticism. Nevertheless, it is still more economical to hunt and try with pencil and paper than with castings.

In addition, nothing can detract from the value of Mr. Caine's work from the standpoint alone of its contribution towards the goal foundrymen want to attain very badly. That goal is to be able to calculate the riser size in terms of the casting's dimensions. No doubt the shortcomings previously mentioned will be overcome by someone with a greater mathematical talent, who it is hoped will present his results in a usable form. My personal attempts to do so have resulted only in supplying an equation which will give the proper riser height for the riser diameter that is judged to be usable. At least we only have to guess that way for the diameter. Trying to go any further, I, too, ended up with the equations to the 6th degree.

If the main premise of Caine's derivation, namely, Chvorinov's principles, is incorrect, then a successful use of the method

with a great many castings in industry must be left to pure chance. It must also be remembered that even a calculating machine such as the heat and mass flow analyzer is fed various guesses until the results check an empirically derived curve.

MR. CAINE: Do you not think that Dr. Schwartz's contribution eliminates the necessity of first arbitrarily selecting a riser and makes it possible to determine the riser dimensions fairly accurately knowing only the surface area and volume of the casting?

MR. LOCKE: Yes, it does help, and please note that I said a "usable" form. I have seen Dr. Schwartz's work and it is most usable; still, in trying it, we found that we could get better yields just by the hunt and try system.

VICTOR PASCHKIS:<sup>3</sup> I want to thank Mr. Caine for a very excellent paper in which he made the mechanism of heat flow in risers very clear also to a non-foundryman. Looking over the paper carefully there are some points which I would like to comment on:

1. Casting and riser exchange heat by conduction and possibly convection. Neglecting this exchange may cause serious errors.

2. The paper gives the impression as if Fig. 2, Eqs. 1 to 3 in short the main part of the paper were based on the correctness of Chvorinov's findings and of Fig. 1 (which is taken from Chvorinov). It will be shown below that the validity of the present paper does not depend on the correctness of Chvorinov's theory. But inasmuch as the latter's paper is quoted repeatedly it may be well to discuss it briefly.

3. Figure 1 of the present paper and Chvorinov's theory imply two facts which are entirely independent from each other:

a. The solidification time for two castings (presumably poured at the same temperature) are the same, if the two castings have the same "size coefficient."

b. The solidification time increases with the square of the size coefficient; e.g.  $R = 0.2$ ; solidification time 0.5.  $R = 2 \times 0.2 = 0.4$ ; solidification time  $2^2 \times 0.5 = 2$ .

It should be noted that in Fig. 1 the line is drawn so thick as to include values 10 per cent apart: in reading for  $R = 8$  one can find a solidification time of 950 or 850, depending on which side of the line one uses for reading.

4. All experiments carried out in this Laboratory as well as theoretical thermal considerations show that Chvorinov's claim are too broad.

a. The concept of the size coefficient holds only for geometrically similar shapes; for example it applies when comparing spheres of different sizes or of cylinders of different diameters etc. In the case of cylinders however the concept is valid only if length and diameter are changed proportionately. Moreover it holds only for "sufficient molds," i.e., molds the outer surface of which does not heat up during the solidification time.

b. Within the validity of the concept of  $R$ , thus restricted the relationship between solidification time and size coefficient as stated by Chvorinov is correct; the solidification time increases with the square of the value of  $R$ .

c. Inasmuch as it is not permissible to speak of a general value of  $R$  independent of shape, it is not surprising to find that the values shown in Fig. 1 are incorrect, as the following examples show:

A sphere, 4.5 in. diam. has a value of  $R = 0.75$ . Its solidification time,\* depending on the temperature is 270-330 sec. against 540 sec (9 min) following Fig. 1. A slab 4.5 in. thick and large enough to have no end effects has a value of  $R = 4$  in. Its solidification time is 695-1105 sec (11.5-18.5 min) against 324 min in Fig. 1.

d. The question might be asked how Chvorinov can claim validity for his findings on the basis of Briggs and Gezelius tests if the above mentioned report\* confirms Briggs tests and disclaims Chvorinov's findings. The answer is obvious from the report.\* Briggs and Gezelius have in their paper presented average solidification times, taken at quite different pouring temperatures. From Figs. 11-20 of the "Study on Solidification of Steel Spheres" it is obvious that such averaging is not permissible. Moreover Briggs and Gezelius, in their bleeding test could never observe directly beyond approximately  $2/3$  solidified thickness (for a total radius of 4.5 in. the last observation shows 3 in. solidified). Hence plotting "complete solidification" as Chvorinov calls for considerable extrapolation of questionable accuracy.

\* Columbia University, New York

\* See Heat Transfer Committee Report 1948, "Study on Solidification of Steel Spheres" by V. Paschkis, Table VI, p. 377.

<sup>2</sup> Armour Research Foundation, Chicago

5. As mentioned before, the present paper does not stand and fall with the correctness of the curve Fig. 1, which is taken from Chvorinov. The essence of the new paper is contained in Fig. 2. In this figure the ratio of "R values" of riser and casting are correlated with the ratio of the volumes of riser and casting. The latter ratio does not appear at all in Chvorinov. Now, assume that Chvorinov would (in Fig. 1 of the present paper) have drawn with the same arrangement of scales the relationship between time and "R value" differently: e.g., it might be a straight line, but with a different slope; or it might be a curve, either convex or concave with respect to the abscissa axis. Any such change would in no way influence the character of the curve Fig. 2.

6. As the Fig. 2 is drawn now it merely presupposes that castings of equal "freezing ratio" solidify in the same time independent of their shape.

7. The curve Fig. 2 and Eq. 3 are by no means a unique solution for the two conditions given on grounds of physical considerations; the two conditions are:

- Relative freezing ratio  $x = 1$ ; volume ratio  $y = \infty$
- Relative freezing ratio  $x = \infty$ ; volume ratio  $y = b$  (amount of contraction on solidification)

Now there is an infinite number of different equations which satisfy these two conditions; e.g.

$$x = \frac{a}{(y-b)^s} + 1.0; \quad x = \frac{a}{(y-b)^s} + 1.0; \quad x = \sqrt{\frac{y-b+1}{y-b}}$$

Mathematically speaking, two points are not sufficient to determine a curve.

Hence Eqs. 3 and 4 are arbitrary and the good matching with the experimental values is an empirical finding and valid only within the range verified by experiments.

8. Now from reference 4 of the paper it appears as if most of the risers which were investigated were of a shape similar to that of the casting. For such instances the relationship of Fig. 2 may hold. But it would be interesting to see experiments where a thin slab is risered with a sphere, or at least with a cylinder with height equal to the diameter.

It is suspected that such a case would not fit Eqs. 1 and 2.

9. Dr. H. A. Schwartz in his written discussion states: "If the cylinder is to be attached to the casting by an entire end the most efficient form has a height of  $r$  and the same  $V/A$  ratio, Eq. 3." This presupposes that heat is flowing through the end of attachment at the same rate as through the perimeter of the cylinder or through the opposite end. It appears to the writer that this assumption contradicts that of the paper of "no heat exchange between parts of different cross section."

J. B. CAINE (Author's Reply to Dr. Paschkis' Discussion): Dr. Paschkis' discussion has put this subject on a firm mathematical foundation, one that as the author I was incapable of doing. I am perfectly willing to stand corrected as to the validity of Chvorinov's relation as long as Dr. Paschkis agrees that there is some relation between volume and freezing ratio and the relative freezing rates of the casting and riser.

However the range of castings that can be risered from Figs. 2 and 3 covers about all castings encountered in production. Contrary to Dr. Paschkis' suspicions as expressed in item 8 of his discussion,  $1\frac{1}{2}$ -in. slabs were risered with cylindrical risers whose height was equal to the diameter. The experimental results checked the curves of Figs. 2 and 3. Infinite slabs can never be risered in steel, for, as stressed in the paper, the feeding range of a riser is limited. Slabs less than  $1\frac{1}{2}$ -in. in section are usually not risered, for they can be cast sufficiently solid for all practical purposes without a riser.

Item 1 of Dr. Paschkis' discussion is due to my not expressing myself clearly. The area of contact between the casting and riser must be taken into consideration. This area of contact is not included in the surface area of either the casting or riser. This assumes that there is no, or at least balanced, heat exchange between the casting and riser. Perhaps later, more precise work must take into consideration a flow of heat between the casting and riser. This can be compensated for by a change in constant  $c$  of Eq. 3. However, our knowledge at present is so elementary that we must learn much more about the whole problem before such preciseness is needed.

E. C. TROY:<sup>6</sup> There are two things that should be made clearer in the paper. Is the calculated yield based on weight of final casting divided by weight of casting and risers?

MR. CAINE: Yes.

MR. TROY: This means that metal losses due to melting, chipping, grinding, etc. averaging about 6 per cent must be deducted to indicate true yield.

MR. CAINE: But that is a separate item for any foundry.

MR. TROY: But our yields in foundries are figured as casting weight divided by metal charged to furnace.

MR. CAINE: True, but with furnace and scrap loss subtracted from the yields given, they are still much higher than those obtained now.

MR. TROY: We know too, that the gates in small castings often take up 20 per cent of the total metal poured. But that is not as important as this other thing. The whole contention of your paper being based on the fact that the freezing rate of the riser for any given area and volume will be in the same ratio as that of the casting, assuming that the heat is being extracted uniformly, both on the riser face and on the casting face.

While this is probably true of simple shapes, few commercial castings have unlimited coolant or sand on all surfaces. Rapidly cooling external corners, slow cooling internal corners, small internal sand cores together with powerful temperature gradients introduced through manner of gating will tremendously complicate any attempts to calculate proper riser size.

Introduction of small radii at internal corners can change a casting that has a shrink into a casting that does not have a shrink with the same riser system.

On the other hand, I want to thank Mr. Caine for his work. It is hard word and he has tried to approach some definite method of reasoning in the selection of riser sizes.

C. W. BRIGGS:<sup>5</sup> In commenting on Mr. Troy's remarks I should like to advise that the records of the Steel Founders Society indicate that there are several dozen different commercial casting designs for which risers have been calculated by the foundries. These foundries have been able to increase their yields from 40 to around 65, or from 50 to 70 per cent and in doing so they are merely following the curves as they have been set forth. It is easy to follow the curves and a few calculations are simple. These castings of which I speak are commercial castings; they are not only simple shapes. Our biggest problem now is to get steel foundrymen to actually go through the calculations, but I will assure you that if foundrymen will take the time and effort to make a few studies along the lines of the paper you will find that the curves and the calculations you have done will be tremendously valuable to you.

MR. LOCKE: As pointed out by Dr. Paschkis, even though the theoretical equation does not satisfy the boundaries used in its derivation, we cannot deny the check of the curve by points calculated from very many actual castings. We just cannot overlook the fact we have experimental verification.

Those of you who have read the Steel Founders' Society Research Report 13 will recall that complications, such as Mr. Troy pointed out, did occur. But when areas of retarded cooling, internal cores, or flanged castings entered the picture an adjustment in calculation had to be and was made.

There is nothing wrong with that method. The greatest scientists use it. If they have something presented to them by experiment and it is not in accordance with a theoretical equation, the equation is altered.

As an illustration consider the work of Dr. Paschkis. In one case with a set of data on steel where the heat analyzer does not check the experimental results, a value for an air gap is fed into the machine. This allows his results to check experimental observations. In another set of data with aluminum, where the heat analyzer without a value for an air gap does check experimental results, the air gap is forgotten.

DR. PASCHKIS: I think you are mistaken regarding the air gap. In all sand castings, the casting of metal against sand, we have consistently found the air gap is not there. In casting against a metal surface, chill, or a metal mold, there is an air gap and it has to be introduced.

MR. CAINE: I think the answer to most of this discussion is the fact that we must remember that we are in the early stages of this problem. Mr. Troy's step corresponds to a slightly advanced stage. Many risering questions are admittedly unanswerable at present, but that should not discourage using these relations on simpler castings, for it is only by use that we can increase our knowledge of this problem.

<sup>5</sup> National Engineering Co., Philadelphia

<sup>6</sup> Steel Founders Society, Cleveland

# A FLUIDITY TEST FOR ALUMINUM CASTING ALLOYS

By

W. E. Sicha\* and R. C. Boehm\*

## ABSTRACT

*Details of a standardized procedure for measuring the relative fluidity or mold filling capacity of aluminum alloys on different lots of the same aluminum alloy have been described. The careful control necessary in performance of the fluidity test may appear formidable but actual experience with the test will demonstrate that the required technique can be established with little difficulty.*

## Introduction

FLUIDITY MEASUREMENTS on aluminum alloys, for determination of mold filling capacity, can provide useful information for foundrymen. The results of fluidity tests may be helpful in diagnosing casting problems and are particularly valuable in alloy development. Essential requirements of fluidity test method are that it should supply data consistent with fluidity ratings based on production casting experience and that it provide reproducible results in duplicate tests.

A fluidity test that generally conforms to these stipulations was described in the paper entitled, "The Measurement of Fluidity of Aluminum Casting Alloys" by L. W. Eastwood and L. W. Kempf.† This paper described the flat spiral fluidity test casting shown in Fig. 1. Considerable use of this test during the intervening period has demonstrated the advantages of minor revisions in the pattern design and more complete definition of molding and pouring practices, particularly for the purpose of gaining improved reproducibility of test results. The material included in this paper describes these details and constitutes an amplification of the earlier paper.

## Test Casting

The test casting pattern consisted of a  $\frac{1}{16}$ -in. thick by  $1\frac{3}{4}$ -in. wide ribbon of an arbitrarily selected spiral shape mounted on the drag face of a wooden match plate. Green sand was used in molding and a  $\frac{3}{8}$ -in. diameter sprue, cut through the cope, was located so as to open into the spiral cavity at about its mid-point.

\* Aluminum Research Laboratories, Aluminum Company of America, Cleveland, Ohio.

† TRANSACTIONS, American Foundrymen's Association, vol. 47, pp. 571-582 (1939).

One  $\frac{3}{16}$ -in. diameter vent was punched through the cope of the mold at each end of the spiral. A pouring basin made as a core and positioned over the sprue was employed to maintain uniformity in the conditions under which molten metal entered the mold. Dimensions of this pouring basin are shown in Fig. 2. Fluidity comparisons were based on the volumes of the flat spirals.

## Pattern Change

It appeared that some of the discrepancies in results of duplicate tests might be caused by variations in flow of the first metal that reached the base of the sprue. In an effort to insure more uniform conditions of flow into the spiral arms, a  $1\frac{3}{4}$ -in. x  $1\frac{1}{2}$ -in. x  $\frac{7}{8}$ -in. deep well was provided in the drag and directly under the sprue, as shown in Fig. 3. Tests demonstrated that addition of the well at the base of the sprue was effective in improving the consistency of fluidity data. Incorporation of this pattern change necessitated removal of the spiral arms at the edges of the block of

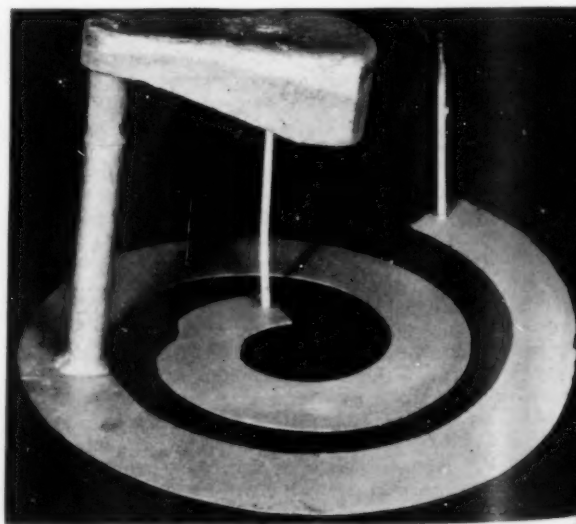


Fig. 1—Flat spiral test casting of original design for determining fluidity of aluminum casting alloys.





Need for standardization of molding material and procedures for production of fluidity test molds were emphasized in the earlier paper on this subject. A more detailed discussion of practices adopted to mini-

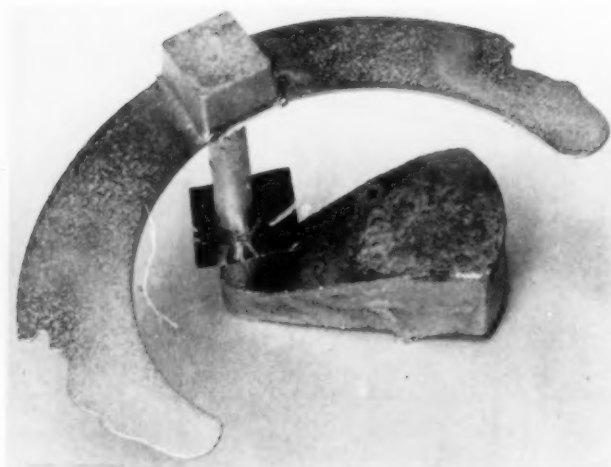
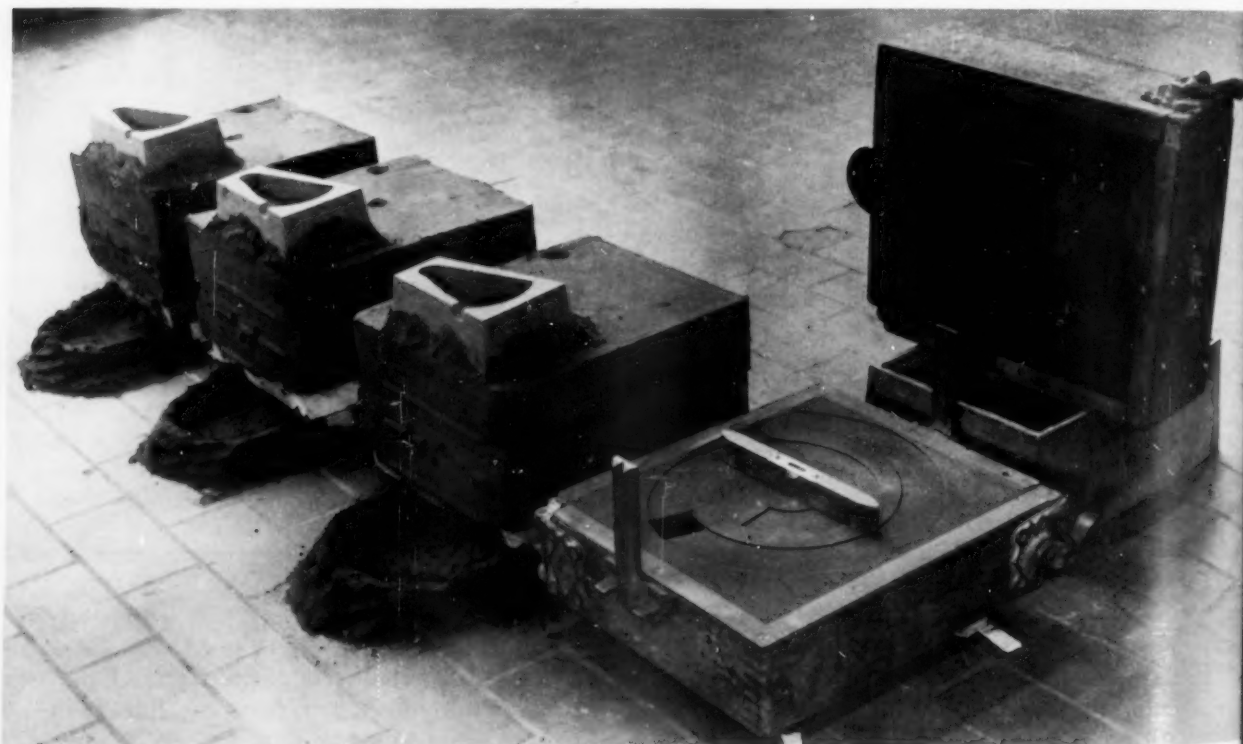


Fig. 3—Flat spiral fluidity test casting of modified design inverted to show the well at base of sprue.

mize the effects of variables in mold preparation probably would be of interest to those contemplating use of this test.

The same base molding sand or sand mix, with a controlled moisture content, should be employed constantly in order to obtain comparable test results. Minor variations in the physical properties of the particular molding sand have no significant effect on fluidity values. However, marked differences in molding sand properties can influence fluidity data materially.

Fig. 4—Fluidity spiral molds in position for pouring in background and a mold drag being leveled in foreground.



A snap flask or tight metal flasks may be used. Convenient dimensions for the flask are 12 in. x 14 in. with a 5-in. deep cope and a 4-in. deep drag.

Hand ramming should be limited to packing the sand adjacent to the flask walls. Squeezing the molds at a fixed pressure is a desirable procedure for attainment of consistent mold hardness and permeability.

The sprue can be formed most readily, as a properly located and uniformly cylindrical channel, by means of a tube or rod suitably placed in the flask cope before filling with sand and squeezing the cope. This sprue form can be held in position at the bottom by a pin mounted on the match plate and at the top by allowing it to extend through a properly located and slightly over-size hole in the cope squeeze board.

#### Preparation of Mold for Pouring

An important detail in mold preparation is the necessity for leveling. This can be accomplished readily by inserting small wedges where needed under the bottom board cleats while checking with a small spirit level placed on the open drag parting surface. The leveling operation is illustrated in Fig. 4.

Institution of the control details that have been described as well as those pertaining to melting and pouring the metal still permitted an undesirable degree of scatter in duplicate test results. Actual pouring of the molds was found to be too subject to variation even with the control exercised by use of the pouring basin core on top of the mold. The direction from which metal entered the pouring basin, the exact location at which the stream of metal encountered the pouring basin, and the rate of pouring influenced test results.

Efforts to minimize the effect of these factors resulted in adoption of the practice of inserting a piece of

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aluminum foil over the sprue opening on top of the mold and under the pouring basin core. An approximately 2 in. square piece of 0.005-in. thick 2S-O foil served as a temporary dam which was effective in reducing the influence of variations in pouring technique.

The core pouring basin must be placed carefully on top of the mold to insure exact alignment of the portion of the sprue in the core with that in the mold. Inscribe a line in the green sand around the properly aligned core affords a means for relocating the core following placement of a piece of aluminum foil over the sprue and on top of the mold. Molding sand should be banked around the pouring basin core after it is replaced to prevent run-outs and to aid in holding the core in position. A group of molds ready to be poured is shown in Fig. 4.

### Metal Handling

The importance of precise control of the metal pouring temperature was stressed in the previous paper and further comment on this point is in order because it cannot be overemphasized. Deviation from the standard pouring temperature by more than 5° F can alter spiral arm volumes much more than relatively large departures from other standard practices.

Pyrometric equipment must be accurately calibrated and should be checked for adjustment just before it is to be used. Beaded chromel-alumel thermocouples mounted in small diameter steel protection tubes or "baby" pyrod couples have proved to be satisfactory for determining melt temperatures. Mounted or stationary pyrometers are considered necessary in this work because adjustment can be maintained much more readily than in production types of portable equipment. Two couples should be available and connected through a selector switch to the pyrometer.

A crucible with a capacity of about 20 lb. of aluminum alloys is suitable because it can be handled by one man for pouring. Employment of charges that practically fill the crucible is advantageous. Adjustment of the melt temperature to that desired for pouring should be performed in the melting furnace.

The rate of heating the alloy in the crucible should be retarded as the pouring temperature is approached. Through brief and repeated operation of the furnace burners, the molten metal can be heated gradually to the pouring temperature. Use of two thermocouples with one placed in the center of the melt and the other against the side of the crucible, assists in observing the magnitude of the heat "head" in the furnace and in controlling operation of the furnace burners during the latter stages of temperature adjustment. This manipulation is necessary to bring the alloy exactly to the pouring temperature at the moment when crucible and furnace are at practically the same temperature. Attainment of this temperature balance prevents heating or cooling of the alloy as it is poured over the lip of the crucible. Melts prepared in this manner can be poured promptly into several fluidity spiral molds in succession with assurance that the temperature drop during the entire operation will be negligible.

TABLE 1—FLUIDITY OF THE ALUMINUM-5% SILICON ALLOY IN SEVERAL TESTS<sup>1</sup>

Melt No.	Fluidity Spiral Weights, Grams					Average Weights, Grams
	1	2	3	4	5	
Duplicate Melts of Ingot of Aluminum-5% Silicon Alloy						
1	57.5	56.5	54.9	54.5	56.5	56.0
2	64.5	69.9	53.5	62.1	61.2	62.2
Combined Average						59.1
Remelts of Pigged Metal and Castings from Melts 1 and 2						
3	61.2	55.5	60.7	60.8	62.4	60.1
4	50.8	57.9	66.3	60.7	59.2	59.0
Combined Average						59.6
Remelts of Pigged Metal and Castings from Melts 3 and 4						
5	66.0	54.6	58.9	59.9	60.1	59.9
6	59.3	54.5	58.4	62.2	61.3	59.1
Combined Average						59.5
Remelts of Pigged Metal and Castings from Melts Prepared with a Different Heat of Aluminum-5% Silicon Alloy Ingot						
7	59.5	66.2	62.5	58.0	64.6	62.3
8	57.7	59.4	59.1	58.6	51.0	57.3
Combined Average						59.8

<sup>1</sup> A pouring temperature of 1350 F was employed and aluminum foil was inserted across the sprue.

TABLE 2—EFFECT OF GAS CONTENT ON THE FLUIDITY OF THE ALUMINUM-5% SILICON ALLOY<sup>1</sup>

Melt No.	Gas Contents of Melts, <sup>2</sup> % Voids	Fluidity Spiral Weights, Grams			Average Weights, Grams
		1	2	3	
Spiral Castings Poured From Melts with Low Gas Contents <sup>3</sup>					
1	0	66.3	63.2	57.4	62.3
2	0	63.6	53.4	63.6	60.2
3	0	64.9	67.0	62.0	64.6
Grand Average					62.4
Spiral Castings Poured From Melts with Higher Gas Contents <sup>4</sup>					
1A	0.36	...	62.7	61.7	62.2
2A	0.97	60.6	63.6	63.6	62.6
3A	0.72	66.4	58.5	61.9	62.3
Grand Average					62.4

<sup>1</sup> The three melts were prepared with ingot from a single heat of 43 alloy. A pouring temperature of 1320 F was employed and aluminum foil was not placed across the sprue.

<sup>2</sup> The values listed are per cent voids, determined by density measurements on standardized test specimens poured with the fluidity spiral castings.

<sup>3</sup> The melts were fluxed with chlorine to substantially remove hydrogen.

<sup>4</sup> The remaining molten metal in each melt was treated with 0.2% NH<sub>4</sub>Cl to introduce hydrogen.



The melt should be skimmed with a preheated skimmer prior to the last step in temperature adjustment and further skimming should be avoided. A suitable temperature for pouring is 1350 F for aluminum alloys although some other or more than one standard pouring temperature may be adopted. Pouring should be performed at a fairly rapid rate and a small amount of metal should overflow through the slot in the top edge of the core.

### Test Results

The fluidity rating of a particular heat of ingot or a certain aluminum alloy should be based on the volume of the spiral in precision work. However, informative comparisons frequently can be made on the basis of spiral weights. Satisfactory fluidity values for individual lots of metal can be established by pouring from three to five flat spiral castings from each of duplicate melts.

Adherence to standardized practices such as have been described still yields some variation among weights of individual spirals. The order of uniformity of results obtainable is indicated by the values included in Table I. Variation in weights of individual spirals poured from a single melt usually is less than 15 grams. The object in pouring several castings from each melt is to permit averaging of the individual spiral weights. This procedure provides more reliable values for comparison and as is evident in Table I, average weights of ten spirals from separate tests on an aluminum alloy are in close agreement.

Fluxing the heats of molten metal undoubtedly should be avoided in procuring fluidity comparisons on different lots of ingot of the same aluminum alloy which have displayed non-uniform casting characteristics in production use. In fact, fluxing for removal of hydrogen probably can be omitted in practically all fluidity testing because results appear to be uninfluenced by the gas content of a melt. This observation is substantiated by data given in Table 2. Substantially identical fluidity ratings were obtained on melts of an aluminum alloy prepared with ingot from the same heat but which had been treated so as to provide melts with drastically different gas contents. The thin spiral ribbons probably solidify too rapidly to allow formation of gas bubbles in sufficient quantity to influence metal flow and the related spiral weights.

### Summary

Details of a standardized procedure for measuring the relative fluidity or mold filling capacity of aluminum alloys on different lots of the same aluminum alloy have been described. This information supplements that provided previously in an A.F.A. paper on a fluidity test for aluminum casting alloys prepared by L. W. Eastwood and L. W. Kempf in 1939. The careful control necessary in performance of the fluidity test may appear formidable but actual experience with the test will demonstrate that the required technique can be established with little difficulty.

It is hoped that discussion of this fluidity test

method will stimulate interest to such a degree that its use will be extended. More general application of the fluidity test should provide information leading to development of refinements that would further increase its usefulness in evaluating aluminum alloys with respect to this important casting characteristic.

### DISCUSSION

*Chairman:* HIRAM BROWN, Solar Aircraft Co., Des Moines, Iowa

*Co-Chairman:* H. R. YOUNGRANTZ, Apex Smelting Co., Chicago

*WALTER BONSAK (Written Discussion):*<sup>1</sup> The authors' contribution to the measurement of fluidity is noteworthy since a test of this kind is very much needed. It is gratifying to find that with a small change in design this pattern can be made to work as well as shown in the test data. To get such closely checking results on one and the same alloy is really a convincing proof that the test seems to work. We have not had a chance to revise our pattern yet and try the authors new method, but any step towards a "standard" fluidity test is highly welcome. This change looks good and it would be interesting to have many people try it and to have a comparison and a discussion on it a year from now.

*VICTOR PASCHKIS (Written Discussion):*<sup>2</sup> The authors emphasize the importance of pouring temperature for accuracy of fluidity tests. This statement, made for aluminum castings, is in agreement with the report by the writer regarding steel castings.\* The prevalent influence of the pouring temperature on the solidification pattern makes it desirable to have definite information on the limitations of fluidity tests. How does a variation in alloy influence the fluidity as compared with a change in pouring temperature? To what extent are the variations of fluidity (Table I) the result of imperfect temperature control?

Inasmuch as the temperature is so important, is the introductory statement still correct, that "the results of fluidity tests . . . are particularly valuable in alloy development"?

*D. LAVELLE:*<sup>3</sup> What were the results on some other alloys as compared to the alloys described? What was the magnitude of difference found?

*MR. SICHA:* We have some additional information on that point which was obtained subsequent to preparation of the paper. The range of variation can be indicated by comparison of the Al-5% Si (43) alloy with the Al-4.5% Cu (195) alloy. As I recall the figures we had a fluidity value difference of a little less than for these two alloys. I might give a few values for comparison. In the standardized test with No. 43 alloy we had a fluidity value of about 57 grams and for No. 195 alloy the value was about 50 grams. In the case of the Al-6.3% Si-3.5% Cu (319) alloy, the fluidity value was of the order of 85 grams and the Al-7% Cu-3.5% Si (C113) alloy provided a fluidity value of about 82 grams.

### MR. SICHA'S REPLY TO MR. BONSAK

*MR. SICHA (Reply to Mr. Bonsack's Discussion):* We hope that the interest of Mr. Bonsack and others in working with this fluidity test specimen will contribute improvements and revisions that will provide an even higher degree of sensitivity and reproducibility in test results. However, we believe that the test can be very useful in its present stage of development.

*J. G. MEZOFF:*<sup>4</sup> Did the authors check the thickness of the casting to see if there was a variation in thickness of the spiral which might be related to the variation in the weight of the casting from one pour to the next on a given test?

*MR. SICHA:* Yes, that point was checked carefully and in general the spiral arms were practically identical in thickness. As is indicated by your comment, distorted fluidity values would result from variations in mold cavity thickness and the attendant non-uniformity in spiral arm thickness. It was because of such factors as this that I stressed so heavily the importance of control of molding procedures.

<sup>1</sup> Apex Smelting Co., Cleveland

<sup>2</sup> Columbia University, New York

<sup>3</sup> V. Paschkis, "Studies on Solidification of Castings," TRANSACTIONS, A.F.S., Vol. 55, p. 74 (1947).

<sup>4</sup> American Smelting & Refining Co., Barber, N. J.

<sup>5</sup> Saginaw Bay Industries, Inc., Bay City, Mich.

Mr. MEZOFF: I was thinking also of the possibility of introducing error by not closing the mold properly. A little sand in the pouring keeping the cope up a little would be really significant percentage-wise on the thickness of the casting. That is the point I wanted to make.

Mr. SICHA: That undoubtedly is a condition which must be avoided. In line with Mr. Mezoff's suggestion, it is necessary to be extremely careful in the molding and metal handling operations to prevent introduction of that type of variable. Once established, the control is not as formidable as it might sound.

Co-CHAIRMAN YOUNGKRANTZ: Our mechanism for getting uniform flow into the mold was a little different from that of the authors. We mounted a bottom-pour crucible over the top of the pouring basin. Then we did not have to worry too much about what the temperature of the metal in the crucible was. We poured from the crucible into the bottom-pour basin having our metal overheated slightly. Then as it got down to the

pouring temperature in this bottom-pour crucible which was heated, we pulled the plug and let it run. We felt we got better reproducibility through such a mechanism.

Mr. SICHA: Use of a bottom-pour crucible is another means of controlling metal flow which we accomplished by introduction of a sheet of aluminum foil across the sprue. Our experience with bottom pouring has not been particularly satisfactory, and I think anyone who has had experience with bottom pouring has encountered a problem in maintaining a good seal at the plug in the bottom of the crucible. If you have means for using bottom pouring successfully that practice probably could be employed instead of using aluminum foil.

Co-CHAIRMAN YOUNGKRANTZ: We experienced no better reproducibility than that reported by the author.

Mr. SICHA: If anyone is interested in performing experiments with this fluidity spiral, a full-scale drawing that can be used as a template for production of a pattern will be supplied by the authors on request.

# GRAY IRON HARDENABILITY AND ITS RELATION TO AIR QUENCHING OF CASTINGS

By

R. A. Flinn\* and R. J. Ely\*

## ABSTRACT

*This report discusses first, the range of hardenability available in unalloyed and nickel-molybdenum gray iron and secondly, the use of hardenability values in planning the quenching of castings.*

*End-quench hardenability and other test specimens with slower cooling rates demonstrate that gray iron follows with good reproducibility the same relationships found in steels even at retarded cooling rates.*

*The hardening of castings by air quenching to minimize distortion and cracking is then related to these hardenability values and to the surface to volume ratio of the casting.*

*Spot hardening by flame and induction heating followed by air cooling is also correlated with the above data.*

## Introduction

AS GRAY IRON receives increasing attention as a quality material for engineering use, it is inevitable that complete information will be required concerning its engineering properties. Since hardenability is a dominating consideration in the application of a large group of steels, parallel knowledge of the suitability of gray iron for similar service is required.

The purpose of this paper is first, to indicate the range and reproducibility of hardenability that is available in gray iron and secondly, to demonstrate the use of the higher hardenability irons in castings where air quenching is desirable to avoid distortion or cracking obtained during liquid quenching.

### I. Range and Reproducibility of Hardenability

Murphy, Wood, and D'Amico<sup>1</sup> illustrated in 1936 that the same principles govern the hardening of gray iron as for steel by describing an "S" or isothermal transformation curve similar to eutectoid steel. Others<sup>8,9,10</sup> have demonstrated the effect of alloys upon the gray iron "S" curve. Timmons, Crosby, and Herzig<sup>2</sup> have contributed a discussion of the application of the Jominy or end-quench hardenability test to gray iron.

Although a significant amount of gray iron is now hardened, two questions are often asked:

a. What is the reproducibility of hardenability in commercial, controlled cupola irons?

b. What range of hardenability is available in the above irons?

### A. Reproducibility

The end quench test developed by Jominy and Boegehold<sup>3</sup> is now used widely as a standard of hardenability and form an excellent basis of comparison for the alloys discussed in this paper.

Briefly, a 1-in. diameter x 4 in. long bar is heated to the standard quenching temperature for the material, e.g. 1600 F, then removed from the furnace and quenched at one end-face by a standard water jet. This procedure develops a variety of cooling rates and therefore a variation in hardening as shown in Fig. 2.

The hardness at different points may be then related to the effects to be expected in different size castings or at various locations in a large casting.

In these experiments twelve specimens of Class 40 cupola iron obtained during two days production were subjected to the above test, Fig. 2. The analyses of the irons are given in the same graph. The test bars used to insure sound specimens are illustrated in Fig. 1. The Jominy test was conducted by soaking the specimens for 1/2 hr at 1550 F and then water quenching one end in the standard fixture.<sup>3</sup> Hardness was determined on the side of the bar after grinding 0.015 in. deep.

All the specimens hardened to 45 Rockwell "C" hardness minimum to a depth of 1/2 in. In Fig. 2, the allowable variation in an H type steel of similar hardenability has been plotted, to show the comparable range of hardness.<sup>4</sup>

### B. Range

#### 1. Unalloyed Irons

Figures 3 and 4 illustrate the variation in end-quench hardenability for three types of gray iron ranging from a low carbon, low silicon Class 50 iron to a higher carbon and silicon Class 30 iron. While comparatively little difference exists, it is opposite the general effect produced in steel by increasing carbon

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and silicon. This anomaly may be explained as follows:

a. Increasing silicon in gray iron reduces carbon solubility at the austenitizing temperature, a predominating factor in determining hardenability.

b. The higher graphitic carbon associated with increased total carbon and/or silicon reduces overall hardness even if matrix hardness remains constant. Hence the distance along the Jominy bar to a specified Rc hardness (a typical method of evaluating hardenability) would also be reduced.

It should be noted that the hardenability data reported are applicable only, for example, to Class 50 irons of the analysis given. If the class specifications are met by the addition of alloys rather than changing the carbon-silicon balance, a different characteristic hardenability would be obtained.

## 2. Alloyed Irons

A pronounced increase in hardenability is obtained with moderate alloy additions. In Fig. 5, the Class 30 and 40 irons have been maintained for reference and the results for nickel-molybdenum additions to the same and to another base iron have been added.

To make a quantitative comparison of the hardenability of the different compositions, the distance from the quenched face to point at which the hardness falls below a certain value, Rc 45 for example, may be measured and used as an index of hardenability. As Rc 45 (430 BHN) represents the minimum hardness for many specifications, the distance along the end-quench bar to this value has been used as the basis of Table 1. In each case the cooling rate past 1300 F at the same distance has also been tabulated.

The combined carbon content of the quenched structures should be mentioned here briefly. The alloyed irons, although exhibiting 0.52-0.53 per cent combined carbon in the as-cast conditions, rose to 0.73-0.78 per cent, close to eutectoid composition, upon heating to 1600 F. This characteristic has been noted for other alloyed and unalloyed compositions and contributes to both hardenability and reproducibility of hardening.

Since the alloyed materials of Table 1 hardened fully at the slowest cooling rate of the end-quenched bar, the air hardenability bar<sup>5</sup> illustrated in Fig. 6 was used. This consists of a 6-in. diameter x 4-in. long cylinder into which the 1-in. diameter x 7-in. long test bar is inserted for 3 in. This assembly is then heated to the same temperature as used for end-quench tests and then air cooled rather than water quenched. While the Jominy end-quench bar provides a series of cooling rates from 600 F/sec to 4 F/sec, the air hardenability bar cools 4 F/sec to 0.67 F/sec. All rates given are measured at 1300 F. The hardness values of Fig. 8 for rates less than 4.0 F/sec were obtained by using the air hardenability bar.

## II. Air Quenching of Gray Iron

### A. General

In the quenching of steel or gray iron from elevated temperatures to produce hardening, cracking or distortion may occur due to two fundamental causes.

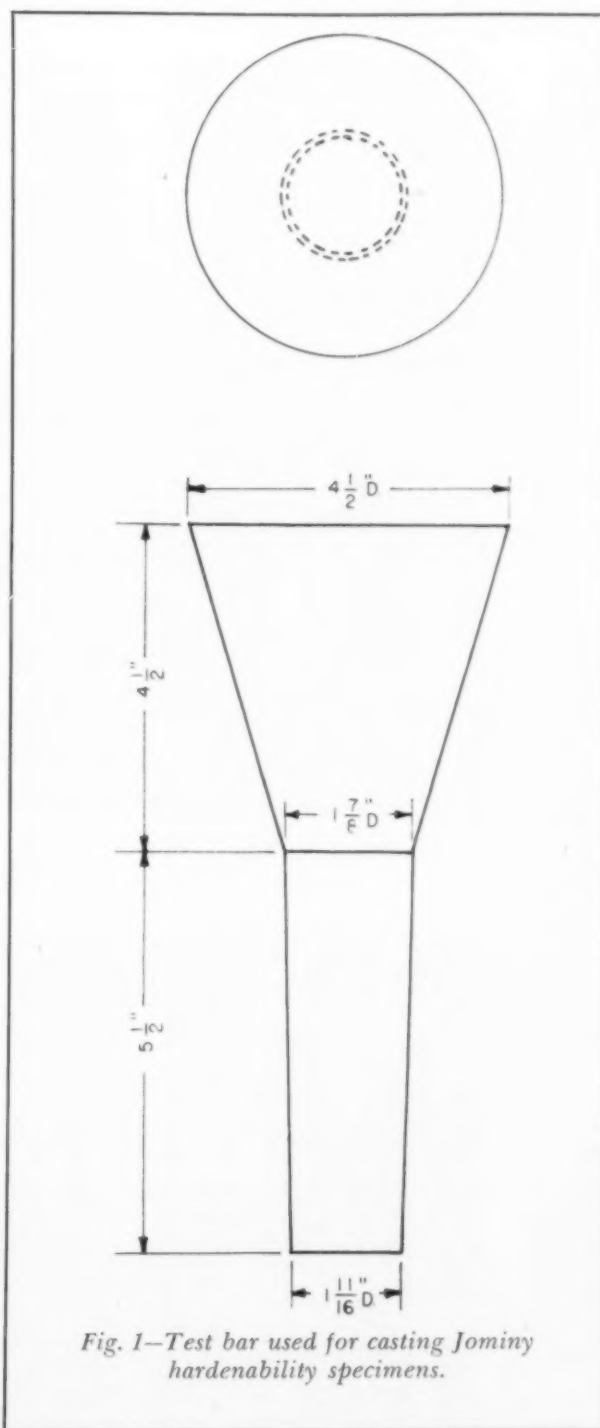


Fig. 1—Test bar used for casting Jominy hardenability specimens.

1. A temperature difference results from the rapid cooling of the surface layers. The thermal expansion of most steels and irons is between 0.0007-0.0012 in. per in. per 100 F. Since each 0.001 in. per in. strain develops 15,000 to 30,000 psi. stress, a 200 to 400 F gradient may cause deformation and in brittle materials, cracking. Even in normally ductile steels, cracking is often encountered because the hardened layer does not possess the ductility of the material in the softened state.

2. Structural changes are accompanied by dimensional changes. The change in length due to the

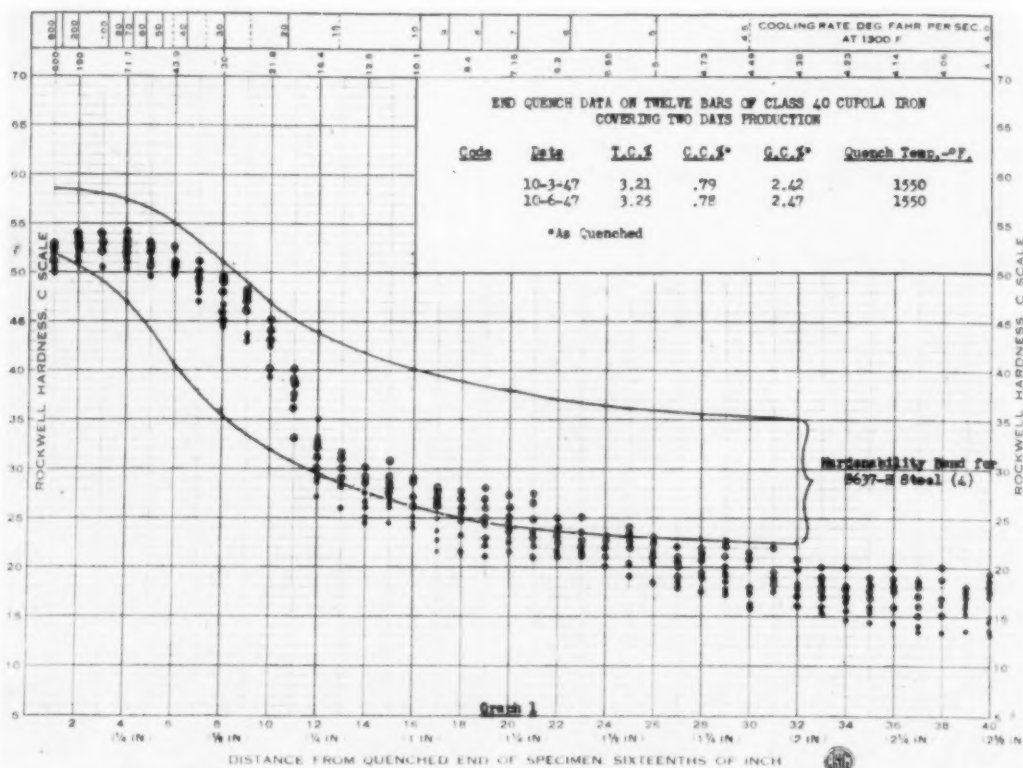


Fig. 2—End-quench data on 12 bars of Class 40 iron from 2-days' production.

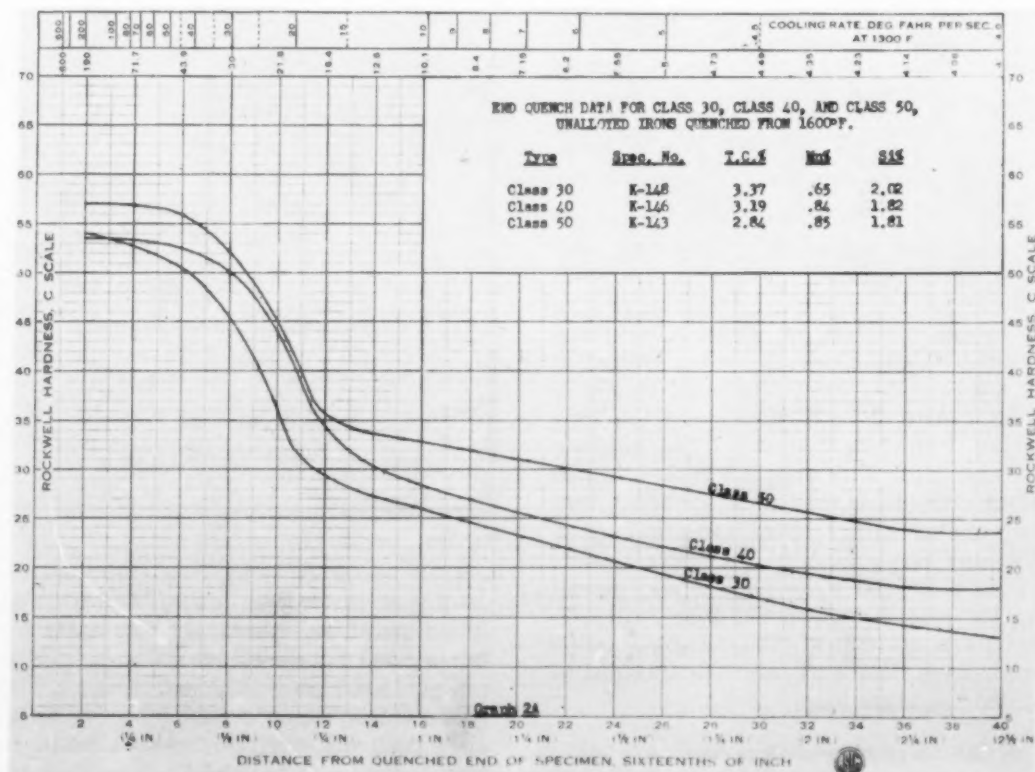


Fig. 3—End-quench data for Class 30, Class 40, and Class 50, unalloyed irons quenched from 1600 F.

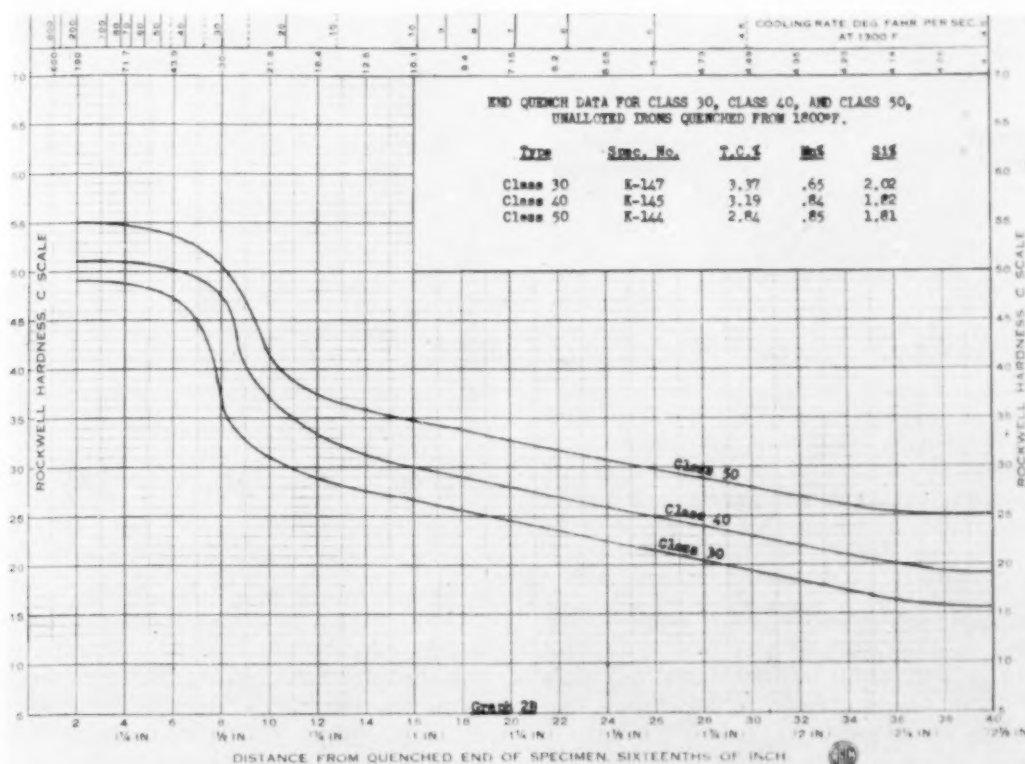


Fig. 4—End-quench data for Class 30, Class 40 and Class 50, unalloyed irons quenched from 1800 F.

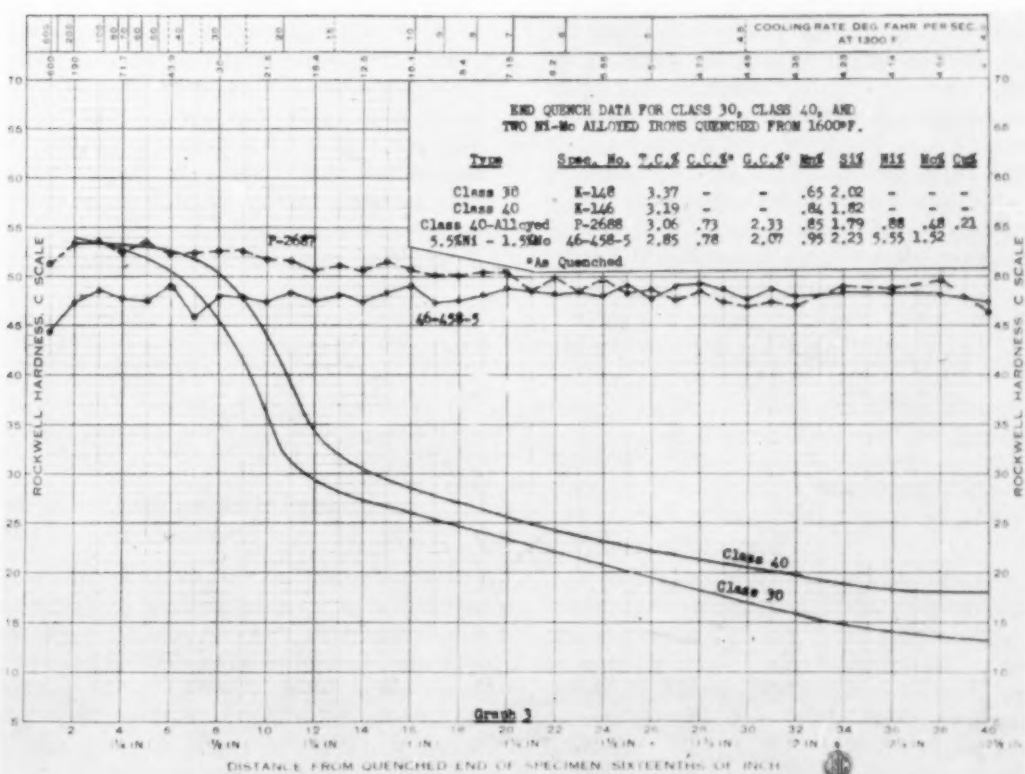


Fig. 5—End-quench data for Class 30, Class 40, and two Ni-Mo alloyed irons quenched from 1600 F.



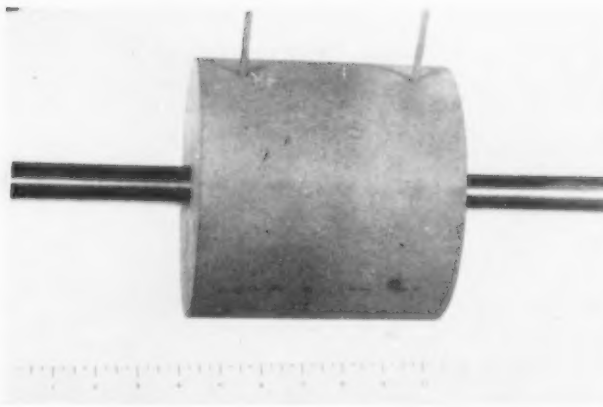


Fig. 6—Test bar for determining air hardenability. Specimens 1 in. in diameter by 7 in. long are inserted in 6-in. diameter by 6-in. long cylinder to a depth of 3 in. The complete assembly is heated to the desired austenitizing temperature and air cooled.

austenite to martensite transformation alone during the hardening of alloy steel or gray iron is approximately 0.010 in. per in.<sup>6</sup> If transformation occurs at different times in separate parts of the casting, the resulting stresses are of larger magnitude than caused by temperature gradients. In general, however, the principal stresses from transformation are in compression.

Since the thermal gradients may be reduced by air quenching, the accompanying stresses may also be lowered or eliminated. This process also decreases the likelihood of transformation at different parts of the casting at different times. Air hardening or air quenching therefore provides less tendency to cracking than liquid quenching and has been used extensively for intricate steel parts such as dies. In the following section the application of this method to gray iron is described. The data of Table 1 and Fig. 8 demonstrate that certain alloyed gray irons would harden

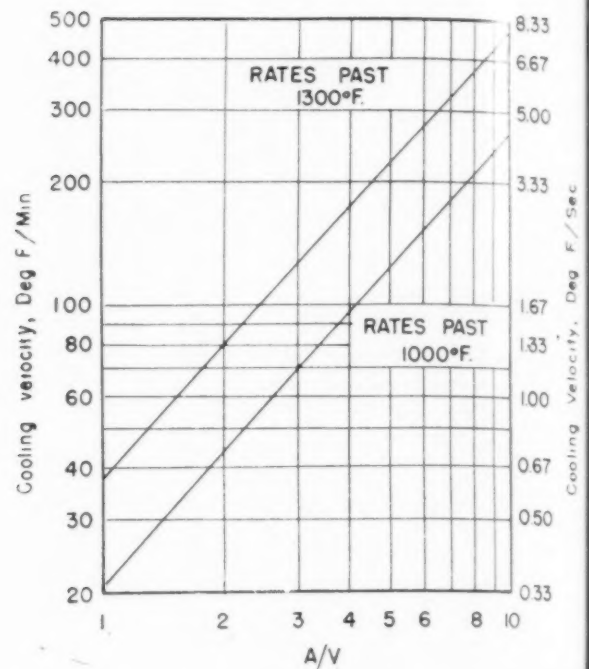


Fig. 7—Center air cooling rates as a function of the area to volume ( $A/V$ ) ratio for regularly shaped masses.<sup>5</sup> (While the cooling rates given in this curve and on the end-quench curves were determined on steels, the use for irons is justified by the similar thermal conductivity and specific heat.)

even when air quenched from the hardening temperature. The minimum cooling rate through 1300 F necessary to harden the 0.88 per cent Ni, 48 per cent Mo iron to above Rc 45 was 4 F per sec while for the 5.55 per cent Ni, 1.5 per cent Mo iron, the rate was slower than 0.67 F/sec. Both of these compositions may be used for air quenching, the lower alloy for lighter sections, the higher alloy for heavy castings. A modified type of air quenching in which only a

TABLE 1—COMPARATIVE HARDENABILITY OF DIFFERENT IRONS\*

Type	T.C. Per Cent	C.C. Per Cent		Mn, Per Cent	Si, Per Cent	Ni, Per Cent	Mo, Per Cent	Cu, Per Cent	Distance** Rate*** To Rc 45 °F/Sec	
		Cast	Quenched							
Class 30 } Unalloyed }	3.37	...	...	0.65	2.02	...	...	...	8.0	30
Class 40 } Unalloyed }	3.19	...	...	0.84	1.82	...	...	...	9.8	23
Class 50 } Unalloyed }	2.84	...	...	0.85	1.81	...	...	...	10.0	22
Class 40 } + Ni and Mo }	3.06	0.53	0.73	0.85	1.79	0.88	0.48	0.21	> 40.0	> 4.0
Induction Furnace } Iron + Ni and Mo }	2.85	0.52	0.78	0.95	2.23	5.55	1.52	...	> 40.0	> 4.0

\* A more complete summary including the data of Timmons, Crosby, and Herzig<sup>2</sup> is given in Table 4 at the end of the paper.

\*\* Distance in sixteenths of an inch from the quenched end of the Jominy bar at which the hardness falls below Rc 45.

\*\*\* These values represent the cooling rate (°F/sec) through 1300 F at the distances defined above. The 1300 F rate was selected because of common usage. The rate in the regions of more rapid transformation is often more significant, but the temperatures of these regions vary depending upon composition.

TABLE 2—ANALYSIS AND MECHANICAL PROPERTIES OF CAST DIE

Heat No.	T.C. Per Cent	Mn Per Cent	P Per Cent	S Per Cent	Si Per Cent	Ni Per Cent	Mo Per Cent
46-497	2.46	0.96	.110	.159	1.81	5.30	1.39
Condition	Elongation						
	C.C. Per Cent	G.C. Per Cent	BHN	Rc	Tensile Psi	Per Cent Total	Per Cent Plastic
As Cast	0.47	1.99	315		64,125	0.40	0.05
Softened 1200 F-15-F.C.	0.49	1.97	241		55,250	0.40	0.11
Hardened 1600 F-4-A.C.	0.64	1.82	444		40,000	0.31	0.07
Tempered 600 F-15-A.C.			418		63,250	0.37	0.04
							Modulus x 10 <sup>6</sup>
							20.7
							20.0
							19.2

portion of the casting is hardened will be discussed later.

### B. Typical Air Quenching Cycle

The cycle for air quenching an entire casting is composed of the following steps:

1. Anneal to relieve stresses and increase dimensional stability for machining operations; also to soften if necessary. In cases where an excess of alloy is used, the casting may harden at the mold cooling rate.
2. Heat uniformly to 1550 to 1600 F for hardening.
3. Remove from furnace and cool uniformly in air to room temperature.
4. Temper if desired at 400 to 600 F.

As an example, a cast die weighing 232 lb and having a maximum section of 4 in. may be considered. The casting was poured of a 5.5 per cent Ni-1.5 per cent Mo analysis similar to that previously discussed. To soften the casting for machining a tempering treatment at 1200 F for 15 hr was used to provide a Brinell hardness of 240. After machining, the casting was heated to 1600 F, held 4 hr and air cooled producing 444 BHN. All the above hardness values were uniform to  $\pm 15$  BHN throughout the casting as disclosed by sectioning.

The mechanical properties of the material are provided in Table 2.

To air harden smaller castings much less alloy is required. A general method for determining the amounts required is included in the following section.

### C. Estimation of Hardenability Required for Air Quenching

The cooling rate of a casting in air as in other media is related to its surface to volume ratio. For simple shapes this ratio may be easily determined and the cooling rate at 1300 F may be estimated from Fig. 7.<sup>5</sup> As the data presented in Fig. 7 are based on center cooling rates, they represent the slowest rate to be encountered and hence if sufficient hardenability is planned for this point, no difficulty should be encountered in obtaining adequate hardness at other locations. For more complicated shapes, the data of Fig. 7 can only serve as a first approximation of cooling rates. Whether required hardenability is obtained

by experimental determination of cooling rates and selection of minimum alloy necessary, or by over-alloying on an approximation basis, would depend upon the quantities or tonnages involved.

For example, assume that it is desired to air harden a casting  $\frac{1}{2}$  in. diameter x 2 in. long. The surface to volume ratio ( $A/V$ ) of this bar is  $3.53/.392 = 9.0$ . The cooling rate past 1300 F is 6.65 F/sec (Fig. 7). The 0.88 per cent Ni, 0.48 per cent Mo alloy with a hardenability of 50 Rc at this rate (Fig. 8) will be satisfactory while an unalloyed iron would require quenching in a liquid medium.

### D. Spot and Differential Hardening by Air Quenching

Much less alloy is required if only certain locations of the casting are to be hardened. For example, let only one region of the 3-in. Y block shown in Fig. 9 be heated to 1600 F at  $\frac{1}{8}$  in. beneath the surface by an oxyacetylene torch. The cooling rate under these conditions will be far greater than if the entire casting were cooled from the austenitizing temperature as shown in Table 3.

The colder surrounding metal provides a rapid mass quench whose cooling rate is 54 times faster than for the same block uniformly cooled from 1600 F.

Since the cooling rate past 1300 F is 45 F/sec with the spot heating just described, an unalloyed Class 40 iron would be expected just to harden (Fig. 8). Examination of an unalloyed Class 40 iron treated in this manner showed, however, the presence of some traces of pearlite in the hardened zone, indicating that a marginal condition exists. In the 0.88 per cent Ni, 0.48 per cent Mo alloyed iron treated similarly no pearlite was observed.

A like condition occurred in practice during the induction hardening of 14-in. ring casting of the section shown in Fig. 10. Water quenching was required to produce a fully martensitic structure to a depth of  $\frac{1}{8}$  in. in an unalloyed Class 40 iron, while a moderately alloyed iron could be air quenched to full hardness.

The relatively small difference in hardenability of irons quenched from 1600 F and 1800 F indicates that the temperature gradients of similar magnitude existing during induction or flame heating should not have adverse effects upon the expected hardenability.

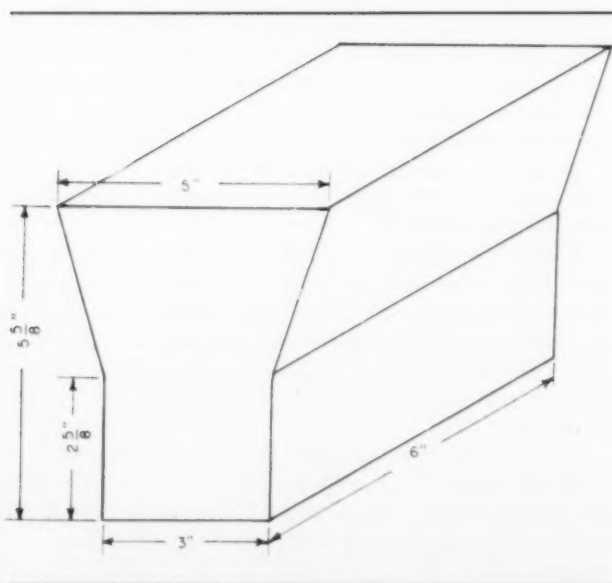
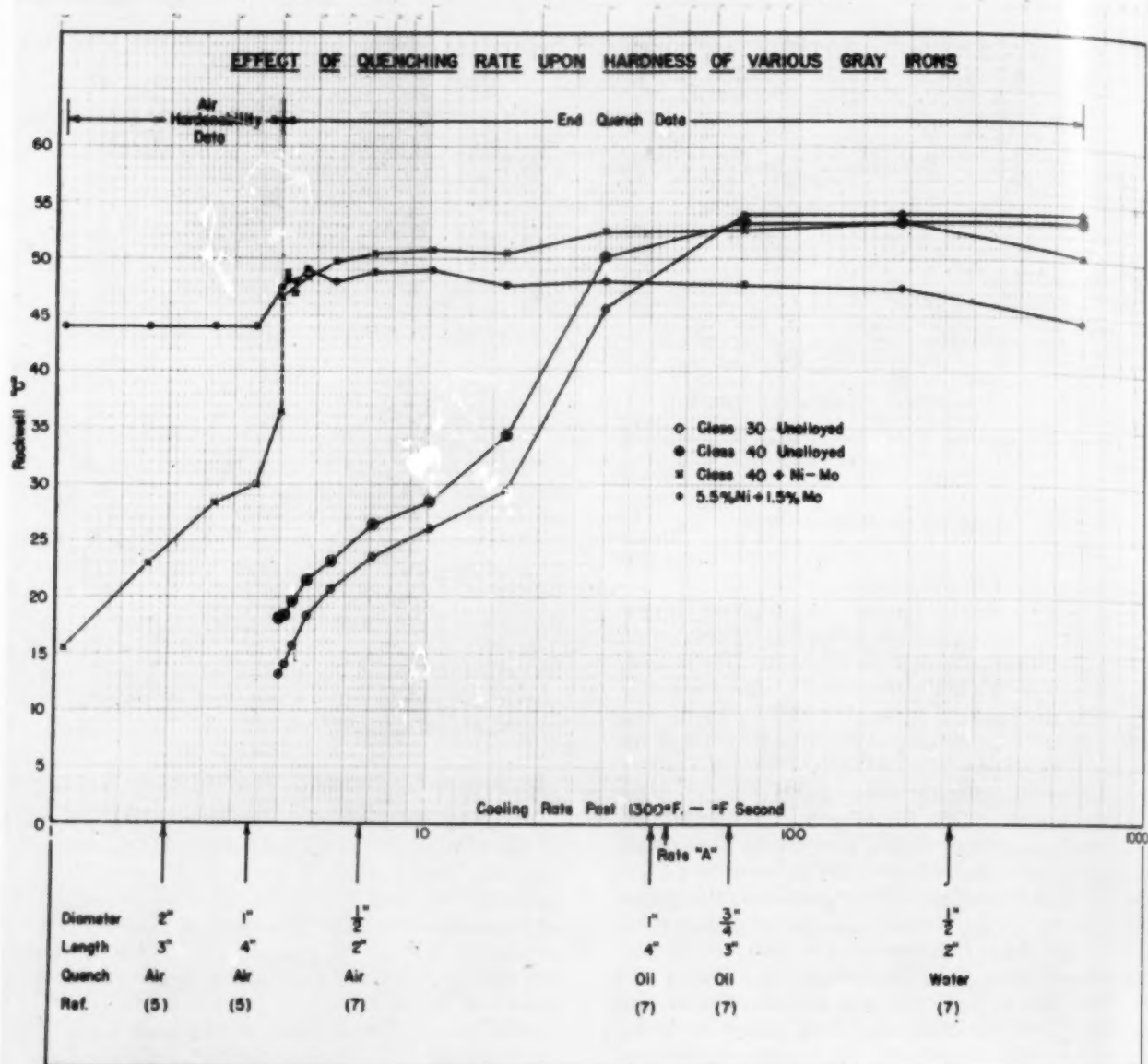


Fig. 8—Effect of quenching rate upon hardness of various gray irons.

TABLE 3—COOLING RATE OF 3-IN. V-BLOCK UNDER VARIOUS CONDITIONS

Condition	Rate at 1300 F. °F/sec
Mold Cooled	0.12
Air Cooled	0.83
Mass Quenched After Spot Hardening	45.0*

\*Rate "A," Fig. 8

Fig. 9—Test bar used for spot hardening. Hardening was performed on bottom face and cooling rate determined 1/8 in. below surface.



TABLE 4—SUMMARY OF HARDENABILITY OF VARIOUS IRONS

Type or Class Iron	T.C. Per Cent	C.C. Per Cent		Mn Per Cent	Si Per Cent	Ni Per Cent	Cr Per Cent	Mo Per Cent	Cu Per Cent	Distance To Rc 45	Rate °F/Sec
		Cast	Quenched								
Plain iron*	3.19	0.69	...	0.76	1.70	...	0.03	0.013	...	7.5	33.0
Class 30 } Unalloyed }	3.37	0.77	...	0.65	2.02	...	...	...	...	8.0	30.0
Class 40 } Unalloyed }	3.19	0.82	...	0.84	1.82	...	...	...	...	9.8	23.0
Class 50 } Unalloyed }	2.84	0.97	...	0.85	1.81	...	...	...	...	10.0	21.8
Mo-B*	3.20	0.58	...	0.64	1.76	Trace	0.005	0.48	...	18.0	8.4
Mo-A*	3.22	0.65	...	0.75	1.73	...	0.03	0.47	...	22.0	6.2
Cr-Mo*	3.21	0.60	...	0.67	2.24	0.06	0.50	0.52	...	29.0	4.6
Ni-Mo*	3.22	0.53	...	0.66	2.02	1.21	0.02	0.52	...	38.0	4.1
Cr-Ni-Mo*	3.36	0.61	...	0.74	1.96	0.52	0.35	0.47	...	> 4.0	< 40
Class 40 } + Ni and Mo }	3.06	0.53	0.73	0.85	1.79	0.88	0.05	0.48	0.21	> 4.0	< 40
Induction Furnace } Iron + Ni and Mo }	2.85	0.52	0.78	0.95	2.23	5.55	...	1.52	...	> 4.0	< 40

\* From Data of Timmons, Crosby and Herzig. These bars soaked 1 hr at 1575 F prior to quenching.

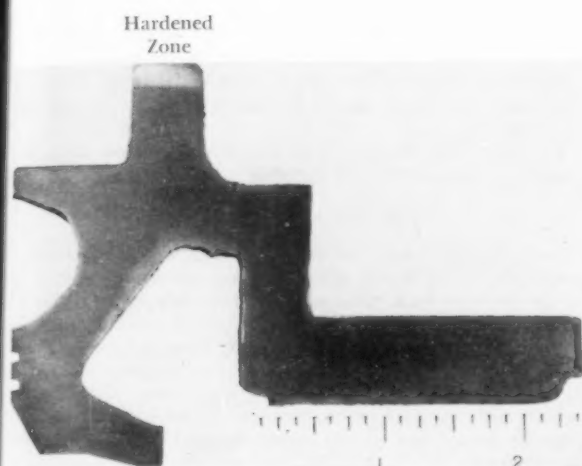


Fig. 10—Section of 14-in. ring casting showing induction hardened zone.

### Conclusions

1. Properly controlled gray irons have both reproducible hardenability and a wide range of hardenability similar to alloy steels.
2. Hardenability may be adjusted by the use of alloys to provide compositions that will harden by air cooling from 1600 F, thereby reducing cracking and distortion.
3. The alloyed compositions may also be used for differential hardening by induction or flame heating followed by air cooling.

### Acknowledgment

The authors are greatly indebted to the entire staff of the Metallurgical Department of the American Brake Shoe Company, especially to Mr. R. H. Schaefer, Chief Metallurgist for many helpful criticisms and suggestions.

They also appreciate greatly the metallographic work of Mr. R. J. Gray and the mechanical testing and drafting by Misses M. A. Moran and S. A. Decker.

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### DISCUSSION

Chairman: J. S. VANICK, International Nickel Company, New York

Co-Chairman: T. E. EAGAN, Cooper Bessemer Corp., Grove City, Pa.

D. J. REESE (Written Discussion):<sup>1</sup> I like the way Flinn and his associates explore the iron-carbon-silicon system and associate their findings with the more explored iron-carbon system. Applying the Jominy end-quench test to cast iron appears to have awaited the initiative of these investigators. As castings are most likely to be irregular in shape with wide variations in sections, drastic quenching procedures are not desirable in most cases. These investigators direct attention to the magnitude of strain and stress when either temperature differentials or structural changes occur.

<sup>1</sup> International Nickel Co., Inc., New York

Although air quenching of alloyed cast iron component parts has found commercial application, for example, in automotive camshafts, the procedures used have usually been developed by trial and error methods for the specific part to be treated. Though these investigators pack a lot of information into a brief paper dealing with the heat treatment of cast iron, sufficient information to void the need for trial and error approaches to the heat treatment of cast iron, we can still hope that these investigators will record their observations on other types of plain and alloyed cast irons at some near future date.

I would like to extend my compliments to Messrs. Flinn and Ely for their contribution to increased knowledge on this engineering material.

#### MR. JOMINY'S DISCUSSION

W. E. JOMINY (*Written Discussion*):<sup>2</sup> This paper on gray iron hardenability is very interesting and shows a thorough understanding of the subject on the part of the authors. Cast iron being a much more complex material than steel requires more careful consideration in applying the laws of hardenability.

For those who are expecting to use hardenability tests in connection with oil quenching or air quenching of gray iron castings, it would be well to heed the two points made at the top of page 509 labeled "a" and "b." Under "a" the statement is made that "increasing silicon in gray iron reduces carbon solubility at the austenitizing temperature, a predominating factor in determining hardenability." This is of greatest importance since dissolved carbon increases both the hardenability and the hardness in cast iron. Since rate of cooling of the casting as well as silicon content greatly affects the amount of combined carbon, these two factors will have a profound influence on the hardenability of the iron.

Another factor which will affect the amount of carbon in solution is the amount of time the iron is held at the austenitizing temperature. The longer the iron is held at the austenitizing temperature the more nearly it will approach the equilibrium between silicon, combined carbon and graphitic carbon. Consequently, in applying these principles, it is essential that the rate of cooling of the casting in the mold as well as the length of heating time and temperature of heating time must be carefully considered to obtain uniform results.

The other point referred to previously is "b", "The higher graphitic carbon associated with increased total carbon and/or silicon reduces overall hardness even if matrix hardness remains constant". This is likewise a very important principle to keep in mind. It is probably simpler to consider hardenability from the point of view of microstructure rather than the point of view of hardness. If distance on the end-cooled bar is measured at which the matrix is all martensitic we get the most accurate measure of the hardenability of the iron. We use hardness measurements merely because they represent an easier method of determining at what distance the microstructure changes from all martensite and at what distance it contains 50 per cent martensite, etc.

Since the hardness is affected by the amount of graphite present, it is safer in cases of any misunderstanding to examine the test piece under the microscope. For instance, in Fig. 5 specimen No. 46-458-5 is shown to have a definitely lower hardness at the water cooled end of the standard end-quenched bar than the other irons. This is the most highly alloyed cast iron which also has the lowest graphitic carbon of the group. I wonder whether examination of the microstructure of this iron would not reveal retained austenite as being the cause of this lower hardness.

The writers do not state whether the test bar shown in Fig. 1 is cast in a graphite mold, green sand or core sand. I should like to ask the authors what they would consider the ideal mold material for this test piece.

In Fig. 6 the authors show the 6-in. dia. x 6-in. long test cylinder for determining air hardenability. In the *Metals Handbook* of ASM (1948 edition) this test piece has been standardized so that the hole through the 6-in. cylinder is bored 0.858 in. in diameter and tapped to a depth of 1 in. so that the test piece may be screwed into the cylinder. The threads on the test piece are at a location so that when

screwed in the cylinder the test piece extends 3 in. into the cylinder. The cylinder is made of a 32 per cent nickel iron alloy to give freedom from scale formation but it should be noted that this alloy will not cool at the same rate as a plain carbon cylinder. It would be well to adopt this standard in making tests so that we can duplicate one another's results.

In translating data from the standard hardenability bars to a commercial casting, it will of course be necessary to make some actual comparisons between the casting under consideration and test pieces made from the same iron. As stated previously, these will be very definitely affected by the section size of the casting and in some cases by the location on the casting.

#### MR. NAGLER'S DISCUSSION

C. A. NAGLER (*Written Discussion*):<sup>3</sup> The authors are to be complimented for the fine paper that they have written on the hardenability of gray iron.

The method of casting the Jominy hardenability specimen is very good. It is assumed that the bars were then machined to the standard 1-in. diameter and 4-in. length. Were the bars copper plated prior to the Jominy test?

Did the authors note any graphitization in the cast irons on heating at 1550 F for 1/2 hr? Was there any ferrite formed?

Have the authors given any thought to the following type of hardenability data? Austenitize the Jominy bar at same predetermined temperature and then quench the bar to same predetermined isothermal temperature for a predetermined length of time and then remove the bar from the salt and quench in the standard Jominy fixture. Then determine hardness values for the length of the bar at stations 1/16 in. apart.

The users of gray iron have been well aware of the air hardening characteristics of the unalloyed and alloyed gray cast irons.

To clarify Fig. 8 to me will the authors please further interpret the data contained therein?

#### MR. EAGAN'S DISCUSSION

Co-CHAIRMAN EAGAN (*Written Discussion*): The recommended procedure of hardening gray iron has always consisted of an oil or water quench with a subsequent tempering to the desired hardness. Because of the danger of cracking, only the simplest of shapes could be hardened. The authors have quite cleverly shown that with the proper use of alloys the hardenability can be increased so that a martensitic matrix can be formed with an air quench. Because of the lower temperature gradient during the cooling more complicated shapes can now be hardened.

Flinn has previously demonstrated that by the proper balancing of carbon, silicon, nickel and molybdenum that hardening can be accomplished in the mold. This is the phenomenon that produces acicular iron with such high tensile strength. In this case he adjusted his hardenability by alloying to the point where no martensite is formed. An increase in alloy will of course cause martensite to form and thus harden the piece. By knowing the cooling rate of any given casting through 1300 F and the influence of alloying on the hardenability it is entirely possible to adjust the composition of the metal to produce either the acicular or martensitic structure.

The authors recommend a heat-treating procedure for air hardening alloyed castings. This consists of an anneal, a preheat for hardening, and a temper. This procedure could be modified in many ways to suit individual needs.

I would like to suggest that the heating for quenching be done in two steps, namely a very slow heat to about 1100 F and soak at this temperature until entirely heated through, then either transfer to a furnace at 1550 to 1600 F, or heat rapidly to this temperature in the same furnace. In other words, due to temperature gradient there is just as much chance of cracking a casting during the heating cycle as there is during the cooling cycle. The preheating would help this situation.

#### AUTHORS' REPLY

MR. ELY (*Authors' Written Reply*): We are very grateful to have received the number of thoughtful discussions submitted.

<sup>2</sup> Chrysler Corp., Detroit

<sup>3</sup> Wayne University, Detroit

into the hardenability test from which we obtained much of our basic data. His comment reemphasizing the importance of the carbon-silicon balance in the iron needs no further amplification. Regarding the effect of the original cooling rate, variations will naturally cause some change in combined carbon. If sufficient time for carbon diffusion is allowed at the austenitizing temperature, small variations in initial combined carbon would tend to be leveled out during this phase of the treatment. Metallographic examination of test bar 46-458, which he suggested, was performed and appreciable amounts of retained austenite were observed.

We appreciate Mr. Jominy's fundamental approach from the structural point of view, and we agree wholeheartedly that there is much extra data that can be obtained by this method in addition to the information from the hardness measurements.

The test bar shown in Fig. 1 was cast in a dry sand core. We feel that casting the specimens in a graphite mold, with its increased chilling tendency would change results more than a shift to a green sand mold, for example.

We are pleased to hear that the air hardenability test bar has been standardized. Development work is currently in progress to produce a bar with even slower cooling rates.

Mr. Egan's point concerning the dangers of cracking during

heating is well taken. His suggestion of a two-stage heating cycle will minimize this danger. Whether this is necessary, of course, will depend upon the size and complexity of the casting that is being hardened.

Mr. Reese's point about trial and error methods in planning hardenability is also a good one. It is felt that particularly where there are enough parts involved to warrant it, adequate planning of hardenability so that the minimum necessary alloy content can be used may result in considerable savings in alloy costs.

Mr. Nagler is correct in his assumption that the standard Jominy specimens were machined from the test bar shown in Fig. 1. The bars were not copper plated prior to treatment.

In unalloyed irons graphitization with resultant formation of free ferrite was noted at the air cooled end of the bar. The quenched ends normally had a matrix of high carbon martensite and retained austenite. The addition of carbide stabilizing alloys reduced or eliminated the tendency to graphitize during the heat treatment cycle.

For the present studies, the work was confined to conditions most closely approximating the majority of service applications. However, it is agreed that additional fundamental knowledge would be obtained from isothermal conditioning treatments such as proposed by Mr. Nagler.



# CONSTRUCTION AND OPERATION OF AN OIL-FIRED MALLEABLE IRON HOLDING FURNACE

By

F. Coghlin, Jr.\*

## ABSTRACT

*The paper deals with the design and details of construction and operation of an oil-fired holding furnace used in a malleable iron duplex system. The refractory layout, the oil burner system, the atmosphere control system, operating methods and data, and some brief comparisons with coal firing of the holding furnaces are included. All aspects of the system are covered in a manner which avoids ultimate detail of any point, for each phase of construction and operation of the unit could be broadened into a complete treatise in itself.*

WITH THE MANY DEVELOPMENTS in malleable iron melting processes during the past two decades, the problem of what type unit would serve to greatest advantage in any given foundry cannot be as easily determined as it was when the hand fired camelback air furnace was the mainstay of the malleable industry. During the time straight-line foundry methods were in their swaddling clothes, foundry operators were usually resigned to the fact that the camelback furnace was the logical and only melting medium to be considered, the consequence being that foundries at that time had to be built around the melting unit.

Since then, however, many units, comprising either a single or a combination of furnaces and for use of practically all types of fuel, have been developed that will service essentially any foundry production demand. Thus, the problem of various types of melting equipment to coincide with any foundry application has been answered, but in its place are the problems—what type of melting unit is most suitable for a given foundry and what fuel should be used. The answer to the first question can be found by simply determining the desired capacity of the unit, but the fuel problem entails much more consideration and planning. This relates especially to the malleable iron reverberatory furnace used as a single unit or as a holding furnace in duplexing.

In selecting the most desirable fuel for an air furnace or a holding furnace, many factors must be considered: the initial cost of the fuel burning equipment, the cost of the chosen fuel, the geographic location of the plant and its affect on transportation costs, the economy of the utilization of the fuel, the refining

and oxidizing requirements of the furnace, the size of the furnace, the foundry metal demand and the currently very important point, availability of the fuel. The last item, by reason of acute fuel shortages, is probably a stronger determining factor now than heretofore, because lack of fuel resulting in a foundry shutdown would indeed be false economy if a higher priced fuel could be readily procured to insure constant and uninterrupted foundry operation.

Having gone through a foundry-wide modernization plan, including the installation of three mechanized and one semi-mechanized molding unit, two new sand conditioning units and a new annealing set-up in 1945, the writer's company was faced with the problem of adapting its then present melting system to the requirements of the remodeled foundry at a minimum expenditure.

## Foundry Remodeled

Water jackets had already been installed and other changes had been made on the cupolas to afford longer melting periods for higher output. The big problems were to increase the holding capacity of the air furnaces, to obtain a system that would provide increased and more dependable super-heat of the iron in the holding furnace, to produce greater carbon reduction with greater tonnages but without increasing the hearth area, to replace an antiquated and worn out central coal pulverizing system and to provide a flexible fuel system that would relieve somewhat the ill effects of fuel qualities and shortages.

Inasmuch as a fuel oil heating system, equipped with a large oil storage and super-heating unit had previously been installed in the plant it was decided that an economical approach to the problem, at least from the standpoint of initial cost, would be to install fuel oil burners on the reverberatory furnaces and use the available fuel supply. After determining the size and type burners best suited to the installation, it was further decided that one burner in addition to the indicated number required on the basis of a straight Btu conversion from past coal consumption figures would be provided. The purpose of the extra burner was not to compensate for any expected drop

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in efficiency for an increase was expected and obtained. The extra burner was included to furnish more accurate control and economical operation of the furnaces through the elimination of the need to ever force the burners.

As a precaution against fuel oil shortages, the burners were mounted on a plate and piped up in a manner which facilitates rapid removal of the oil burners and re-installation of the pulverized coal burners.

This paper contains an account of our experience with the oil fired units, and is based on a compilation of notes on construction, operating data and observations since the installation of the system. Although it contains a list of the advantages we have gained since changing to fuel oil, it is not the intention of the writer to go into detail on the relative merits of either fuel.

On the basis of statistics it is quite evident that very few malleable foundries are using oil for either straight air furnace or duplex melting. Reports from other users, although favorable in some instances, still seem to indicate general dubious results and questionable acceptance of the fuel for use in the reverberatory furnace. However, since the inauguration of oil firing of the holding furnaces at the writer's company, the process has been found very acceptable, and for our purposes and application far superior to the previous method of pulverized coal firing.

The complete duplex melting unit consists of two identical holding furnaces, both having an established overloaded capacity of 65 tons, and two 84-in. water cooled cupolas lined down to 60 in. in the melting zone. The air furnaces are operated on alternate days, but, in order to meet increasing foundry production demands, both cupolas are used for each day's heat.

Daily production, over a 16-hr. two shift period, averages between 350 and 375 tons. Peak hourly metal demand averages 30 tons per hour. This maximum hourly demand, being less than the total possible dual cupola output, enables the operator to maintain a nearly constant metal bath in the holding furnace during the time the cupolas are in operation.

Chemical analysis of the iron must be held to close specifications due to the nature and size of some of our production castings and to insure susceptibility of the metal to the short cycle anneal. Metal temperatures at the furnace spout are maintained close to 2900 F.

#### Furnace Construction

The shell of the reverberatory furnace, of home design and construction, is the semi-air-cooled bottom type. Originally the furnace was constructed with floating buckstays and sidewall sections. Dire results proved this design faulty and made it necessary to stiffen the construction with sunken buckstays and the addition of several cross members between the side wall and bottom sections. Internally the shell is 36 ft. long and 11 ft. wide. An overhanging plate is provided at the front of the furnace for oil burner mountings. The rear of the furnace adjoins the flue 6 ft. out from the stack entrance; the entrance being

diametrically opposite the second furnace at the base of a common stack. The furnace bottom plates are flat and join the side walls at right angles.

#### Refractory Construction

Refractory construction of the furnace furnishes a hearth area of 280 sq. ft., the completed furnace measuring 32 ft. in length between the bridge walls and 8 ft. 9 in. in width.

The front bridgewall is layed up from the bottom furnace plate with 9 in. of brick backed up by a 4½-in. safety wall and 4 in. of sand. As no combustion chamber is required for the burners, four arched openings, 14½ in. in diameter and 17 in. deep, directly behind the bridgewall, form simple burner ports. Sidewalls are 13½ in. thick from the bottom plate to the metal line and 9 in. thick above the metal line. It has been found that this mode of wall construction lends itself especially well to our operation. The heavy wall below the metal line provides adequate protection against breakouts, whereas the 9-in. wall above the metal line, due to faster heat dissipation, displays exceptionally good service life. The top of the sidewalls are level and describe a straight line forming no hung restrictions the entire length of the furnace. See Fig. 1 for sketch of furnace design and construction.

A 9-in. sq. test door is located 10 ft. from the front bridgewall and is used for taking flame temperatures, iron samples, observing flame characteristics and for making additions to the bath. The slag door is located approximately 8 ft. from the rear bridgewall of the furnace. Neither the slag door nor the test door are closed during operations. A slight pressure within the furnace prevents infiltration of air.

Two sets of two 1¼-in. tap out blocks are centered 10 ft. out from the burner ports and are directly opposite the test door. Three of the blocks are layed in 3 in. above the brick bottom for regular tapping. The remaining block opening is located at bottom level for drainage of the last iron and slag from the furnace.

The back bridgewall is built out from the stack to form a restriction 36 in. wide and 27 in. high, or, in terms of the area, 972 sq. in. The top of the bridgewall is level with the front bridgewall, 14 in. from the bottom of the furnace at the centerline. Both bridgewalls are 2½ in. above the metal line when the furnace is full to capacity.

A concave or inverted arch type brick bottom has been found very serviceable. The bottom shell of the furnace being flat does not lend itself very readily to this bottom layup method, but this is partially compensated for by a 4½-in. safety sub-bottom layed in dry on sand to the proper contours. Only on rare occasions do conditions necessitate removal of the safety sub-bottom. Hence, in addition to being an excellent safety factor from the standpoint of a possible bottom failure, it also makes an ideal and practically permanent backing for the 9-in. main bottom.

To act as a cushion for expansion and to serve as a sealer for the sub-bottom approximately 1 in. of sand is placed between the sub and main bottoms. The main bottom, Fig. 1, consists of rows of 9-in. brick layed in across the sand cushion between the side-

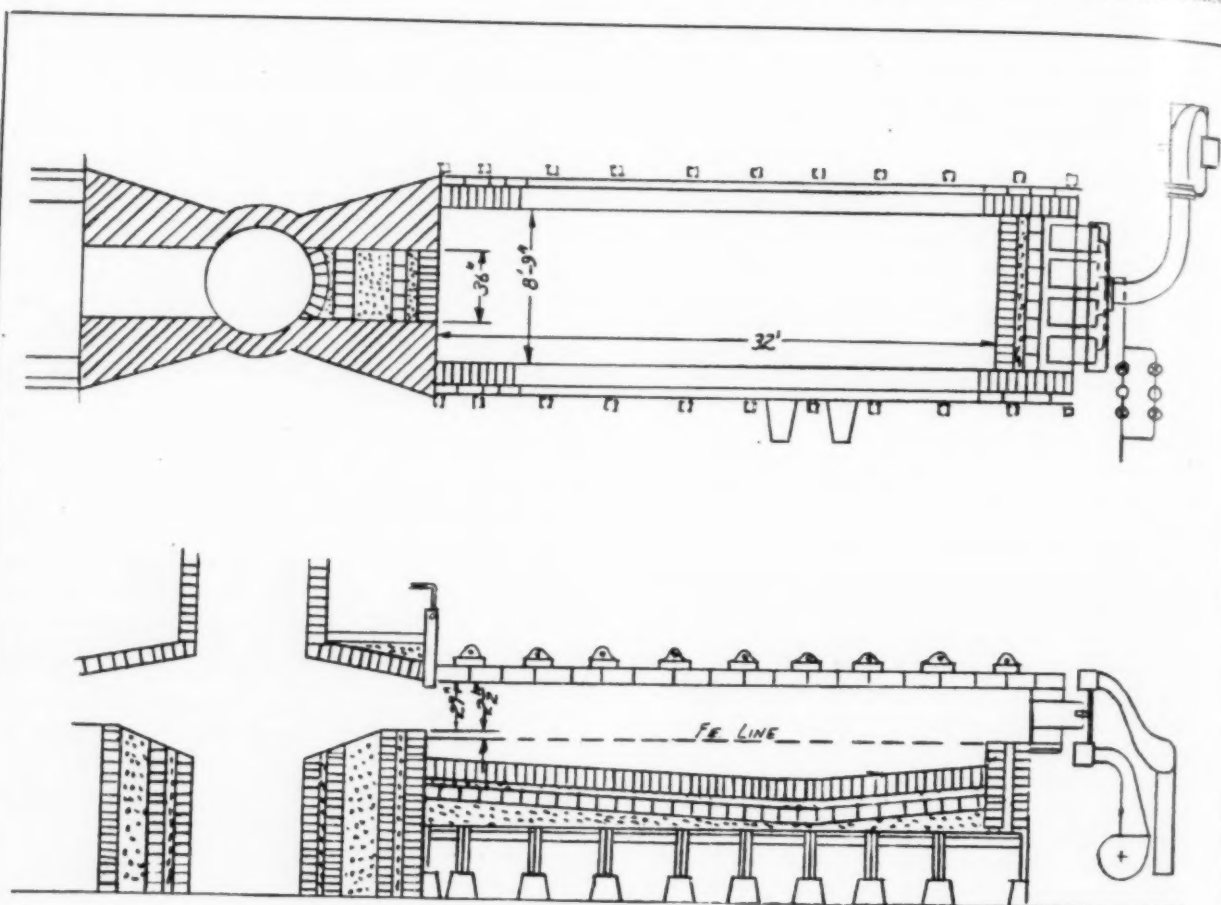


Fig. 1—Sectional sketch of oil-fired holding furnace.

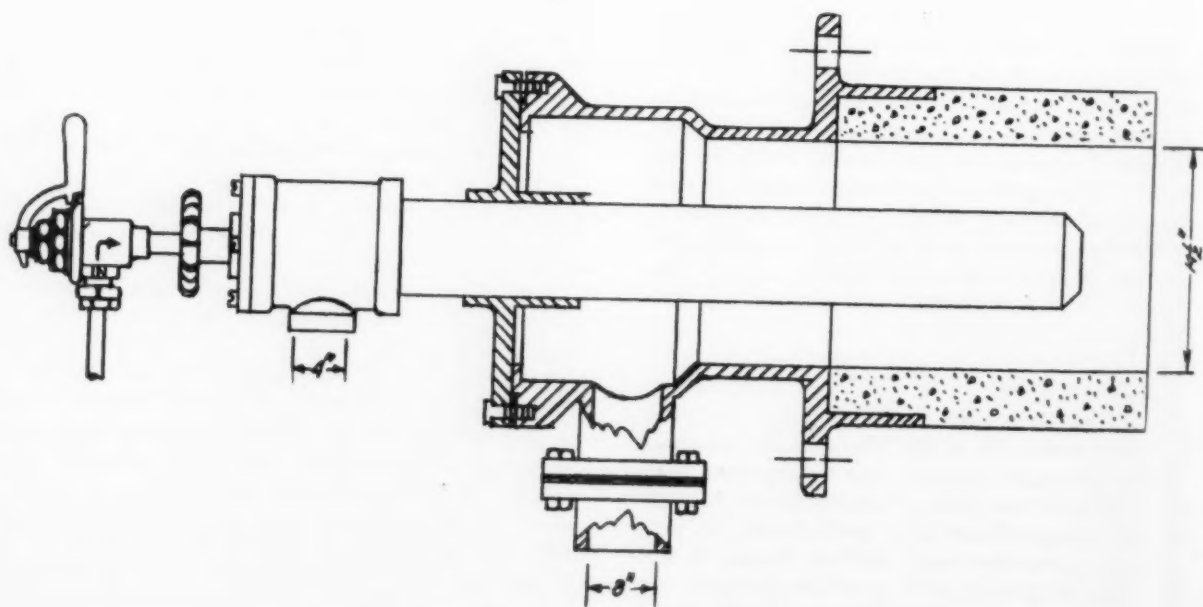


Fig. 2—Sectional view of luminous flame external mix oil burner.



walls to describe a 4-in. inverted arch. Only one wedge brick is used to key in each course. A slope of  $1\frac{1}{8}$  in. per ft. from the back bridge and  $\frac{1}{4}$  in. per ft. from the front bridge assures complete drainage of iron from the furnace.

Cupola iron enters the furnace a short distance out from the rear bridgewall just above the metal line. The delivery trough is set in at an angle away from the center of the furnace to prevent direct flow of cupola metal to the tap holes.

Bungs are of conventional design and are made up of a double row of 9-in. bung arch brick layed in a 10-ft. 9-in. frame with a 10-in. spring. Bungs are layed up dry.

All refractories used in air furnace construction and for repair work are of super duty quality. For general lay-up, a medium consistency slurry of high temperature thermal setting cement has been found ideal. High operating temperatures and long operating periods make the practice of using super duty refractories a necessity and economically sound.

### Oil Burner Equipment

Four 8-in. low pressure luminous flame external mix oil burners of 80 gal. per hr. capacity are mounted on the front burner plate with centers equidistant. A sectional view of the burner is shown in Fig. 2. To get proper flame impingement for highest possible flame temperatures above the tap holes and to cause movement of slag the burners are set in at an angle of  $5^\circ$  from level. The two outside burners are also faced in at an angle of  $5^\circ$  to protect the side walls from direct sweep of the flame. The burner body openings are  $14\frac{1}{2}$  in. in diameter and enter the furnace at a point 7 in. above the metal line and 38 in. behind the bridgewall.

For the first few months of oil burner operation a  $1\frac{3}{8}$ -in. diameter oil atomizing air orifice was used, but it was found, due to a slight deficiency of combustion air, advantageous to increase the size of the orifice to  $1\frac{1}{2}$ -in. diameter and use the extra air thus gained to supplement the combustion air when necessary. The oil nozzle has a straight  $\frac{1}{4}$ -in. bore.

Combustion air is supplied by a 5200 cu. ft. per min. blower operating at 7-oz. pressure. A fan type cupola blower with a rated capacity of 5100 cu. ft. per min. at  $1\frac{1}{2}$ -lb. pressure is used for atomizing. This value is greatly in excess of the required 680 cu. ft. per min. at  $1\frac{1}{2}$ -lb. pressure, but some of the superfluous air is used to advantage when critical carbon reduction is required and to augment the combustion air when necessary.

### Fuel Oil

No. 6 residual fuel oil of A.P.I. gravity 9-18 and a heating value of approximately 147,000 Btu per gal. is used for firing the furnace. The oil is stored in two heated sub-surface 30,000 gal. tanks and is supplied to the burners by means of an insulated recirculating oil line. The pumps and super-heater included in this line provide a constant supply of 170-200 F. oil at 60 psi to the burner main. The pressure is reduced to 38 lb. at the burners.

In view of the possible dangers that could result from an oil supply failure during operation, several precautionary measures have been taken to avert trouble. A stand-by steam boiler fired by combination gas-oil burners and two extra oil pumps are kept ready for operation should the regular equipment fail. Solenoid activated safety shut off valves stop oil flow to the burners in the event of a power failure. Other safety features include by-passing all filters, pressure regulators and meters. All uninsulated pipe lines have drain cocks.

The oil burner controls consist of adjustable port oil valves, graduated butterfly valves on the atomizing air lines and blast gates on the combustion air lines. Figure 3 is a view of the burner control arrangement.

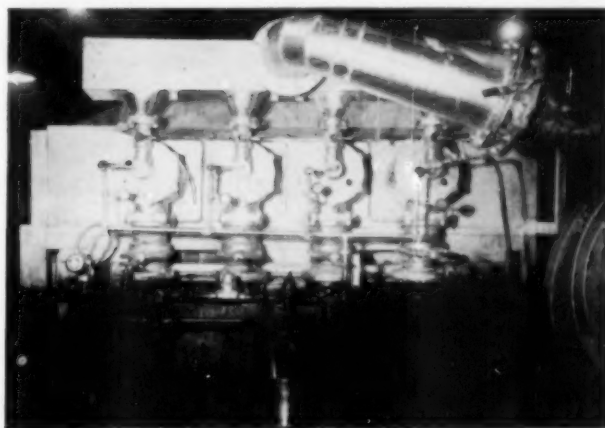


Fig. 3—Photo shows rear of oil burners and controls.

Pressure gauges are included on the individual burner oil delivery lines, on the main oil line and on both air manifolds. An oil line thermometer is located in the main distribution line a short distance from the burners. Oil consumption is indicated by a simple displacement meter which can be used for determining consumption per hour, consumption per ton as well as total oil consumption. Burner pilots are of the ordinary gas fired type.

### Furnace Atmosphere Control

During early experiments with oil it was found that furnace atmosphere control could not only be used to good advantage but is actually a requisite to efficient operation of the furnace and proper control of the metal bath. The instrument used, as shown in Fig. 4, is a differential type indicator and recorder which gives constant  $\text{CO}_2$  percentages of the discharged furnace gases. The gases are drawn in through a high-alloy water-cooled sampling tube, the end of which protrudes into the furnace to a point 4 in. below the bung line and directly above the center of the back bridgewall. To avert actual drawing in of burning gases, the sample tube is mounted at a slight angle away from the path of the flames. A cooling condenser and filter are in the line between the sample tube and the  $\text{CO}_2$  instrument to remove condensation and particles of foreign matter. Figure 5 displays layout of entire system.

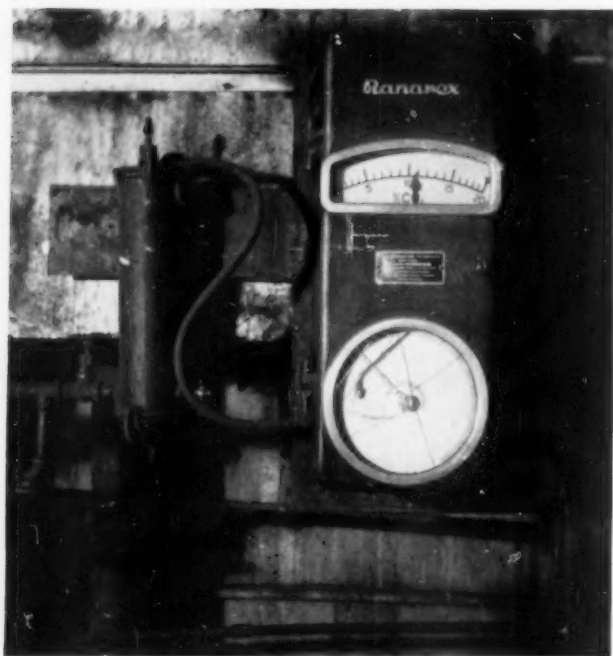


Fig. 4—Differential type  $\text{CO}_2$  indicator and recorder used for atmosphere control of the holding furnaces.

In operation the  $\text{CO}_2$  analyzer is simple and very accurate. The accuracy of the machine is checked at regular intervals against Orsat analyses. On very few occasions have any major discrepancies existed and these were apparent from mere observation before resorting to the use of the Orsat.

Through experience with the  $\text{CO}_2$  analyzer and the establishment of operating standards, the furnace operators are able to create practically any desired oxidizing or reducing atmosphere to produce proper effects on the bath of molten metal. The chart shown in Fig. 6 giving the percentage of excess air plotted against  $\text{CO}_2$ ,  $\text{CO}$ ,  $\text{O}_2$  and  $\text{C}$  analysis is the basis for the established standards for atmosphere control. By use of the chart the operator is able to determine fairly accurately the  $\text{CO}$ ,  $\text{O}_2$  and  $\text{C}$  percentages from the indicated  $\text{CO}_2$  analysis. Evaluation of this analysis in terms of its effect on the chemistry of the bath makes possible intelligent regulation of conditions.

Standard operating procedures have been derived by simple computation of excess air requirements to supply oxygen for oxidation of carbon and the attendant but usually undesirable oxidation of manganese, silicon and iron. This of course is in addition to the excess air required by the fuel depending upon its combustion characteristics.

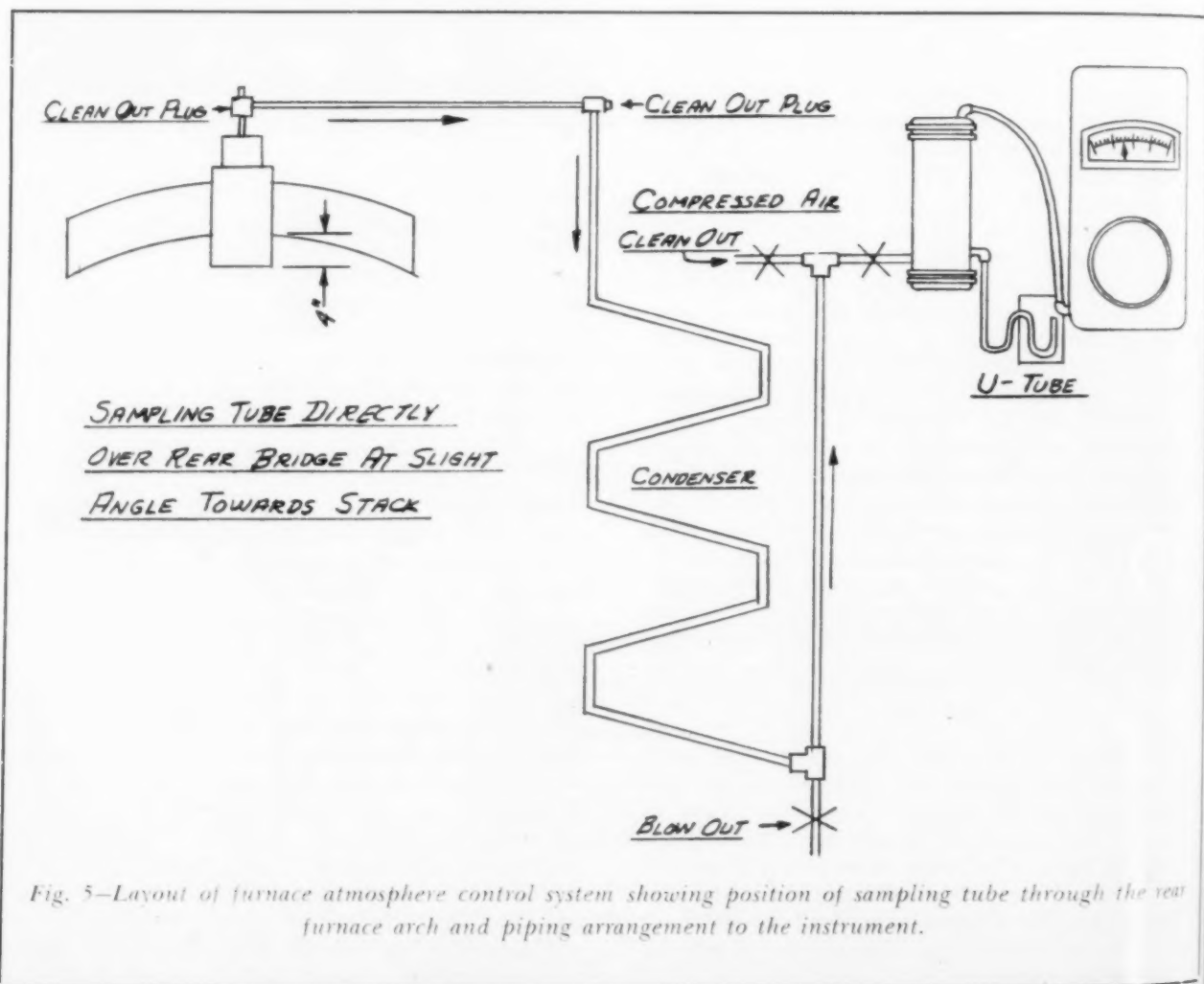


Fig. 5—Layout of furnace atmosphere control system showing position of sampling tube through the rear furnace arch and piping arrangement to the instrument.

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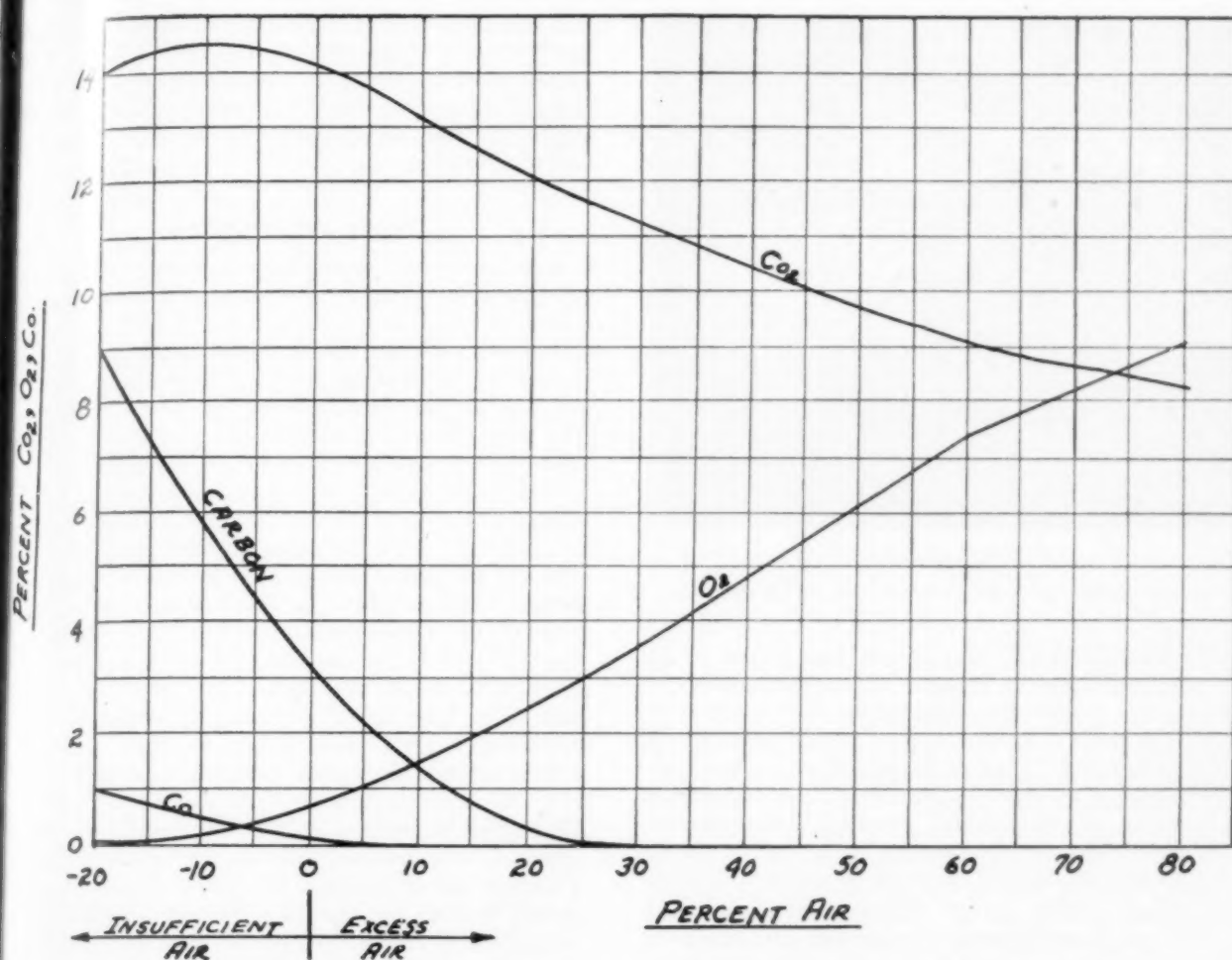


Fig. 6—Chart showing percentage of excess air plotted against  $\text{CO}_2$ ,  $\text{CO}$ ,  $\text{O}_2$  and  $\text{C}$  analysis.

#### Furnace Operation

Melting operations are regulated closely to coincide with foundry shift schedules. Since the foundry operates on a two-shift 16-hr. schedule the melting department must operate on a three-shift 24-hr. schedule utilizing the non-productive time for cupola maintenance, furnace preheat, starting and tapping the first cupola and to build up and adjust the chemistry of the furnace bath.

The oil furnace is started for preheat  $\frac{1}{2}$  hr. before tapping the first cupola. To prevent spalling of refractories, heavy firing is avoided until the furnace has been under fire for approximately 1 hr. The remaining 3 hr. is ample time to attain the standard operating temperature of approximately 3200 F at the test door. Temperature at the slag door will average close to 3100 F. Due to close time requirements necessitating a rapid filling of the furnace, together with over-capacitating the furnace, little divergence from standard temperature requirements can be tolerated if adequate metal temperature is to be obtained. Since the installation of oil firing, however, little difficulty has been experienced in obtaining or maintaining metal temperatures.

As the bath increases to approximately 30 per cent capacity, or 20 tons, repeating 30-min. chemical analy-

sis of the common control elements carbon, silicon and manganese is started and continued throughout the entire heat. Any required adjustments in the chemistry of the iron are made by ferro-alloy additions to the furnace or by manipulation of the furnace atmosphere as the bath approaches full capacity.

During the filling of the furnace and throughout the entire heat the  $\text{CO}_2$  analyzer serves its essential purpose in guiding furnace operation. If an oxidizing atmosphere is required the air is increased, the oil is decreased, or both, until the desired atmosphere indicated by the instrument and based on previously mentioned standards is obtained. Conversely, a reducing atmosphere can be created by increasing the oil, reducing the air or both. All burner adjustments are made with due regard to the maintenance of furnace and metal temperature.

Following the furnace tap-out, and as the trend of daily production demand is established, the cupolas are operated at an output rate calculated to preserve a bath of approximately 50 tons during the first 10 hr. of production. Consistent with the repair schedule, a single cupola is used to supply iron for the last 6 hr. of the second shift. This practice serves very well to gradually diminish the furnace bath without altering cupola operations until the second cupola is dropped.

The large hourly turn-over of iron in a furnace of





Side view of one holding furnace showing position of tap holes and burner mounting end plate.

such greatly deficient hearth area has proven to be no problem with oil firing. Carbon control, as explained in a preceding paragraph, is very effective even at periods of highest cupola output and highest foundry demand. A cupola output of 30 tons per hour containing an average of 2.55 carbon can consistently be reduced to 2.35. Table 1 shows the average oxidation losses and other pertinent data for a 1-week period. Furnace silicon and manganese analyses are not greatly altered except during periods requiring a very oxidizing furnace atmosphere. In these instances, silicon losses of over 3 points or manganese losses of over 6 points have not been experienced. Sulphur pick-up has been prevented through careful control of the selection and analysis of the fuel oil. Furnace slag is kept to a minimum until both cupolas have been dropped as an aid to the action of a desired oxidizing or reducing atmosphere on the chemistry of the iron. After cupola operations have ceased, slag is allowed to build up on the bath somewhat for protection against too great a loss of carbon. Furnace atmospheres must be very carefully controlled during this period and should be reducing.

Metal temperatures are taken at  $\frac{1}{2}$ -hr. intervals at the furnace spout by means of an optical pyrometer. The wide range of casting sizes and the various distances the iron has to be carried to the foundry pouring stations necessitates a metal temperature of not less than 2900 F and preferably closer to 2920 F. Inasmuch as the cupola iron temperature averages 2770-2800 F, the furnace must accomplish a temperature increase of at least 100-130 F. Similar to the uniformly successful attainment of sufficient carbon reduction with oil firing, temperature requirements can also be easily and consistently met.

#### Fuel Consumption

Fuel oil consumption for an iron tonnage output of 350 tons ranges between 165-200 gal. per hr. These figures represent an average hourly consumption, including oil used for preheating and for firing down

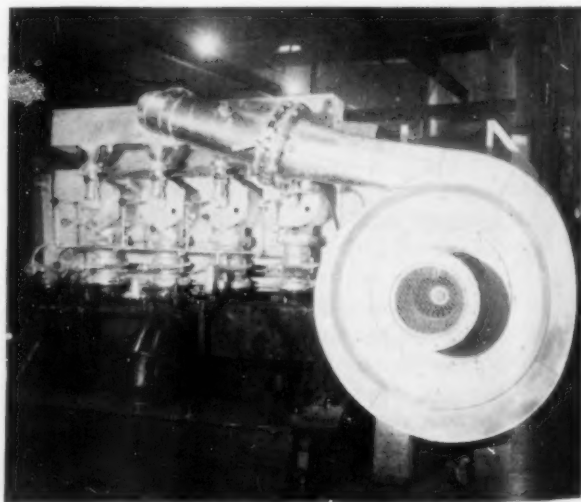
the furnace. This is equivalent to between 12.5-15.3 gal. of oil per ton of iron processed. It has been found, when operating the furnace for two consecutive days, that consumption drops as low as 11 gal. per ton due to the elimination of preheat firing for the second day of operation.

#### Refractory Consumption

Refractory consumption averages one 9-in. fire brick equivalent per ton of iron through the air furnace. Front furnace side walls approach a service life of three weeks which represents 216 firing hours. Rear sections of the sidewalls sometimes double this service life but on the average stand up for five weeks, or 351 firing hours. The front bridgewall and burner openings are usually replaced every six to eight weeks. Present construction of the back bridgewall necessitates greater repair than would be required if structural changes of the flue openings were made. Replacement of the entire bridge face is required every two weeks. Tap-out blocks are replaced every three heats. The blocks in most instances would give reliable service for five heats, but due to the heavy tonnages through the blocks and the extreme depth of the iron bath it has been found, from the standpoint of safety, desirable to change them after three heats.

The subject of furnace bottoms is often discussed in terms of the iron tonnages passed over the bottom. This figure is interesting, but in the writer's opinion provides little information relating to the quality of a bottom for comparison or control figures. For this reason, bottom life and all refractory consumption at our plant is computed on the basis of total hours under fire. Using this system average bottom service life is 500 firing hours, the highest ever obtained being over 900 hours and the lowest 297 hours.

Although any comparison of coal and oil firing has been omitted in the foregoing contents of this paper, the writer has listed below the advantages recognized at the author's company since the conversion from pulverized coal to oil firing. It is to be pointed out



Oil burner layout disclosing atomizing air blower and ducts.

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TABLE 1—HOLDING FURNACE OXIDATION LOSSES BASED ON AVERAGE DAILY ANALYSIS OF 5 CONSECUTIVE HEATS

Heat No.	Carbon			Silicon			Manganese			Max. Hourly Metal	
	Cupola Iron	Furnace Iron	Oxidation Loss	Cupola Iron	Furnace Iron	Oxidation Loss	Cupola Iron	Furnace Iron	Oxidation Loss	Bath, Tons	Turnover, Tons
1	2.56	2.33	0.21	1.66	1.65	0.01	0.62	0.57	0.05	50	30
2	2.51	2.33	0.18	1.62	1.59	0.03	0.62	0.57	0.05	56	30
3	2.60	2.37	0.23	1.60	1.60	0.00	0.57	0.56	0.01	50	26
4	2.57	2.31	0.26	1.63	1.63	0.00	0.59	0.56	0.03	50	30
5	2.56	2.36	0.20	1.67	1.64	0.03	0.60	0.56	0.04	52	32

and emphasized that the advantages gained by this company in resorting to fuel oil firing would not necessarily be realized in other installations.

#### 1. Lower fuel preparation costs

Two coal pulverizer operators were required per shift for coal drying, pulverizing and blowing. One operator is now required to operate the steam boiler and oil pumps per shift for the furnace oil supply.

#### 2. Lower fuel costs

This is due primarily to greater efficiency of fuel utilization. Computing this improvement on the basis of a straight Btu conversion from coal consumption figures in the old system the efficiency has been increased approximately 20 per cent.

#### 3. Lower refractory and refractory maintenance costs

Furnace refractory life has been markedly improved. Greater control of heating rate to prevent spalling, less abrasion due to absence of ash and controlled cooling are the principal reasons for this improvement in refractory life.

#### 4. Ability to hold greater furnace tonnages

A 65-ton bath is now possible without adversely affecting temperature or control. With coal firing 45 tons was the maximum allowable tonnage on the same hearth area.

#### 5. Improved iron analysis control through atmosphere control

Exact control of air and oil proportions based on constant atmosphere analysis makes this possible.

#### 6. Less maintenance of fuel supply equipment

This is due to less and simpler machinery.

#### 7. More uniform operation

Effected primarily through simpler and more exacting control of the fuel input.

### Control

Oil is generally known to be a very acceptable fuel for an increasing number of applications. As a holding furnace heating medium, it has proven to be a very efficient answer to the melting problem which confronted the author's company two years ago. It is felt that with further use and development of the equipment many more advantages of utilization will be recognized and put to gainful purposes.

## DISCUSSION

Chairman: W. D. McMILLAN, McCormick Works, International Harvester Co., Chicago

Co-Chairman: W. B. McFERRIN, Electro Metallurgical Co., Detroit

MILTON TILLEY (Written Discussion)<sup>1</sup>: The author is to be congratulated on an excellent presentation of the construction and operation of an oil fired malleable iron holding furnace. It may be recommended to anyone contemplating such a piece of equipment. It is especially interesting to note that the author

has found the gas analyzer to be of unquestioned benefit in control work as this point has been at issue heretofore when so many times the opinion has been that theoretically the gas analysis should be helpful while practically it has not been proven by experienced operators.

With regard to the value of oil as a fuel, it has been the writer's experience that, if oil fuel is not a necessity to obtain a metal suitable for the foundry, the choice is merely a matter of economics. It appears from the paper that the oil fuel was a necessity to obtain the results desired, mainly due to the ratio of holding furnace bath surface to the depth. That is, the rate of removal of carbon by coal flame was so low that the carbon of the cupola metal had to be kept low necessitating a low coke ratio which in turn resulted in a low cupola metal temperature. With a relatively low temperature metal entering the holding furnace the time available to raise the metal to the relatively high pouring temperature demanded by the pouring technique, was insufficient. By the use of oil, the difficulty was overcome by raising the temperature of the incoming metal. The higher carbon resulting from the higher temperature is reduced to the desired level by the greater rate of oxidation by the oil flame. In assessing the value of this idea, I am handicapped by not knowing the coke ratio, rate of melt and temperature of the cupola metal during the two periods of operating using coal and oil; and any change in preheated air in both melting units.

From the author's description of the operations it appears that the deep bath is imposed by the time for cupola repair problem and the high holding furnace metal tapping temperature by the loss in temperature between tapping and pouring.

It is unfortunate that this is so because the necessity for producing such high temperatures in a holding furnace of that type imposes a heavy burden on the operators. They are operating on the maximum with no deviation upwards instead of on a more comfortable middle range having a standard deviation to take care of unavoidable fluctuations.

One thing one may do is stagger the tap holes so that the metal could be drawing off with the upper holes when the furnace is full and tap with progressively lower holes as the metal level falls.

An obvious method of improving the situation is to widen the furnace, but in all probability the set-up is such as in most foundries, that it would be an impractical major change. Another is to install larger cupolas.

These suggestions on changes in technique are advanced because it appears that the problem of obtaining the high tapping temperatures is unnecessarily difficult.

The control of metal temperature is so fundamental in foundry economics that no liberties should be taken in setting up the melting technique.

R. P. SCHAUSS (Written Discussion)<sup>2</sup>: Mr. Coghlin has presented an able discussion of the construction and operation of an oil fired malleable iron holding furnace.

From the data of Fig. 7 we find that an average carbon drop of 22 points from cupola to air furnace has been obtained. In view of the limited hearth area and the melting rate, this is certainly well above any reduction in carbon which could be attained with pulverized coal fired equipment. The use of the carbon dioxide recorder in this system is a further indication of the progress which has been made in the operation of malleable air furnaces.

1. After you changed from coal fired to oil fired furnaces, did you get any change in the elongation and tensile strength of your iron for comparable carbon contents?

2. What charge do you use for your cupola? Are you using

<sup>1</sup> National Malleable and Steel Castings Co., Cleveland, Ohio.

<sup>2</sup> Illinois Clay Products Co., Chicago.

northern or southern coke?

3. Is it possible, as is the case with coal fired furnaces, to obtain any carbon pick-up in a clean bath by firing with a heavy reducing flame?

MILTON TILLEY: The kind of fuel used depends largely on the reduction of carbon which is a function of bath area in the furnace compared to the depth of the bath. The author in the paper has a two hours' or more bath in the furnace he speaks of. This lends itself readily to oil firing because oil will burn out carbon more readily than coal. You must be prepared to have a fairly deep bath compared to the surface area on that account.

We used oil at one time and had about one hour's worth of metal in the furnace and we burned out twice as much carbon as the author did on that account. The surface area was much greater compared to the depth which necessitated a higher carbon in the cupola. Sometimes that is convenient when we cannot get a low-carbon coke. You can utilize northern cokes which produce a higher carbon.

So the use of oil as against pulverized coal is connected with the carbon that you want to get out of your cupola as compared to the carbon you want to get out of your furnace. The economy would be pretty much connected with the price of the two fuels.

I would like to ask the author whether he tapped nearer the surface of the metal or whether he staggered the tap holes so as to tap off the higher metal nearer the surface of the bath and progressively lower tapped as the furnace went down towards the end of the day. You would have to be careful about having too deep a bath if you are going to tap out near the bottom. You would not have hot metal near the bottom of your furnace, with either oil or coal firing.

The author made mention of the square end to the furnace near the bridge wall and that he did not like this design. When operating near the peak efficiency of the fuel, it is necessary to utilize all the fine points of heat transfer from the flame to the metal and one of those points is an even flow of the products of combustion. A square end on the end of the furnace interrupts that even flow. Anybody contemplating building a furnace should take that into consideration and not introduce an interruption to the even flow such as the author has on the end of that furnace.

MR. COGHILIN: I did not explain the first point mentioned by Mr. Tilley too well in the paper. We do stagger our tap-out block from high to low. We use the bottom tap-out blocks solely for draining the furnace and for eliminating what slag we can at the end of the heat.

I have a question from Mr. Wurscher in which he asks what my reason was for objecting to the floating buckstays. When I discussed floating buckstays I had reference solely to the floating buckstays we have on the furnaces at our plant. We found it necessary to stiffen them up with 120-lb rails sunken in the floor. It was also necessary to add some cross members. I do believe, however, that if a furnace can be built which will allow free expansion only to the extent that it prevents actual crushing of the brick, it will be approaching the ideal set up.

F. J. WURSCHER:<sup>2</sup> I would like to know what the actual difficulties were; whether or not the floating design of the furnace was conducive to reducing the length of life of the brick bottom.

MR. COGHILIN: Only through actual failure of the bottom; the steel buckstays were not strong enough to take the expansion, so actually we would get the same effect as a cork coming up out of a bottle. The sidewalls would start to move out after a certain amount of expansion of the brick, and then, if the bricks reached the softening point all the way through we would lose the bottom. We went through an epidemic of this lasting about two weeks and finally found it necessary to stiffen up the construction. Since that time, we have had amazingly good results with our bottoms. But, it was only our own design that I was objecting to.

MR. COGHILIN: In reply to Mr. Schauss' written discussion there was no noticeable change in the mechanical properties when the change from coal firing to oil firing was made. We are using northern coke as a fuel. It is very definitely possible to pick up carbon on a clean bath of iron with the oil firing.

We accomplish this by using a reducing flame. We try to maintain at all times an operating temperature of the metal of 2900 F and about 3200 F flame temperature. With a 50-ton bath we can accomplish a great deal of carbon pick up if it is required, through a reducing atmosphere in the furnace.

MR. SCHAUM: For what length of time do you leave that reducing atmosphere in effect?

MR. COGHILIN: We can maintain a strongly reducing atmosphere for a considerable length of time without adversely affecting the metal temperature. We can operate with a reducing flame for 3 or 4 hr depending upon the amount of carbon pick up required. A 4-hr period would be the exception, however, because we rarely want to pick up carbon. The only time we would encounter low carbon analysis would be on a new bottom or through some other irregularity.

W. R. JAESCHKE:<sup>4</sup> I think if you go into the history of this job, you will find that the author did not design or layout the equipment as he is operating it now. In other words, he inherited an unbalanced set of conditions and found a solution for his problem by resorting to oil firing. An analysis, I do not think, can be covered any more clearly than Milton Tilley covered it in his written discussion. If you consider all those points, it is really a matter of economics, and in this case the author took an unbalanced set of conditions and with oil firing he was able to get the results he required.

MR. COGHILIN: That all depends upon what you mean by an unbalanced set of conditions. The conditions that were prevalent in malleable iron foundries years ago and were accepted as standard conditions, any deviation from which being regarded as an unbalanced set of conditions, are constantly changing, so I would like to know what you mean by a standard set of conditions?

MR. JAESCHKE: The conditions are unbalanced for the job at hand. I mean, with pulverized coal firing, your ratio of hearth area to depth of bath, melting rate and tonnage in the furnace was out of balance for pulverized coal firing and you found the solution to that through swinging to oil firing at probably an expense of quite a few dollars per year in increased fuel costs. In other words, if you were going to build the job new today, would you build it as you have it or would you build it properly for a lower priced fuel?

MR. COGHILIN: I believe we would build it as we have it today because we are actually saving money on fuel.

W. R. BEAN:<sup>4</sup> A gallon of fuel oil and 10 lb of bituminous coal have approximately the same heat value. On that basis, the amount of oil used would be approximately the equivalent of from 120 to 150 lb of coal per ton. This 12 to 15 gal of oil is approximately equivalent in heat value to 120 to 150 lb of coal. That much coal is more than is required for malleable duplexing with equipment that is installed with regard to the fundamentals of air furnace operation and air furnace control.

MR. COGHILIN: A Btu study reveals that with an input of 14.5 gal of No. 6 fuel oil we are consuming approximately 2,100,000 Btu's per ton of iron. With coal, we were using, on a maximum 45-ton bath, approximately 230 lb per ton, which is equivalent to about 3,200,000 Btu input. This indicates a great increase in efficiency of fuel utilization.

MR. BEAN: That, I think, goes back to the original question asked by Mr. Jaeschke. You did not have a normal operating condition for pulverized coal. You were operating a central station pulverizing plant, which was installed back in the early 20's. The conditions under which you were using coal were far from normal for air furnace operating, duplexing or direct melting. At the time that you changed from direct melting to duplex melting, the central station plant had long been abandoned, that is, of the type of operation which you had, so that in making a comparison, it is necessary to take into account the factors inherent in the fundamental basis of pulverized coal-fired duplexing.

CHAIRMAN McMILLAN: Does the per cent carbon in Fig. 6 refer to the composition of the metal?

MR. COGHILIN: No, it does not.

CHAIRMAN McMILLAN: Does the carbon curve indicate the amount of soot in the gas?

MR. COGHILIN: That is right.

CHAIRMAN McMILLAN: Is it on the same ordinate as the CO and CO<sub>2</sub>?

<sup>2</sup> Chicago Railway Equipment Co., Marion, Ind.

<sup>4</sup> Whiting Corp., Harvey, Ill.



Mr. COGHILIN: Yes.

MEMBER: Why did you select as your sampling point of the furnace a position 4 in. below the center line of the bung?

Mr. COGHILIN: It seems that our point of sampling is the logical point inasmuch as it is situated in a position that permits the drawing in of the discharge gases as they are about to leave the furnace and after they have passed over the entire bath of metal. Operating the furnace under pressure at all times eliminates dilution of the gases through infiltration of air.

Mr. TILLEY: It has been our experience that without any change in operation or in the furnace, we ran about 10 gal of fuel oil per ton of metal melted. When we changed to oil firing, we ran 100 lb of coal. That is, just 10 times, according to Btu value of the coal (14,000 Btu/lb coal, 140,000 Btu/gal oil).

Mr. COGHILIN: I believe you are running at lower temperatures than we are. That probably accounts for the difference in our respective fuel consumption figures.

In reply to Mr. Bean's comment, I think he brought home one thing I omitted. That is, we can not compare coal firing at other plants with the coal firing we had, nor can we compare our present oil firing on fuel consumption figures alone. It would be absolutely foolish to try to do so. I was merely trying to bring out the advantages that we did gain in changing over from our old coal system to our present oil firing system. That is not to be confused with what we might have gained by putting in a completely new coal unit.

Mr. BEAN: I was interested in the author's comments on the sidewall life, as to the number of hours of service obtained from a 9-in. wall thickness. I believe the author made the statement that he used the 9-in. wall in order to obtain a higher heat transfer to the atmosphere and protect the working surface of the brick. That is of interest to me from the standpoint that it may be possible that fuel is cheaper than brick, plus the cost of putting brick into a furnace. The usual wall in duplexing furnaces is 13½ in. and the same is true of the direct-firing units. An 18-in. wall will not last as long as a 13-in. wall, but there are very few 9-in. wall air furnaces in operation today. In the olden days, there were more. Do you carry that 9-in. wall down to the bottom?

Mr. COGHILIN: No, we do not. We use a 13½-in. wall all the way up to the metal line and from the metal line up, we use a 9-in. wall.

Mr. BEAN: In other words, you get slag erosion on the 13½-in. wall and the heat effect, the actual melting effect on the refractories, from the flame and temperature, comes on the 9-in. wall.

Mr. COGHILIN: Yes.

Mr. BEAN: The 9-in. wall lasts the hours of firing as you mentioned. The hours of firing rather than the tonnage is the basis for air furnace life.

Mr. COGHILIN: The hours of brick life under fire refer to the entire inside of the furnace, the inside 9 in. of brick that extends right to the bottom of the furnace. It includes the section of wall that is backed up by the 4½-in. wall.

# SURFACE GAS PRESSURE OF MOLDING SANDS AND CORES

By

H. W. Dietert\*, H. H. Fairfield\*, F. S. Brewster\*

## ABSTRACT

The gas pressure within a core or molding sand specimen  $1\frac{1}{8}$  in. in dia. by 3 in. in length was determined by ramming a  $\frac{3}{16}$ -in. inside diameter brass tube within the core. The tube extended to  $\frac{1}{8}$  in. of the surface of the core or sand. Thus, the gas pressure  $\frac{1}{8}$  in. from the outer surface of the core or sand could be measured. A sensitive diaphragm type pressure gauge, calibrated from 0 to 50 in. of water pressure, was used. The core and molding sand specimens tested were immersed in a molten lead bath held at 1200 F. This type of bath was used in place of a molten metal held in a ladle because it could be controlled more easily. As may be expected, the gas pressure was found to increase as the moisture, seacoal, oil or cereal content was increased. Extremely high mold hardness also increased gas pressure.

The effect of baking time on core gas pressure was also shown.

The gas pressure test seems to offer a simple technique for measuring an important property of sand mixtures.

## Introduction

IN COMPOUNDING MOLDING SAND OR CORE mixtures, consideration should be given to the gas pressure that will be created by the vaporization and combustion of materials in the sand or core due to the heat of the molten and solidifying metal. The gas pressure permissible at the surface of a green or dry sand mold or core is predicated to a large extent by the ferrostatic pressure and chemical activity of the molten metal. At present, the permissible gas pressure allowable must be determined by the trial and error method in the foundry. Progress may be best made in this field when equipment is at hand to measure the gas pressure produced by each sand or core mixture under standardized conditions. Knowing the gas pressure of selected sand or core mixtures will soon lead to definite information concerning permissible gas pressure providing gas pressure readings as compared with foundry casting results. This paper describes a test procedure for measuring gas pressure and some of the results obtained.

## Method Described

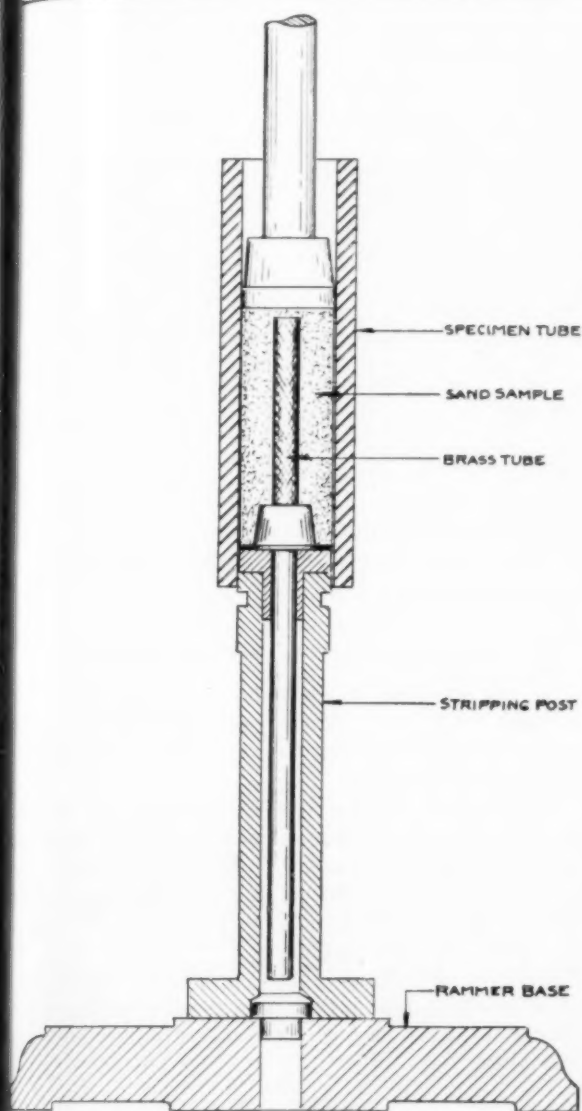
The test specimen selected measures  $1\frac{1}{8}$  in. in dia. and 3 in. in length. It is rammed with the  $1\frac{1}{8}$ -in.

sand rammer unit as illustrated in Fig. 1 with three drops of a 7-lb. weight falling  $2\frac{3}{8}$  in. Within the test specimen is located a  $\frac{1}{4}$ -in. O.D. x  $\frac{3}{16}$ -in. I.D. brass tube. This tube reaches within  $\frac{1}{8}$  in. of the end of the sand or core specimen. The cross-sectional view of the ramming assembly is shown in Fig. 2. A detail



Fig. 1—Sand rammer for forming  $1\frac{1}{8}$ -in. diameter specimens.

\* President, Foundry Consultant, Sales Manager, respectively of the Harry W. Dietert Company.



METHOD OF RAMMING  
SURFACE GAS PRESSURE SPECIMEN

Fig. 2—Method of ramming gas pressure specimens  
 $1\frac{1}{8}$  in. x 3 in.

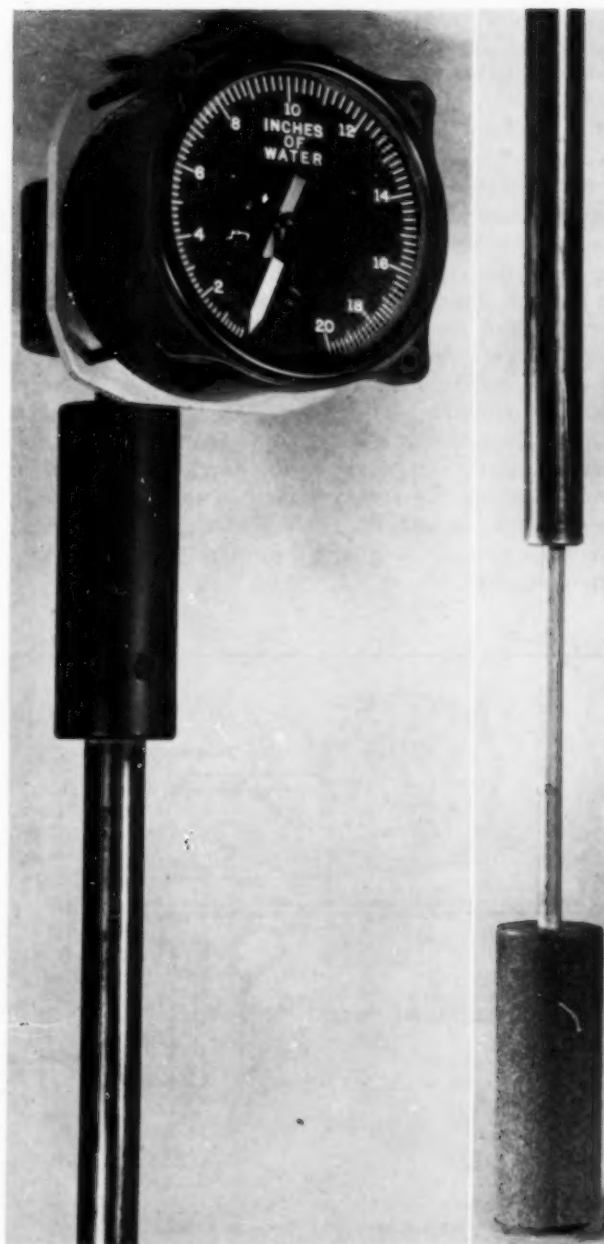
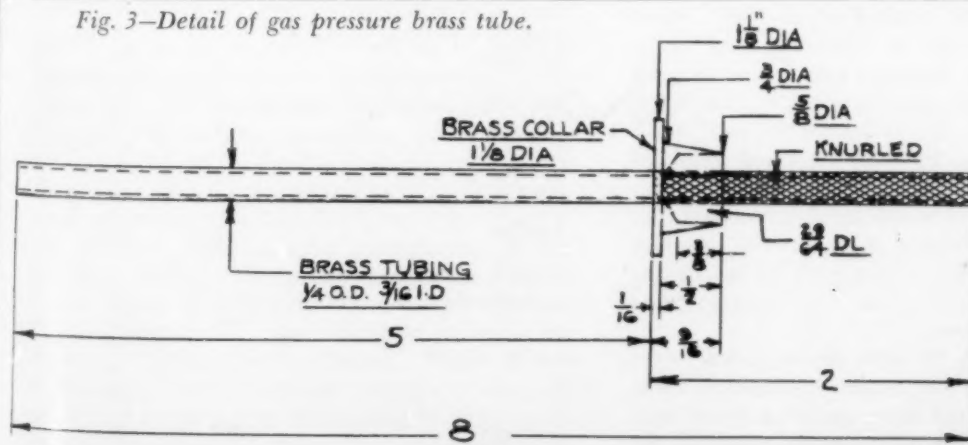


Fig. 4—Portable gas pressure determinator.

Fig. 3—Detail of gas pressure brass tube.





of the 1/4-in. O.D. brass tube that is rammed in the specimen is shown in Fig. 3. The rammed sand or core specimen is removed from the specimen tube by a stripping operation. In order to prevent sand from entering the brass tube, a 1/8-in. brass rod is placed in the tube while the specimen is rammed. This rod is removed after the stripping operation. There is no sand in the brass tube when the test is performed. The exposed end of the brass tube is attached to a pressure gauge reading in inches of water. A portable gas pressure determinator is shown in Fig. 4. This unit is of such length that the specimen may be inserted into molten metal held in a foundry ladle. Another form of gas pressure measuring unit which is simple is illustrated in Fig. 5. The laboratory gas pressure determinator, Fig. 5, uses a molten lead bath held at 1200 F, into which the specimen is inserted. All gas pressure test results given in this report were obtained by immersing the test specimen in molten lead at a temperature of 1200 F. The portable gas determinator may also be employed with a molten lead bath to heat the specimen.

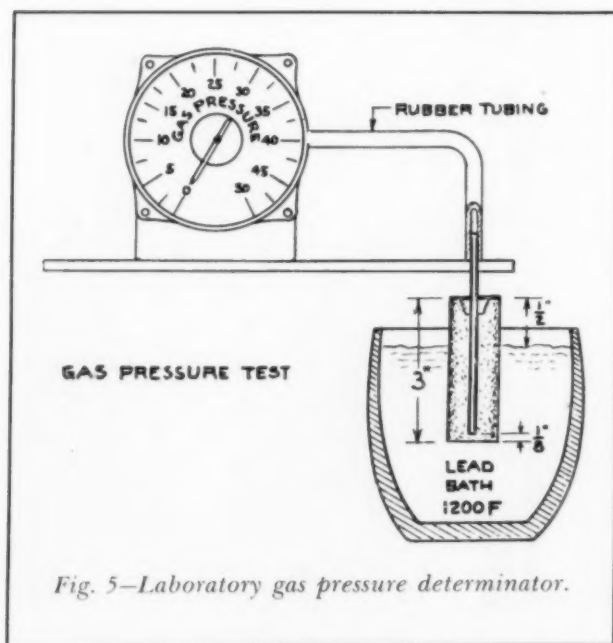


Fig. 5—Laboratory gas pressure determinator.

The molten lead bath method allows for a good laboratory method and has the advantage that high gas pressure sand or core mixtures may be tested without incurring boiling of metal around the specimen as may be the case when lighter metals are used. The lead bath method allows for easy standardization of metal temperatures.

The rammed sand or core specimen is inserted into the molten metal bath either in the green, dried or baked condition as desired, to within 1/2 in. of the upper end. A tolerance of 1/4 in. will not incur any error of practical significance.

Core specimens used in this investigation were baked at 425 F for 1 hr. in a core test baking oven. In most cases five readings were made on every mixture tested. The maximum pressure was observed and

recorded. The time in seconds required to reach the maximum gas pressure was also recorded.

The condensation of moisture in the tubing would be expected to effect maximum pressure obtained. The gas pressure measuring system, however, is an enclosed tube, and there is no flow of gases through the tube or through the gas pressure meter.

The maximum gas pressure was obtained within a period of about 30 to 50 sec. It is assumed that the first surge of gas or steam heats the tube to a temperature in excess of 220 F and forms a vapor-lock. The gas generated after that exerts a considerable pressure but does not condense because the tube is already heated up. It was observed that only the first 10 in. of the tube heated up. The upper part of the measuring device never became warm. Condensation was not observed in the upper end of the gas measuring device after continuous use for several hours.

From the observations made, it was concluded that the steam formed remained in the brass tube which extended through the sand specimen and 5 in. above it. This brass tube was heated hot enough so that condensation did not occur.

#### Sand and Core Composition Studied

In the interest of simplification, only one basic molding sand, No. 3 Albany, was used in this investigation, varying such items as moisture, seacoal content and mold hardness.

Equal parts of Manistee Dune Sand and Juniata Bank Sand both produced in Michigan were used to form the basic core sand. The core oil and cereal content were varied in the core sand mixture to show how gas pressure of the core was effected. The degree of baking was also varied.

A series of five core sand mixtures ranging from 11 to 510 baked permeability was used in this investigation to show how gas pressure changes with permeability.

TABLE 1—EFFECT OF MOISTURE PERCENTAGE IN No. 3 ALBANY SAND ON GAS PRESSURE

Percentage Moisture	0	6	9
Mold Hardness	..	90	87
Green Permeability	..	37.5	69
Green Compression, psi	..	12.0	7.9
Green Deformation, in./in.	..	0.019	0.025
Flowability	..	73	70
Max. Gas Pressure, In. of Water	3.8	16.5	19
	4.1	16.5	17.5
	4.5	16.0	17.4
	4.9	15.7	19.5
		15.9	
Average	4.3	16.1	18.3
Time in Sec. to reach Max. Gas Pressure	41	28	34
	40	27	38
		29	36
		32	35
		26	
Average	40.5	28	35.6

# Effect of Moisture on Gas Pressure

Much attention has been given to moisture control in molding sand and justly so. Appreciating the fact that moisture in sand creates a large amount of gas, many will be interested in knowing how much gas pressure in inches of water is created for each percentage of water added to the sand. Test data is tabulated in Table 1 showing the physical properties of the No. 3 Albany Sand and the gas pressures in inches of water for moisture percentages of 0, 6 and 9. In the case of the dried sand specimen, a time of 40.5 sec. elapsed before maximum pressure of 4.3 in. was created, while for the 6 per cent moisture specimen, the time was only 28 sec. to create 20.5 in. of pressure. The moisture in the dried sand specimen is largely combined and therefore is liberated slowly. Gas pressure is also created by expansion of air in the sand. The gas pressure created for 0, 6 and 9 percentage of moisture is shown by graph in Fig. 6. Note the rapid rise of gas pressure as moisture content is increased.

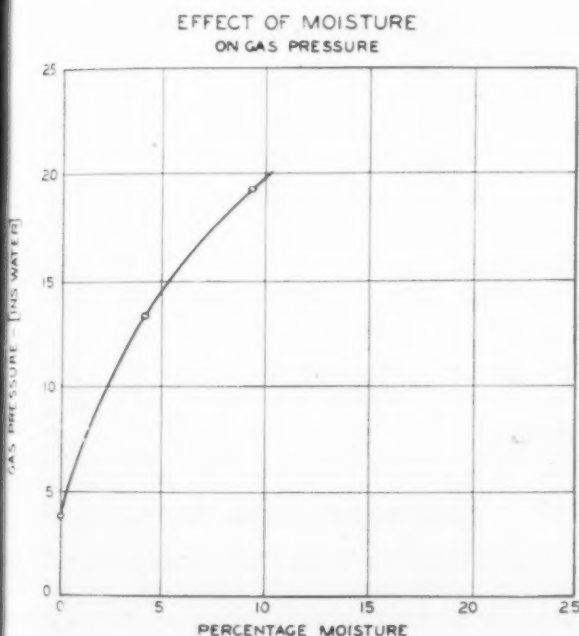


Fig. 6—Effect of moisture on gas pressure.

## Seacoal Additions Increase Gas Pressure

A portion of the facing value of seacoal in molding sand is undoubtedly due to the gas pressure that the seacoal creates. The gas pressure may at times play a very important part in preventing metal penetration by preventing metal flowing into the voids in the mold face. The magnitude of the gas pressure created by seacoal additions to a No. 3 Albany Sand is tabulated in Table 2. The average gas pressure of the No. 3 Albany Sand is 4.3 in. water pressure. When 5 per cent seacoal is added, the gas pressure increases to 20.5 in. of water. The time required to reach the maximum gas pressure is not affected. When 10 per cent of seacoal is added, the gas pressure reaches an average of 29.9 in. of water. This amount of gas pressure seems to slow up heat-travel into the sand specimen

since 96.7 sec. on the average is required for gas pressure to reach the maximum value.

A graphical presentation of how seacoal additions to No. 3 Albany Sand affect the gas pressure is shown in Fig. 7.

## EFFECT OF SEA COAL ON GAS PRESSURE

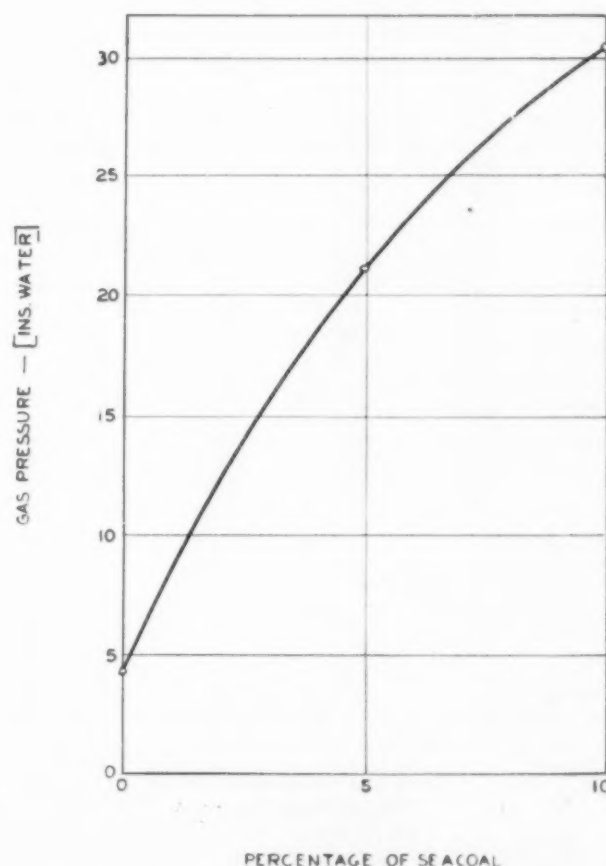


Fig. 7—Effect of seacoal content on gas pressure.

TABLE 2—EFFECT OF SEACOAL CONTENT ON GAS PRESSURE OF NO. 3 ALBANY MOLDING SAND

Percentage Seacoal	0	5	10
Moisture, per cent	5.9	6.0	6.4
Green Compression, psi	12.0	12.15	11.6
Green Deformation, in./in.	0.019	0.017	0.017
Green Permeability	37.5	22	19.25
Flowability	73	69	70
Max. Gas Pressure, In. of Water	3.8	19	30.5
	4.1	21.6	30
	4.5	20.6	30
	4.9	20.5	29.2
		21.2	
Average	4.3	20.5	29.9
Time in Sec. to Reach Max. Gas Pressure	41	42	95
	40	40	100
		40	97
		44	95
		41	
Average	40.5	40.1	96.7

### Ramming Affects Gas Pressure

Data pertaining to the effect of mold hardness on the gas pressure is tabulated in Table 3. A graphical illustration is shown in Fig. 8. It may be noted that as the mold hardness is increased from 60 to 96, the time required to create maximum gas pressure is considerably increased from 12 to 31 sec. This is fortunate in that it makes this particular sand fairly fool-proof as to the degree of ramming. The longer it requires to reach maximum gas pressure, the less likelihood of experiencing blows. A rapid rise in gas pressure was found when mold hardness was increased from 90 to 96. In this range, the gas passages in the sand were undoubtedly too small for free passage of

TABLE 3—EFFECT OF MOLD HARDNESS ON GAS PRESSURE OF NO. 3 ALBANY MOLDING SAND

Mold Hardness	60	78	90	96
Moisture, per cent	5.9	...	5.9	6.0
Green Compression, psi	3.3	...	12.0	28.5
Green Permeability	120	...	37.5	17
Max. Gas Pressure, In. of Water	15	13.3	12.8	19.3
	15	13.2	13.2	20.4
		13.4	14	19.3
Average	15	13.3	13.3	19.6
Time in Sec. to Reach Max. Gas Pressure	12	16	21	37
		16	20	28
		16	21	28
Average	12	16	20.6	31

### EFFECT OF MOLD HARDNESS ON GAS PRESSURE

#### MOLDING SAND

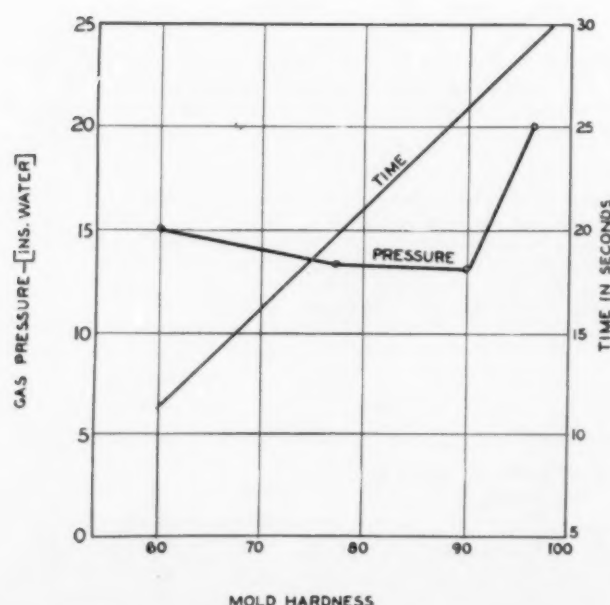


Fig. 8—Effect of mold hardness on gas pressure.

the gas volume created. The phenomena of gas pressure reduction when mold hardness was increased from 60 to 90 was unexpected.

### Core Oil Increases Gas Pressure

The quantity of core oil binder added to a core mixture largely predetermines the amount of gas volume produced in a core, providing other factors are identical. Thus, the gas pressure may be expected to increase as the quantity of oil is increased. This is substantiated by test data as shown in Table 4 and Fig. 9.

TABLE 4—EFFECT OF OIL CONTENT ON GAS PRESSURE OF A CORE MIXTURE

Core Mixture No.	1	2
Raw Linseed Oil, lb	20	40
Manistee Sand, lb	1000	1000
Juniata Sand, lb	1000	1000
Water, lb	40	40
Baking Temperature, F	425	425
Baking Time, hr	1	1
Baked Permeability	90	85
Max. Gas Pressure, In. of Water	1.0	1.8
	1.0	2.2
	1.3	1.9
	1.5	1.6
	1.2	2.1
Average	1.2	1.9
Time in Sec. to Reach Max. Gas Pressure	41	36
	40	36
	44	36
	40	37
	40	35
Average	41	36

### Cereal Binder Increases Gas Pressure

Most binders for cores increase the volume of the gas generated, thereby increasing the gas pressure, when the core is heated by the molten metal. It is of practical value for foundrymen to study the gas pressure created by various additions of each binder that may be of use in his core room.

The rate at which the additions of a cereal binder, corn base, increases the gas pressure is shown graphically in Fig. 10. The test data is tabulated in Table 5. The core containing 2 per cent oil binder and no cereal showed a gas pressure of 1.9 in. of water when immersed in a lead bath at 1200 F. An addition of 0.5 per cent cereal increased the gas pressure to 3.4, and to 4.7 when 1.5 cereal was added.

### Permeability

Baked permeability provides an index of gas pressure, only if binder content and method of baking is held constant.

Figure 11 shows the general relationship between permeability and gas pressure. In these experiments the amount of binder was held constant. In actual



# EFFECT OF CORE OIL ON GAS PRESSURE OF A CORE

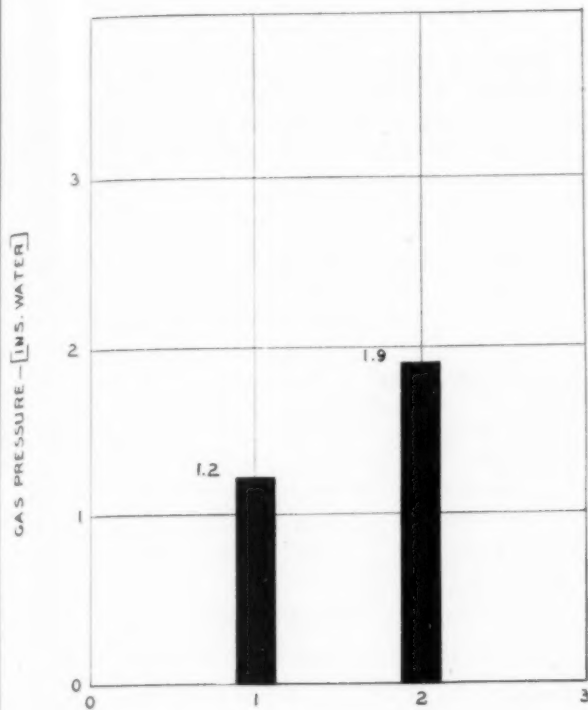


Fig. 9—Effect of core oil on gas pressure.

# EFFECT OF PERMEABILITY ON CORE GAS PRESSURE

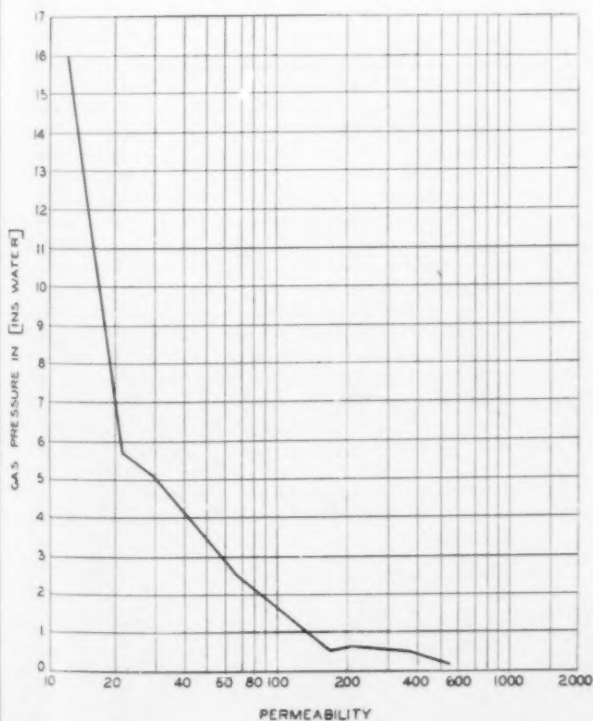


Fig. 11—Effect of permeability on gas pressure.

# EFFECT OF CEREAL CONTENT UPON GAS PRESSURE OF A CORE

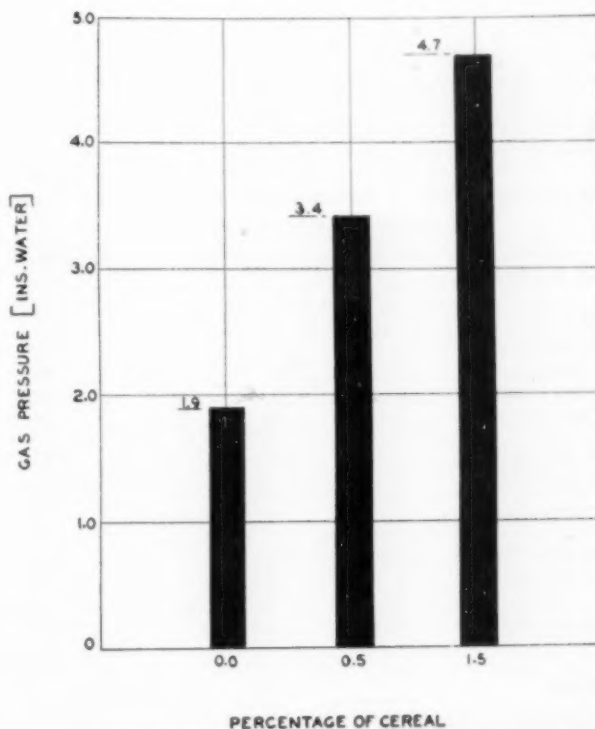


Fig. 10—Effect of cereal content on gas pressure.

# EFFECT OF BAKING TIME ON CORE GAS PRESSURE

CORE MIXTURE:  
100 MESH SILICA SAND 88.5%  
SILICA FLOUR 10.0%  
RAW LINSEED OIL 1.0%  
CEREAL FLOUR 0.5%  
WATER ADDED 3.0%

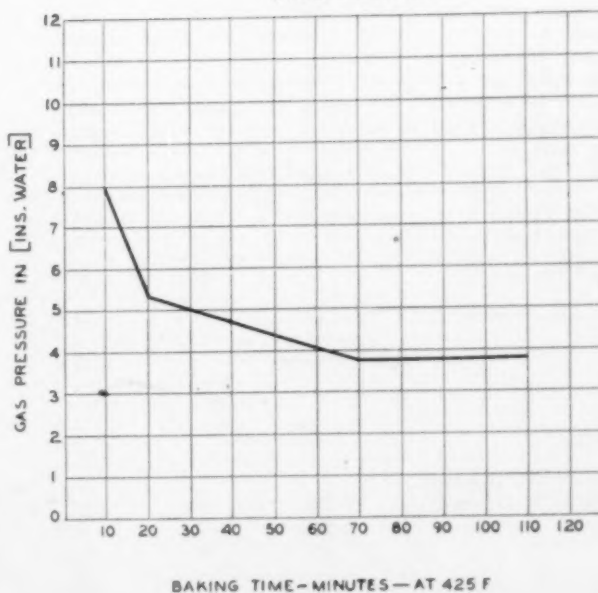


Fig. 12—Effect of baking time on gas pressure of a core.

foundry practice, the finer sands would require a greater amount of binder. See Table 6.

For the test core used, a permeability of 170 will reduce gas pressure to a minimum value.

#### Baking Time Affects Gas Pressure

The gas pressure test proved to be useful in measuring the progress of baking. After 10 min. at 425 F,

TABLE 5—EFFECT OF CEREAL CONTENT ON GAS PRESSURE OF A CORE MIXTURE

Core Mixture No.	3	4	5
Cereal, per cent	0	0.5	1.5
Raw Linseed Oil, lb	40	40	40
Manistee, lb	1000	1000	1000
Juniata Sand, lb	1000	1000	1000
Water, lb	40	40	40
Baking Temp., F	425	425	425
Baking Time, hr	1	1	1
Baked Permeability	85	75	90
Max. Gas Pressure, In. of Water	1.8 2.2 1.9 1.6 2.1	3.1 3.7 3.4 3.3 ...	4.5 4.6 4.9 4.8 ...
Average	1.9	3.3	4.7
Time in Sec. to Reach Max. Gas Pressure	36 36 36 37 35	39 39 35 38 ...	38 38 38 38 ...
Average	36	37.7	38

TABLE 6—EFFECT OF PERMEABILITY ON GAS PRESSURE OF A CORE

Core Mixture No.	6	7	8	9	10	11	12	13	14
Raw Linseed Oil, lb	20	20	20	20	20	20	20	20	20
Cereal, lb	10	10	10	10	10	10	10	10	10
Western Bentonite, lb	..	..	..	20	..	20	20	20	20
Silica Flour, lb	500	..	..	..	..	..	..	20	..
Wedron Sand, lb	1500	2000	1000	500	..	..	..	..	..
A.F.A. Standard Sand, lb	..	..	1000	1500	2000	2000	1000	500	300
Blast Sand, lb	..	..	..	..	..	..	1000	1500	1700
Water, lb	160	80	60	60	60	..	..	..	..
Baking Temp., F	425	425	425	425	425	425	425	425	425
Baking Time, hr	1	1	1	1	1	1	1	1	1
Baked Permeability	11	22	30	67	170	208	231	380	510
Max. Gas Pressure, In. of Water	16.25 16.0 16.25 ...	5.6 5.8 5.8 ...	4.9 5.5 4.7 5.3	2.8 2.4 2.3 2.5	0.6 0.4 0.4 ...	.6 .5 .6 .8	.8 .9 .7 .8	.5 .6 .5 .5	.8 .2 .3 .2
Average	16.1	5.7	5.1	2.5	0.53	.63	.8	.52	.34
Time in Sec. to Reach Max. Gas Pressure	309 300 312 ...	38 38 37 ...	35 35 34 36 35	34 36 39 37 ...	38 38 38 ...	55 53 54 52 ...	35 36 37 38 ...	36 38 39 37 37	40 42 43 41 48
Average	307	37.6	35	36.5	38	53.5	36.5	37.4	42.8

the test cores gave a pressure of 8 in. of water. Gas pressure was reduced to about 4 in. of water by baking for 70 min. at 425 F. See Fig. 12 and Table 7.

The data suggests the use of a test core to check core baking practices in the foundry.

#### Conclusions

Under the conditions of test, namely, 1200 F in a lead bath, the following conclusions are drawn:

1. An increase in the amount of water used to temper a molding sand increases the gas pressure rapidly at the mold face.

TABLE 7—EFFECT OF BAKING TIME ON CORE GAS PRESSURE

Mixture No. 15					
Raw Linseed Oil, lb	..	20	..	..	..
Mogul, lb	..	10	..	..	..
Wedron, lb	..	1800	..	..	..
Silica Flour, lb	..	200	..	..	..
Water, lb	..	60	..	..	..
Baking Temp., F	..	425	..	..	..
Baking Time, min	10	20	40	70	110
Max. Gas Pressure, in. of water	8.1 8.2 7.6	5.3 4.9 5.5	4.8 4.7 4.8	3.4 3.4 4.2	4.0 4.2 3.6
Average	8	5.2	4.8	3.7	3.9
Time in sec. to reach Max. Gas Pressure	41 40 41	44 44 44	44 44 44	46 45 46	48 46 48
Average	41	44	44	46	47

2. The maximum gas pressure in a dried sand is generated very slowly. Several inches gas pressure may be produced in a dried sample of molding sand.
3. Adding seacoal to molding sand causes a rapid rise in gas pressure, for example, 4.3 to 20.5 in. of water pressure as seacoal is increased from 0 to 5 per cent.
4. Fortunately, an increase in mold hardness delays the time at which the maximum gas pressure is created. This delay in time gives the molten metal more time to set and form a chilled skin of metal on casting face reducing the chance of blow defect.
5. The gas pressure at the outer face of a core increases with an increase of core oil used in the core mixture.
6. Additions of cereal binder increases the gas pressure at the outer face of a core rapidly.
7. Increasing baked permeability of a core  $1\frac{1}{8}$  in. in diameter from 11 to 170 reduced the gas pressure from 16.1 to 0.5 in. of water pressure.
8. In compounding core mixtures, one should make studies as to which mixtures will yield the lowest or desired gas pressure. The fineness of sand, grade of core binder, quantity of core binder, baked permeability, size of core, and resulting gas pressure should all be taken into consideration.

## DISCUSSION

Chairman: C. W. BRIGGS, Steel Founders' Society, Cleveland  
Co-Chairman: WERNER FINSTER, American Chain & Cable Co., Reading, Pa.

MEMBER: It was stated that the gas pressure decreased in the Albany type of sand as mold hardness was increased. Was there optimum moisture content in the mold?

MR. DIETERT: The sand was tempered at the workable moisture content. This applies to all of the samples. I do not know why the gas pressure reduced as we increased mold hardness up to 90.

H. H. FAIRFIELD: The moisture content was approximately 6 per cent. It seemed that in the test we ran that the sand generated gas more slowly when it was rammed harder. That might have been due to the fact that we made the test measurement  $\frac{1}{8}$  in. back from the surface and the heat penetrated more quickly with loose rammed sand.

MEMBER: Did your molding sand have relatively low or high pan material?

MR. FAIRFIELD: I think it was a low pan material.

CHAIRMAN BRIGGS: It would be well for the authors to describe what they believe to be the type of defect obtained with high gas pressures.

MR. DIETERT: One particular defect that was eliminated by this study of gas pressure was a blow on an aluminum casting. The casting reached into the cope for a distance of 6 to 8 in. and the metal pressure at this high point in the cope was rather low. Blows were appearing in the cope section of the casting. A green sand core was used in the cope. By reducing the pressure into the green sand core the castings did not show this blow defect or metal agitation.

We had an experience where the metal was agitated due to high gas pressure on the core surface. The metal agitation created a slag-like condition. By changing the gas pressure relief was obtained from this metal agitation.

J. B. CAINE:<sup>1</sup> Were all these tests made in molten lead at 1200 F? Do you not think we can be a little skeptical of the exact position and also of the shape of the curves if we were dealing with, let us say, steel with a rate of heat input of about ten times as much as you have with lead?

MR. DIETERT: The shape of curve would change. In the case of gray iron we have similar curves. We can save time by using a lead bath.

R. E. KERR:<sup>2</sup> I can see how these tests could correlate with cores where the sand core is practically surrounded by metal, but I cannot see a correlation with mold pressures where only a relatively small area of the sand is in contact with the molten metal. It seems that tests similar to these but with the specimen at varying lesser immersion depths might yield results more symbolic of mold pressures.

MR. DIETERT: We endeavored to hold the exposed end of the molding sand sample  $\frac{1}{2}$  in. above the top of the metal bath. One can vary that as much as  $\frac{1}{4}$  in. and the difference is negligible. We have placed pressure tips in molds and find that the gas pressure in the case of molding sand is low. However, from the information given in this paper it will give you an index as to how the gas pressure increases when we vary the ingredients of the material or of the sand. The actual gas pressure that we show from this  $1\frac{1}{8} \times 2$ -in. specimen is not the same value as found in a mold or in a core. However it does show the trend. I think that is all we can hope to obtain from test data contained in this paper. Our work is only a beginning on this subject of gas pressure.

D. C. WILLIAMS:<sup>3</sup> Was your test specimen 3 in. long?

MR. DIETERT: The specimen is 3 in. long. Figure 3 shows details of the gas pressure brass tube. The 2-in. dimension in Fig. 3 should read  $2\frac{7}{8}$  in.

MR. WILLIAMS: Do you use the same ramming equipment as you use for a 2-in. specimen?

MR. DIETERT: The  $1\frac{1}{8}$ -in. sand rammer is used to ram the test specimen.

CHARLES LOCKE:<sup>4</sup> Does the author feel that the magnitude of the pressure against the metal itself is of the same order as that found going out through the core?

MR. DIETERT: We know that heat transfer from the lead bath to the core would be vastly different than, let us say, the heat transfer from steel to the core. The manner of wetting of different metals would not be the same. However it is known from high temperature testing that the heat input into a sand specimen is restricted by the sand itself to a large extent. Thus one might not find too much difference in gas pressure when using different metals for the bath. In case of steel, a thin chilled skin is formed quickly, whereas in the case of iron or lead it is formed slowly.

CHAIRMAN BRIGGS: When you take the effect of mold hardness curve showing 15 to 25 in. of water pressure recorded you certainly would not be interested in some other effects where maximum variation is only about 1 in. of water pressure.

MR. DIETERT: In the case of cores we need much lower gas pressures than for molding sands. Certain casting designs make venting mold or core section difficult. In such cases this gas pressure of the molding sand and cores must be kept very low.

CHAIRMAN BRIGGS: That is a good point.

MR. FAIRFIELD: I might quote a further example in the case of a core that had 7 in. of water pressure. The castings produced with this sand had large blow holes in them. We reduced the gas pressure to 2 in. of water and the blow hole defects disappeared. That is some indication of permissible gas pressure in cores.

<sup>1</sup> Sawbrook Steel Castings Co., Lockland, Cincinnati

<sup>2</sup> Pettibone-Mulliken Corp., Chicago

<sup>3</sup> Ohio State University, Columbus, Ohio

<sup>4</sup> Armour Research Foundation, Chicago



# CONVENTIONAL VS. SALT BATH HARDENING OF CAST IRON CYLINDER LINERS

By

G. M. Lahr\*

## ABSTRACT

*The author compares two salt bath heat treating cycles with the conventional quench and draw treatment. Advantages of the salt bath cycles are herein presented.*

HARDENED ALLOY cast iron cylinder liners (Fig. 1) have been successfully used in the General Motors series 71 Diesel engine for many years. Both laboratory and field tests have shown that the hardening operation greatly increases the wear resistance and consequently the life expectancy of the liner. However, the hardening of this part by conventional oil quench and draw method presents shop problems such as distortion, growth, scaling, and cracking. In an effort to reduce or eliminate these undesirable elements of heat treating, experiments were begun on the adaptation of salt bath quenching procedures to this particular job. The results of these experiments were very encouraging and will very likely result in the adoption of this method of hardening liners for production usage at this plant. It is recognized that the preliminary test results presented here do not represent any great quantity of liners tested and should therefore be interpreted only as a general trend

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of what to expect from the various heat treatments described. Although all test samples were picked at random from many heats of castings, it is possible that greater variations in results will be experienced when more heats are processed.

The specified chemical analysis of the iron now used in all production cylinder liners at the Detroit Diesel Engine Division is as follows:

Element	Content, per cent
Total Carbon.....	3.25 to 3.50
Silicon .....	2.05 to 2.25
Sulphur .....	0.12 Max.
Phosphorus .....	0.25 Max.
Manganese .....	0.55 to 0.60
Copper .....	1.00 to 1.50
Chromium .....	0.55 to 0.80
Nickel .....	0.10 to 0.25

All test liners were made from sand castings that had been normalized at the foundry.

Combined carbon after normalizing is below 0.60 per cent and normally runs about 0.30 to 0.50 per

Fig. 1—Cast iron cylinder liner used in this study.

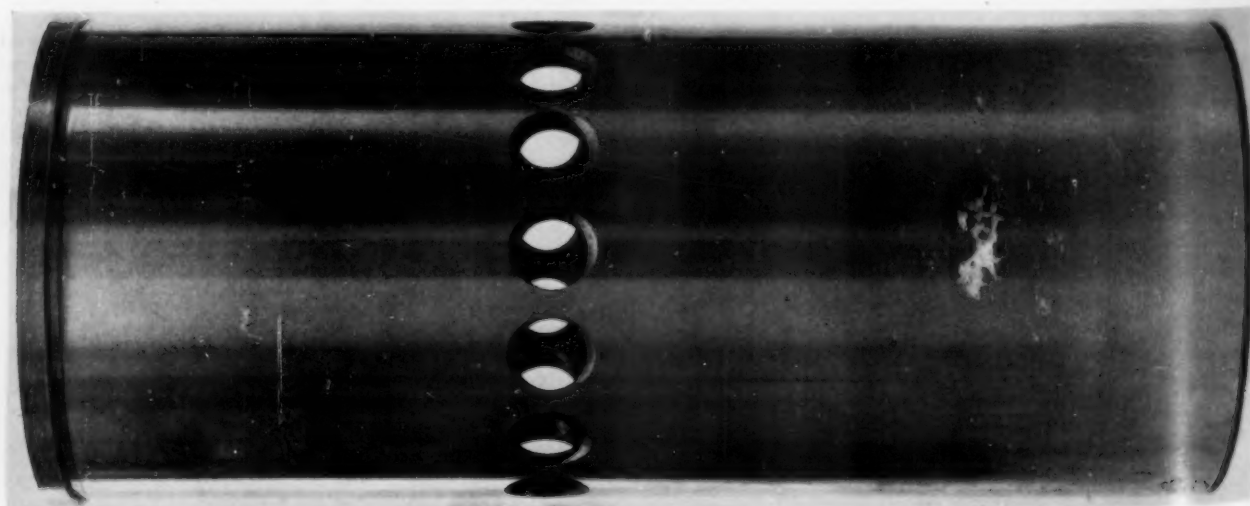




Fig. 2—Typical microstructure of cast iron cylinder liners before hardening.

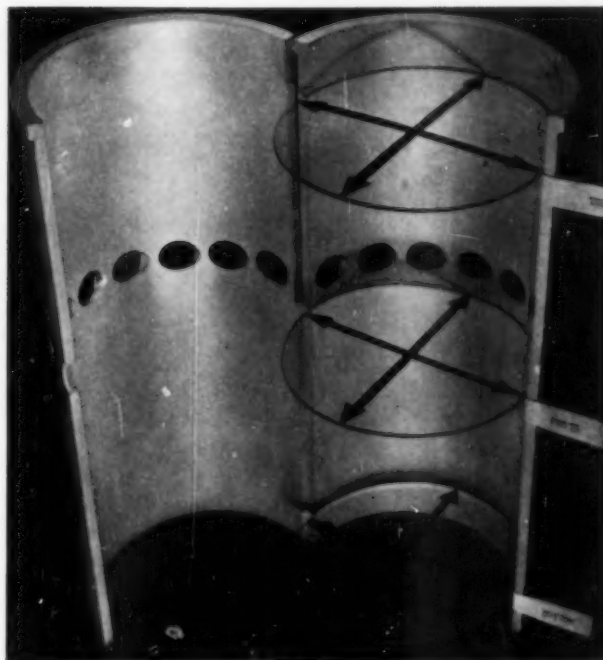


Fig. 3—Photo shows positions where measurements were taken before and after heat treatment.



Fig. 4—Oil quench tank with quenching arbors. Oil was drained before photographing. Normal oil level is approximately 3 in. below "mushroom" top on arbors.

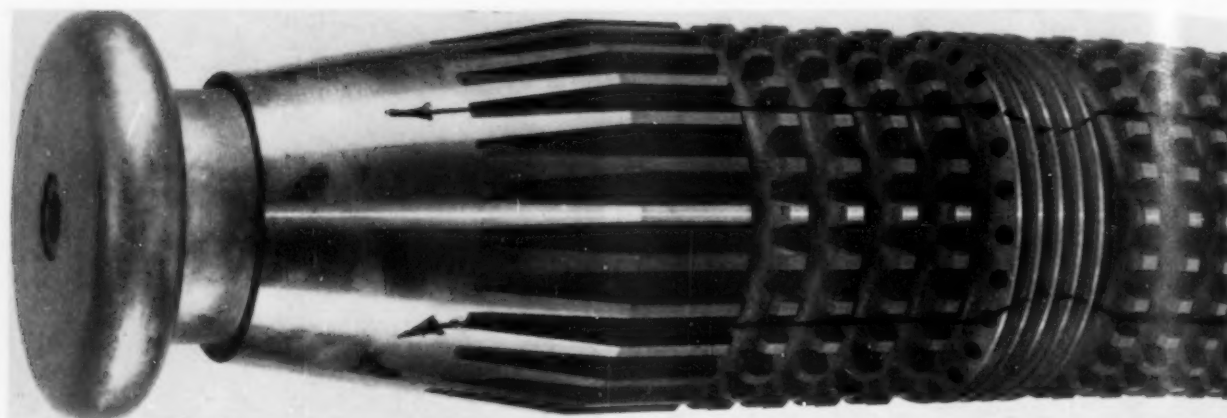


Fig. 5—Arbor used for oil quenching cylinder liners.

cent. The typical microstructure prior to hardening is shown in Fig. 2.

Removal of metal from the inside of the liner after heat treatment is time consuming. Therefore, the primary consideration was to determine growth and distortion of the internal diameter as a result of heat treatment. All measurements made during these tests were taken on the inside diameter before and after heat treatment in accordance with the positions indicated in Fig. 3. Maximum distortion figures were obtained by subtracting the smallest from the largest

of the six measurements after heat treatment. In other words, the distortion figures do not necessarily indicate that the liner is out of round by that much in one horizontal plane, but do indicate the maximum measured variation of the internal diameter size on two axes and three horizontal planes. Maximum growth, on the other hand, was calculated by determining the greatest change in dimension of any single measured diameter during heat treatment. All hardness measurements were taken on the Rockwell "C" scale, since a Brinell test would have caused the liners to be scrapped.

The present production method of heat treating

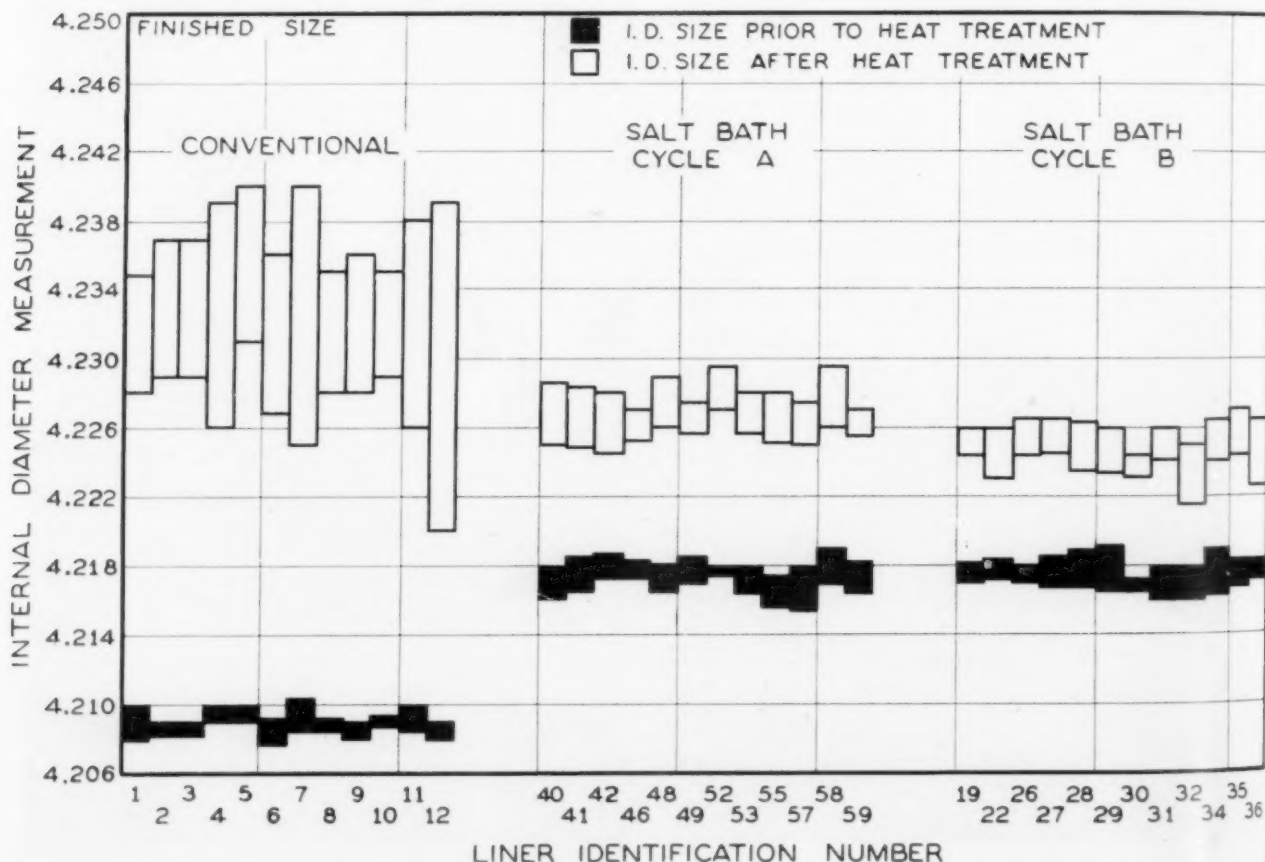


Fig. 6—Pictorial representation of measurements taken of cylinder liners before and after heat treatment by

the quench and draw method and the salt bath method, Cycle A and Cycle B.



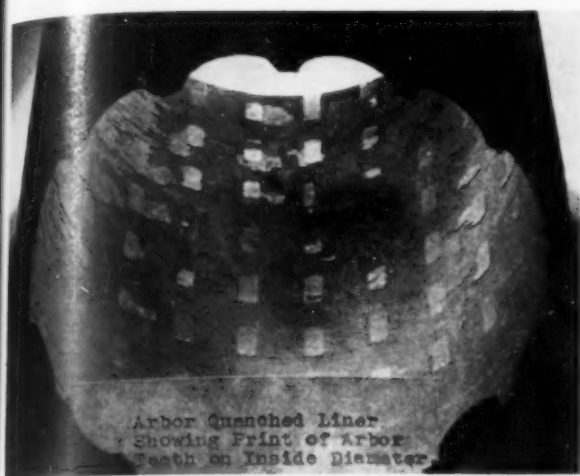


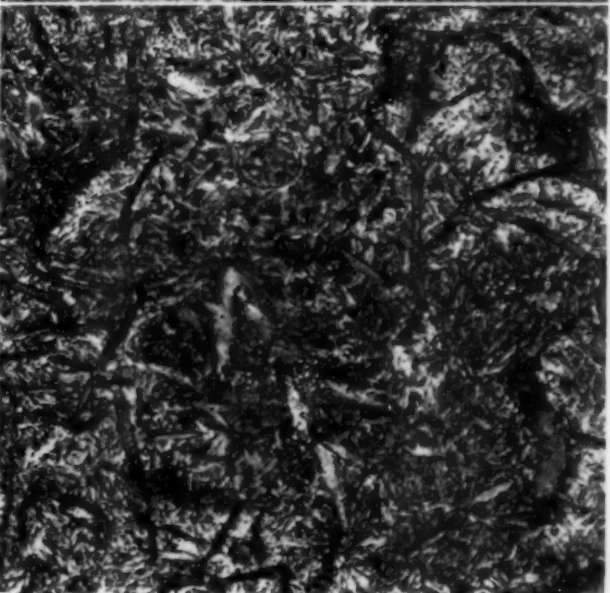
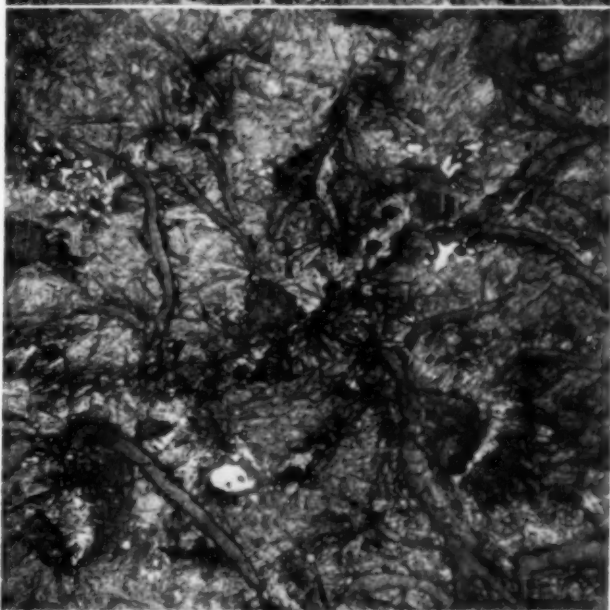
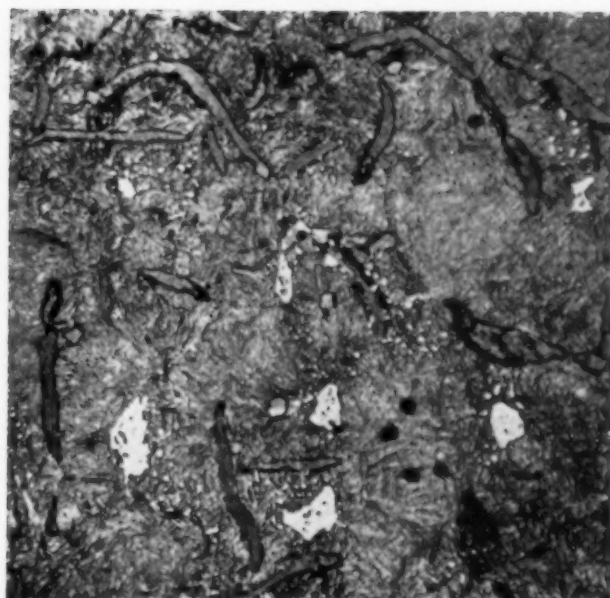
Fig. 7—Arbor quenched liner showing print of arbor teeth on inside diameter.

Typical microstructures produced by heat treatment; Fig. 8 (top) by oil quench followed by 550 F draw; Fig. 9 (center) by salt bath Cycle A; and Fig. 10 (bottom) by salt bath Cycle B. Mag.—500X, nital etch.

cylinder liners involves heating in a gas-fired rotary furnace at 1550 F for 90 min. and then quenching on an arbor submerged in oil at 200 F. Figure 4 shows the quenching arbors at a time when the oil had been drained from the quench tank. The normal oil level in this tank is about 3 in. below the mushroom head on the arbor. The hot quenching oil is pumped through small holes located between the vertical flutes at the base of the arbor and travels upwards until it reaches the port band where it is detoured through holes located adjacent to the liner ports as shown in Fig. 5. From this point, the oil continues upwards in the flutes until it emerges at the top of the arbor. After quenching, the liners are allowed to drain and are then washed and given a 1-hr draw at 550 F. Since the liner was heated originally in a gas-fired furnace without a controlled atmosphere, it requires a grit blasting operation after the draw to prepare it for the grinding operations that follow. Scale thickness has been found to vary from 0.002 to 0.004 in. In addition to removing this scale, the grit blasting operation removes 0.001 to 0.002 in. of stock from the diameter.

Average maximum distortion of 12 production test liners heat treated by this method was 0.0101 in., the range being 0.006 to 0.019 in. A graphical representation on the measurements taken on these liners is presented in Fig. 6A. The average maximum diametrical growth was 0.0279 in., the range being 0.0255 to 0.031 in. All measurements were taken after the grit blasting operation in order to prevent the interference of scale.

The severity with which the liner clamps itself on the arbor before releasing itself and the effect that scale can have on distortion is illustrated in Fig. 7. Note that the scale has adhered in some areas and is gone in others. Assuming that this occurs prior to the time the liner clamps to the arbor, scale could be responsible for a sizeable portion of liner distortion.



As can be seen in Fig. 6A, a liberal amount of stock must be allowed in order to assure that all liners will "clean up" when they are ground. Due to the amount of distortion encountered by this arbor quenching procedure, two internal grinding operations are required for sizing in addition to final honing for proper surface finish. The excessive amount of machining after hardening plus the scrap encountered due to cracks and excessive distortion warranted an investigation of other methods of hardening the liner.

In order to determine the merits of salt bath heat treatment, a test lot of 60 liners was sent to the Ajax Electric Co. Previous experience had indicated that the liners would grow less when salt bath hardened, therefore an additional 0.010 in. was removed from the internal diameter prior to heat treatment. Since no time-temperature transformation diagrams were available for the material being treated, 18 of these liners were used to establish suitable heating and quenching cycles. All heating and quenching was done in molten salt. The two basic types of cycles which were established are listed as follows:

#### Cycle A

Preheat 5 min at 1200 F.  
Austenitize 8 min at 1600 F.  
Quench 20 min at 480 F.  
Cool in air.

Note: The quenching bath was agitated only for the first 3 min. of the total quench time.

#### Cycle B

Preheat 5 min at 1200 F.  
Austenitize 8 min at 1575 F.  
Quench 1 min at 650 F (agitated).  
Cool in air for at least 1 hr.  
Draw at 350 F.

Although cracking is occasionally encountered when oil quenching liners on an arbor, no cracks were found in this test batch of liners. Both distortion and growth were reduced considerably as evidenced in Figs. 6B and 6C and the liners were scale-free. A summary of the results obtained by these two heat treatments is as follows:

#### Cycle A (22 Liners)

Average Maximum Distortion	0.0026 in.
Average Maximum Growth	0.0099 in.
Average Hardness, Rockwell "C"	51.7

#### Cycle B (20 Liners)

Average Maximum Distortion	0.0025 in.
Average Maximum Growth	0.0089 in.
Average Hardness, Rockwell "C"	50.2

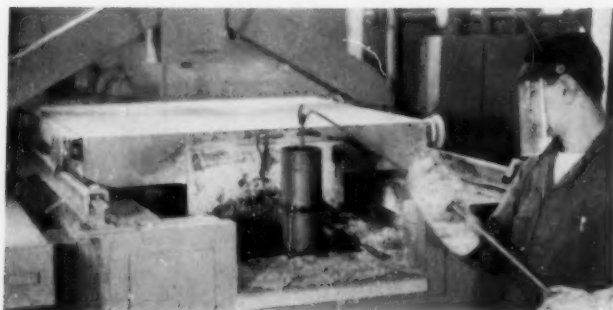


Fig. 11—Salt bath heat treating of cylinder liners.

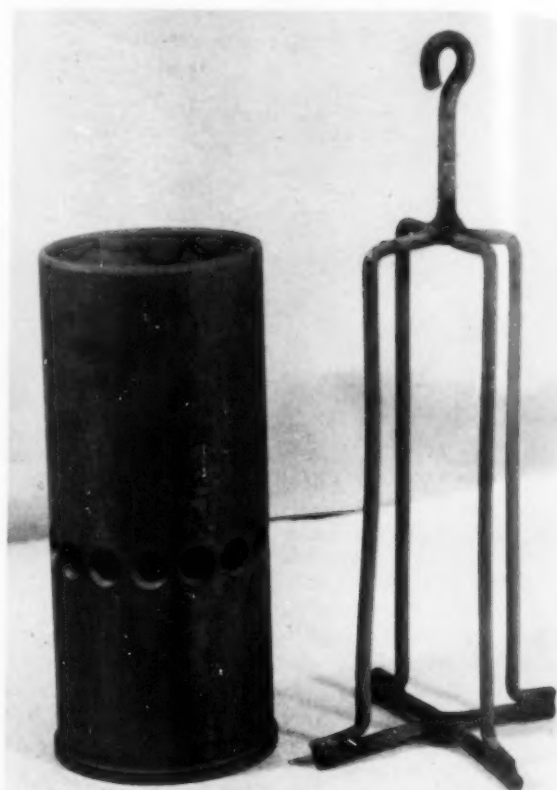


Fig. 12—Hanger for supporting liner in salt bath.

#### Microstructures Produced

The structures produced by oil quenching, and salt bath cycles "A" and "B" are shown in Fig. 8 through 10. These preliminary tests on salt bath hardening of cylinder liners, were carried out in the type of equipment shown in Fig. 11. However, should this system be adopted for production use, a completely conveyORIZED set-up would be utilized. The hanger used for supporting the liner in the various salt baths is shown in Fig. 12.

#### Principles Involved

Since the results obtained by these two types of salt bath heat treatment were so encouraging, it should be interesting to review briefly the principle involved. When a cylinder liner or any other ferrous object that has been heated for the purpose of hardening is suddenly quenched in a cooling medium, such as oil, the surface will cool more rapidly than the interior. Assuming that the liner is cooled rapidly enough to clear the nose of the "S" curve it will transform from austenite to martensite when it passes through a definite temperature range. The location of this temperature range varies with different materials. This transformation is accompanied by a definite volume expansion. Since the surface cools faster, it naturally reaches the transformation temperature range and expands first. Following this, the interior portions of the liner reach the transformation temperature range and expand. The result of this non-uniform transformation is invariably distortion and occasionally cracking. If instead of quenching into oil, the liner is quenched into

TEMPERATURE, °F  
1600-  
1400-  
1200-  
1000-  
800-  
600-  
400-  
200-  
0

Fig. 13—

a molten  
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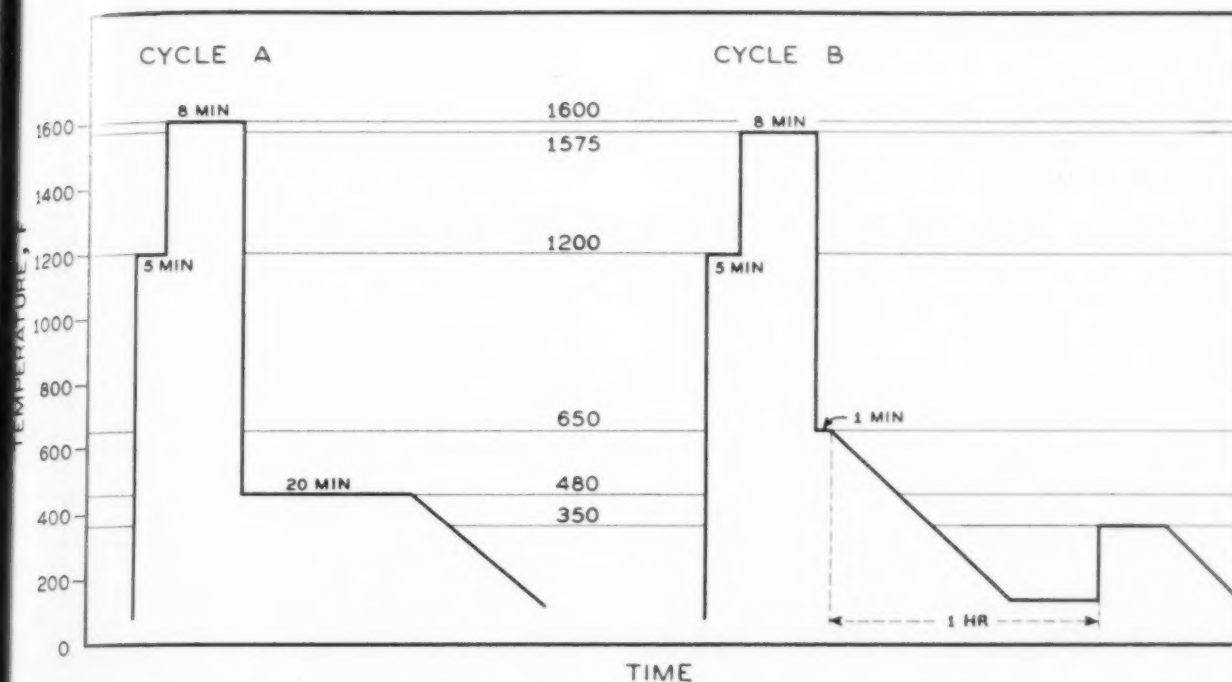


Fig. 13—Representation of salt bath Cycles A and B.

a molten salt bath held at a temperature slightly above the transformation temperature range and held there long enough for the entire cross section to reach the temperature of the bath, uniform transformation can be accomplished by cooling in air. By cooling in air, the temperature differential between the surface and the interior will be minimized and transformation will be reasonably uniform. This process is termed martempering. Austempering, although performed in similar steps, involves different principles. In martempering, the liner hardens during the air cooling period after being withdrawn from the quenching bath. Austempering, on the other hand, involves complete transformation of the austenite while still in the quench bath.

Cycle A as applied to the cylinder liners comprises a combination of the principles of austempering and martempering in that it embodies a rapid quench to 480 F, a holding time of 20 min at 480 F to effect partial transformation to lower bainite, and a final transformation of austenite to martensite during air cooling to room temperature. See Fig. 13. Cycle B is essentially modified martempering treatment in that a rapid quench to 650 F is employed to prevent transformation of austenite to pearlite or high temperature bainite and that the transformation to martensite is effected during air cooling from 650 F.

#### Summary

The advantages of salt bath treatments over the conventional gas furnace heating and arbor quenching procedure can be summarized as follows:

1. The absence of scale eliminates the necessity of grit blasting.
2. Since no arbor is required in Cycle "A" or "B" tracks due to stresses imposed by the shrinking of the liner on the arbor are eliminated.

3. Distortion is minimized, thus reducing precision grinding difficulties.

4. By increasing the internal diameter size prior to heat treatment in accordance with the growth data obtained in these tests, the machining operations following heat treatment can be minimized. Adjustment of the outside diameter dimension prior to heat treatment will also result in a minimum of stock removal after heat treatment.

5. The time required for hardening is reduced considerably.

#### DISCUSSION

Chairman: H. BORNSTEIN, Deere & Co., Moline, Ill.

Co-Chairman: D. A. PAULL, Scaled Power Corp., Muskegon.

GARNET P. PHILLIPS (Written Discussion):<sup>1</sup> Mr. Lahr's paper on hardening of cast iron cylinder liners is of considerable interest to us. Experience had by us confirms Mr. Lahr's findings.

During the latter part of the war we were confronted with the problem of producing hardened, thin, dry-type cylinder liners for heavy duty trucks that would be finished to a maximum out-of-round and taper of 0.001 in. Normal procedure was to insert the thin hardened dry liners in cylinder blocks and to finish-hone to desired tolerance after insertion. The request mentioned above was made with the idea of inserting the liners, already honed to size, in the blocks with no honing required after insertion.

During the war period a non-alloyed low total carbon gray iron was used in these thin liners. Hardenability was satisfactory with the thin sections involved ( $\frac{1}{16}$  in. through  $\frac{5}{16}$  in.). In working out this proposed requirement, checking of I.D. was done in the pot chuck after finish honing, again after removal from pot chuck and again after reinsertion in pot chuck and clamping. The latter step would simulate insertion in machined cylinders in cylinder blocks.

Using the non-alloyed low total carbon gray iron then in use and using conventional heat treatment (oil quench from 1600 F and draw at 400 F) resulted in liners that would not meet the maximum out-of-round tolerance of 0.001 in. For example, the following out-of-round readings were obtained:

ID Finish honed in pot chuck	0.0003—0.0007 in.
ID Removed from pot chuck	0.0020—0.0045 in.
ID Reinserted in pot chuck	0.0007—0.0015 in.

Drawing at higher temperatures (up to 700 F) would give

<sup>1</sup> International Harvester Co., Chicago



liners with a high percentage meeting the out-of-round tolerance but they would fall below minimum hardness specified of 45 R<sub>c</sub>.

Returning to the use of alloy irons, copper-chrome or molybdenum-chrome-nickel, both of which had previously been used by us in regular production, resulted in liners about 90 per cent of which would meet the out-of-round requirement after conventional oil quench and draw.

Hot quenching of the non-alloyed low-carbon iron was tried. The liners were austenized at 1600 F, quenched in salt at 420 F, held for varying lengths of time (from 1 to 7 min) and air cooled. Some were drawn at 450 F for 2 hr. The cycle that resulted in best results from viewpoint of both hardness and minimum out-of-round was to hold for 1 min in salt at 420 F and draw at 450 F. Hardness ranged from Rockwell C 45 to 51 and out-of-round of honed ID after reclamping in pot chuck varied from 0.001 in. to 0.0009 in.

Impact tests on round bars showed that impact resistance was much higher on the hot quenched iron as compared to the conventional oil quenched and drawn iron.

CHAS. A. NAGLER (Written Discussion):<sup>2</sup> Mr. Lahr is to be complimented on his fine paper concerned with the application of isothermal treatment to an industrial application.

The author, in the application of salt bath hardening to the cylinder liners, is using the method of heat treatment that was suggested in the early 1940's by Nagler and Bowdell. It is well recognized that the distortion on hardening can be kept to a minimum if the case and core are at the same temperature prior to the actual austenite-martensite transformation. This method of hardening keeps cracking to a minimum due to quenching stresses. The use of quenching fixtures is discarded whenever possible for they increase the cost of hardening by increased handling cost and maintenance of the quenching fixtures.

Has the author carried out any performance tests to determine what Rockwell C hardness produced by salt bath quenching gave the ideal cylinder life?

The successful application of the isothermal transformation method of heat treatment is wholly based on a knowledge of the isothermal transformation curve of the cast iron involved. This knowledge will definitely fix the maximum section size that can be isothermally heat treated, that is to make sure the section size can be quenched beyond the various upper noses of the transformation curves.

There is much useful information that can be gathered from study of the relationship between quenching rates, in cooling from the martempering temperature. In the application of the isothermal heat treatment to cast iron one must be extremely careful that no ferrite is formed by a slight slowing of the quenching in getting by the upper nose of the curve.

On the basis of microstructure the acicular troostites appear to be ideal for use in a cylinder and that would allow the hardness to be increased to some 55 Rockwell C.

MR. LAHR: We more or less feel that the harder we can get our liner the better. The liners mentioned in the paper were put in engine tests out in the field and we have had no wear data reported. The hardness measurements on these liners averaged around 2 to 3 points Rockwell C higher than we normally get from our conventional method of hardening. However, our draw temperature is also higher than that used in the salt bath process.

As far as the wear resistance of these various structures is concerned, I have no knowledge of data available\* on the wear resistance of the Bainitic type of structure produced by austempering at low quench temperatures. The martensitic structure produced by martempering should, in my opinion, show corresponding wear resistance when compared to a martensitic structure produced by the conventional quench and draw, assuming that the hardness obtained are equal.

CO-CHAIRMAN PAULI: Generally speaking the idea has been "the harder the better", and as the engineering characteristics of the particular liner will take that hardness. We have had cases where they have had to drop back to 40 to 45 Rockwell C due to some particular engine design that just could not stand the increased brittleness of the harder liner.

J. T. MACKENZIE:<sup>3</sup> If you can hang these cylinders on a holder and churn them in hot salt why can you not do the same thing with the oil quench?

MR. LAHR: We tried quenching liners into oil at 200 F without an arbor and found that distortion was far in excess of that produced by quenching on an arbor. It is possible that a martempering type oil could be used at a higher temperature with resulting lower distortion assuming that the quench rate was maintained at a temperature above or near the M<sub>s</sub> point.

F. E. KASCH:<sup>4</sup> Why should you have a 90-min hold on the oil quenching cycles and 8-min on the salt bath cycle? Is the most growth due just to that time differential at temperature?

MR. LAHR: The basic reason for the large difference in heating time is the comparative heating rates and uniformity of heating of the two mediums involved. Molten salt transfers heat to the liner at a much faster rate than is possible in a gas-fired rotary furnace. In our rotary furnace the gas flames are directed well above the tops of the liners, thus heating the upper portions first. In addition to this, the hearth plate is noticeably cooled near the loading door of the furnace. Those two factors make it necessary to leave the pieces in the furnace 90 min in order to insure that the liner is fully at heat throughout its entire length before quenching. Attempts to decrease this time have resulted in scrap due to unequal diametrical growth from top to bottom. On the other hand experiments on cast iron samples 3/16 in. thick by 1 1/2 in. square have shown that full hardness can be obtained throughout the cross-section by quenching after 3 min in molten salt at 1550 F. In addition to its rapid heating rate, molten salt also heats the liner uniformly from all surfaces.

Since this work was done to solve a shop problem, we were forced to contend with many factors that would have been eliminated had the work been more theoretical in nature. For instance, all liners hardened by the conventional method were scaled and had to be grit-blasted before measuring which removed the scale plus some metal. This would then show as growth in the graph. They were also quenched on an arbor which was specifically designed so that they would clamp tightly to it on cooling and release themselves upon transformation. This could account for some "forced" growth. For these reasons, the data given in this paper should not be interpreted as meaning that salt bath methods will always result in less growth. I have been unable to reproduce these variations in growth using 10-in cast iron strips heated in molten salt and in an atmosphere furnace. In fact the samples heated in the molten salt for 10 min grew more than those heated in the atmosphere furnace for 90 min. All of these measurements were taken after quenching in oil.

A. E. SCHUH:<sup>5</sup> Were these salt baths agitated or not?

MR. LAHR: The high temperature baths were agitated only by the stirring action produced as a result of current passing between the immersed electrodes. The quenching bath was mechanically agitated. On the 1-min quench it was agitated for the full length of the quench, while on the 20-min quench it was agitated for only the first 3 min. A pump was used to force salt up through a large vertical pipe immersed in the quench salt. The liner was then quenched into the mouth of this pipe.

CO-CHAIRMAN PAULI: Have you gone into this study to a point where you have been able to find out different rates of different amounts of growth from one casting to another or from one particular type to another, for example, sand cast irons vs. centrifugally cast irons?

MR. LAHR: We have also been working with centrifugally cast liners, and although I have no actual figures like those in the paper to show you, we have found that the centrifugally cast liner grows less than the sand cast even with the same analysis. The centrifugally cast liner I am talking about has an AFA Type D Graphite as compared with the Type A in the sand cast liner. As for the variation of growth from heat to heat, I am inclined to believe that this variation is tied up somewhat with the original combined carbon as received from the foundry. In other words, if you would start out with a combined carbon of 50 per cent, and increase it to 0.65 per cent during hardening, you would get "X" amount of growth. If you would start out initially with a combined carbon around 0.10 per cent, I am inclined to believe that you would get more growth since there is a greater change in combined carbon.

<sup>2</sup> Wayne University, Detroit

<sup>3</sup> American Cast Iron Pipe Co., Birmingham, Ala.

<sup>4</sup> Pettibone Mulliken Corp., Chicago

<sup>5</sup> United States Pipe & Foundry Corp., Burlington, N. J.

# TECHNIQUES OF QUALITY WELDING OF PLAIN CARBON STEEL CASTINGS

By

E. LaGrelus \* and J. D. Wozny \*

## ABSTRACT

The object of this study was to develop welding techniques to produce welds which are sound and possess minimum physical properties required of plain carbon steel castings and to study the metallurgical effects of size of weld, section size welded and subsequent heat treatments on the hardness and microstructure. During the course of this study, it was found that mineral or lime-coated electrodes generally produce sound welds free from pinhole porosity. With the cooperation of electrode manufacturers, electrodes were developed whose weld metal deposit met the minimum required physical properties of plain carbon steel castings.

## Introduction

THE EVER INCREASING DEMANDS for quality and strength in various materials for industrial and engineering applications will necessitate production of castings of increased quality and increased strength. At the present time, some specifications are requiring that weldments made in the repair of steel castings possess physical properties similar to the parent metal and no doubt more specifications will follow this trend in the near future. In addition there is a trend to include radiographic inspection for the quality of weldments in respect to cracking, fusion and soundness (pinhole porosity).

To determine the optimum conditions necessary for sound welds that will meet all the quality and strength demands so that procedures can be set up, it is necessary to know the effect of type of electrode, section size, weld size and the effect of subsequent heat treatment on microstructure and hardness.

The initial step in this work was to find electrodes which would give (1) welds possessing a satisfactory degree of soundness, and (2) would deposit weld metal which would possess certain minimum physical properties. After electrodes were found that met the above two requirements, a metallurgical study was made to study the effects of section size, single and multi-pass welding, and post heat treatment on the hardness and microstructure of the weld metal and heat affected zone. A plain carbon cast steel of the following analysis was used in this investigation:

C-0.28, Mn-0.74, Si-0.57, P-0.024, S-0.040.

It was deoxidized with 2½ lb of aluminum per ton of molten metal.

The above steel is used to meet the following minimum physical properties:

Yield Strength, psi.....	38,000
Tensile Strength, psi.....	70,000
Elongation, per cent.....	24.0
Reduction of Area, per cent.....	36.0

## Procedure for Investigating New Electrodes

The first problem entailed in the investigation of new electrodes was to set up a uniform procedure of welding. After considerable preliminary study and a thorough review of literature, a welder's qualification test plate was adopted (Fig. 1). The plate was to be

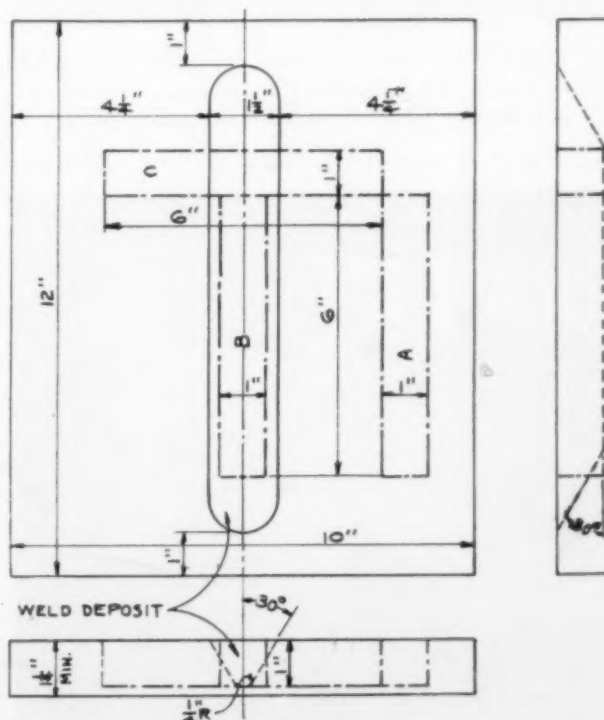


Fig. 1—Welding qualification test specimen.

\* Research Metallurgists, American Steel Foundries, East Chicago, Ind.

TABLE 1—EFFECT OF TYPE OF COATING ON SOUNDNESS OF WELD DEPOSIT

Electrode	Electrode Coating	Radiographic Findings
A (AWS E-6010)	Organic	U
B (AWS E-6012)	Organic	U
C (AWS E-6010)	Organic	U
D (AWS E-6010)	Organic	U
E (Stainless)	Lime	A
F (AWS E-6012)	Organic	U
G (AWS E-6013)	Organic	U
H (AWS E-6020)	Organic	U
I (AWS )	Organic-Asbestos	U
J (AWS E-6015)	Lime	A
K (AWS E-6015)	Lime	A
L (AWS E-6015)	Lime	A

U—Unsatisfactory—Does not comply with Radiographic Standard.  
A—Acceptable—Complies with Radiographic Standard.

Grade of Steel—Plain Carbon in above tests.

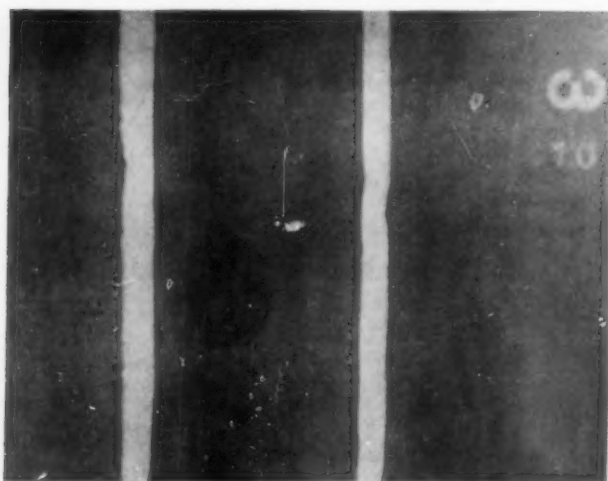
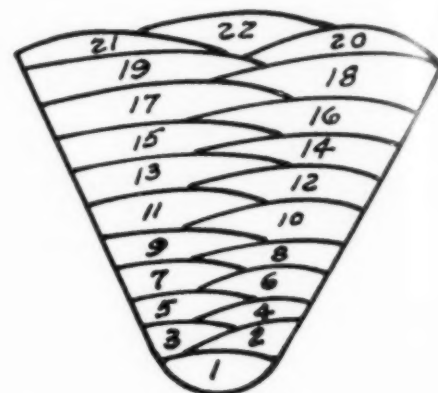


Fig. 2—Recommended standard exograph for soundness of welds. This soundness is typical of weld made with lime or mineral coated rods. Reduced  $\frac{1}{2}$ .



Fig. 3—Exograph showing a typical example of pin-hole porosity found in plates welded with organic or cellulosic coated electrodes. Reduced  $\frac{1}{2}$ .



Pass No.	Size of Electrode	Amperage
1	$\frac{5}{32}$	190
2	$\frac{5}{32}$	190
3	$\frac{5}{32}$	190
4	$\frac{3}{16}$	230
5	$\frac{3}{16}$	230
6	$\frac{3}{16}$	230
7	$\frac{3}{16}$	230
8	$\frac{3}{16}$	230
9	$\frac{1}{4}$	310
10	$\frac{1}{4}$	310
11	$\frac{1}{4}$	310
12	$\frac{1}{4}$	310
13	$\frac{1}{4}$	310
14	$\frac{1}{4}$	310
15	$\frac{1}{4}$	310
16	$\frac{1}{4}$	310
17	$\frac{1}{4}$	310
18	$\frac{1}{4}$	310
19	$\frac{1}{4}$	310
20	$\frac{1}{4}$	310
21	$\frac{1}{4}$	310
22	$\frac{1}{4}$	310

Fig. 4—Method of laying passes, electrode sizes, and amperages used in welding the qualification test plate shown in Fig. 1.

of cast steel of the grade of steel being repaired in production with the groove either cast or machined to size.

A standard exograph of soundness was recommended (Fig. 2) after making a study of production welds. This radiographic standard did not permit any cracks, lack of fusion or pinhole porosity of any greater concentration than shown in the standard.

The physical properties of the weld deposit were obtained from 0.505-in. tensile bars which were removed from the weld plate as shown in Fig. 1, "B." All plates were welded in the normalized condition (1650 F, 2 hr, air cooled) and the physical properties of the welds were obtained in the as-welded condition and after a heat treatment similar to that which the castings received in production.

#### Influence of Electrode Coating

During the preliminary part of this investigation, considerable porosity was noted in welds made in cast steels welded with electrodes having organic



cellulose coatings. This type of defect was less prevalent when the same steels were welded with electrodes having mineral or lime type coatings. Figures 2 and 3 are exographs illustrating these conditions.

This preliminary work stimulated an investigation to determine the effect of coatings on pinhole porosities. Therefore, a study was made of several electrodes that were being used, to determine if this effect was produced by all organic-coated electrodes. All welds were made to eliminate as much as possible the variance in welding technique. Manufacturers' recommended amperage, polarity and special precautions were adhered to so that the best possible welds were obtained with each electrode. Figure 4 is a sketch of the number of passes, sizes of electrodes and amperage used.

Table 1 is a partial list of various grades of electrodes tested, showing AWS designation, type of coating, radiographic findings, etc. From the above study it is noted that only electrodes that are lime or mineral coated would comply with the proposed radiographic standard. This study also revealed that for Grade "B" steels no electrodes normally manufactured at that time would be acceptable. Electrodes J, K, and L included in this table were not available in quantities at the time this study was made.

Since the above study indicated that there were no commercial electrodes available for welding of Grade "B" steel that would produce the desired soundness or the required minimum physical properties of plain carbon steel, the authors planned a long range program to obtain electrodes that would:

1. Produce welds which would comply with the proposed radiographic standard of soundness.
2. Produce welds in which the weld deposit would have the minimum physical properties of the parent metal.

3. Have good welding characteristics, so that uniform welds could be produced by the welders in the plants.

The authors discussed this problem with several electrode manufacturers advising them of their revised specifications, showing them exographs of welds made with their electrodes, and discussing with them their proposed studies and investigations.

#### Experimental Results—

##### Physical Properties of Weld Metal

One electrode manufacturer proceeded to make sample electrodes almost immediately and later several others cooperated. The first manufacturer submitted a total of eight types of experimental electrodes before an electrode was found which was finally adopted and placed in use at the authors' various plants. The authors supplied all the necessary steel for the manufacturing investigation. They also made all the final experimental welds and performed all the X-raying, heat treatment and physical testing.

Table 2 lists in chronological order the physical properties and chemical composition of weld metal of the various experimental electrodes submitted by one manufacturer during the development of a satisfactory electrode for welding of plain carbon cast steel. Each of the electrodes listed was used to weld the groove of the plate shown in Fig. 1. After welding, all plates were machined smooth before X-raying to remove any false indications during radiographing. No physical properties were obtained from any weldment unless the radiographic standard of soundness was met. All physical properties listed are from all-weld-metal longitudinal specimens removed as shown in Fig. 1.

TABLE 2—PHYSICAL PROPERTIES AND CHEMICAL COMPOSITION OF WELD METAL FROM EXPERIMENTAL ELECTRODES SUBMITTED BY ONE MANUFACTURER IN THE DEVELOPMENT OF A SATISFACTORY ELECTRODE FOR PLAIN CARBON STEEL (LONGITUDINAL SPECIMENS)

	Electrode Code	Yield Strength, psi	Tensile Strength, psi	Elongation, Per Cent	Reduction of Area, Per Cent	C	Mn	Si	P	S	Ni	Cr	Mo
1.	AW	58,000	69,000	32.1	74.2								
	N	41,100	57,200	44.5	72.6	.08	.52	.11	.030	.025	..	..	..
2.	AW	65,500	75,000	30.0	64.7								
	N	43,000	61,000	39.0	74.0	.10	.52	.27	.009	.032	..	..	.40
3.	AW	85,500	103,000	24.0	58.2								
	N	51,000	78,000	31.0	59.8	.12	.68	.22	.010	.031	..	.90	.45
4.	AW	74,500	86,300	28.0	62.3								
	N	47,700	70,300	31.5	73.5	.14	.70	.36	.019	.022	.12	..	.52
4A.	N	45,500	67,300	39.0	74.9	.11	.65	.32	.014	.026	..	.02	.40
5.	AW	82,000	93,500	20.0	50.0								
	N	44,800	77,400	28.8	68.8	.17	.74	.43	.012	.026	.18	.25	.39
6.	AW	87,300	101,000	21.0	49.1								
	N	43,000	74,500	30.5	56.2†	.18	.73	.42	.014	.024	.19	.41	.39
7.	AW	80,000	95,700	23.5	66.8								
	N	44,000	76,500	37.5	73.9	.13	.71	.40	.014	.025	..	.41	.48

N—Normalized after welding 1650° F for two hours and air cooled.  
All bars aged at 400° F for 16 hours and air cooled.

Defect in weld metal  
AW—As welded

Electrodes 1 and 2 produced welds which did not meet the minimum tensile strength in the normalized condition, while electrode 3 although it possessed satisfactory physical properties in the normalized condition, was thought to have exceedingly high tensile strength in the as-welded state. Electrode 4 met the physical properties both in the as-welded and normalized condition and was recommended to the plants for the qualification of their welders. Electrode 4A was a sample from the manufacturers' production shipment to the plants. Because of the low tensile properties obtained after normalizing, each plant proceeded to qualify their shipments. The following table shows the results obtained:

CHEMICAL COMPOSITION									
Plant	C	Mn	Si	P	S	Cr	Ni	Mo	
1	.11	.65	.32	.014	.026	.02	.54	.40	
2	.10	.50	.22	.015	.010	.05	..	.46	
3	.13	.66	.29	.014	.022	.03	..	.46	

LONGITUDINAL PROPERTIES AFTER NORMALIZING				
Plant	Yield Strength, psi	Tensile Strength, psi	Elongation, Per Cent	Reduction of Area, Per Cent
1	44,600	66,000	37.0	70.9
2	45,000	68,500	34.0	72.5
3	42,500	67,000	38.0	74.8

Because of the low tensile properties, the electrode manufacturer submitted two more electrode samples, 5 and 6. Both of these electrodes complied with the revised specification.

Sample 7 was a sample of electrodes received by the plants from the manufacturer's production shipment and was also satisfactory. It is now being used in the repair welding of production castings.

#### Transverse and Longitudinal Physical Properties of Weld Metal

We have included in Table 2 only the longitudinal all weld-metal specimens in reporting the physical properties although the transverse weld metal properties were also studied. However, the weld metal in the transverse pull bars was not 2 in. wide, so that in testing the bar, the weld metal, the heat-affected zone and the unaffected parent metal within the 2-in. gauge mark, were all tested simultaneously. This brought a complexity into the testing. Following are the transverse physical properties of the weld metal from electrodes 4, 5 and 6, the longitudinal properties of which are included in Table 2.

TRANSVERSE PHYSICAL PROPERTIES				
Electrode No.	Yield Strength, psi	Tensile Strength, psi	Elongation, Per Cent	Reduction of Area, Per Cent
4 (AW)	52,700	83,400	18.0	38.8
(N)	46,300	71,300	24.0	67.7
5 (AW)	49,000	81,700	15.5	50.8
(N)	41,000	71,000	27.0	71.6
6 (AW)	48,000	82,300	16.0	48.6
(N)	42,000	70,500	37.0	71.9

AW—as welded state.  
N—after normalizing at 1650 F for 2 hr. and air cooled.

It will be noted that in some cases the transverse elongation is much lower than the longitudinal elongation in the as-welded and normalized condition. The as-welded reduction of area and yield strength also run lower in the transverse properties.

To study this effect further, transverse pull bars were removed from welds of varying widths (Fig. 5) so that 1/2-in., 1-in., 1 1/2-in. and 2-in. widths of weld metal were obtained within the 2-in. gauge mark. The following physical properties were obtained from these bars in the normalized condition:

	100% Parent Metal	Width of Weld Metal in 2-in. gauge mark			
		1/2 in.	1 in.	1 1/2 in.	2 in.
Yield Strength, psi	50,100	44,000	42,100	42,100	41,000
Tensile Strength, psi	80,000	73,800	71,500	71,000	70,000
Elongation, Per Cent	30.0	23.0	27.0	28.0	30.0
Reduction of Area, Per Cent	54.7	69.0	71.2	70.8	72.5

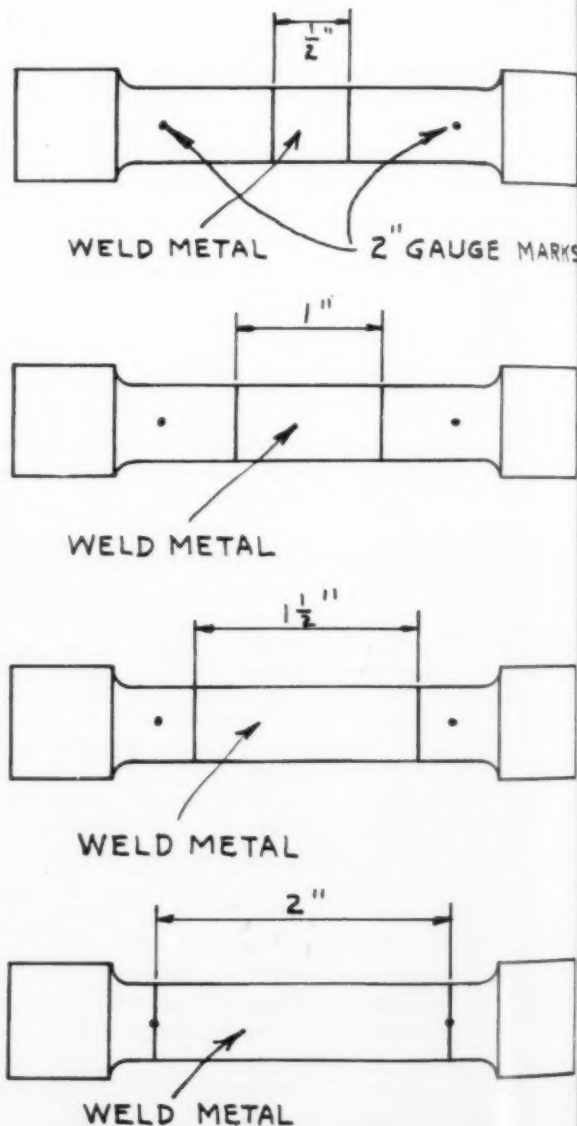


Fig. 5—Sketches showing position of welds in tensile test bars used to study effect of the width of the weld on transverse ductility properties.

The elongation increased with the width of the weld metal within the 2-in. gauge mark until it compared with the parent metal at the 2-in. width. Observing the above results, one would say that since the tensile strength of the weld metal is lower than that of the parent metal, practically all the elongation takes place in the weld metal. To obtain the elongation value of the weld metal, the elongation would have to be based on the width of the weld in the pull bar and not on the entire 2 in. This would increase the elongation value of the transverse weld metal. However, the same phenomenon (that of low elongation values in the transverse pull bar) was noticed in cases where the tensile strength of the transverse weld metal was equal to or higher than the parent metal.

#### Procedure for Metallurgical Investigation

Three cast plates in the normalized condition were prepared for welding by chipping "U" grooves of the following dimensions:

Size of Plate Inches	Small Groove Inches	Large Groove Inches
$\frac{1}{2} \times 9 \times 11$	$\frac{1}{4} \times \frac{1}{2} \times 1$	$\frac{3}{8} \times 1 \times 2$
$1 \times 9 \times 11$	$\frac{1}{2} \times 1 \times 2$	$\frac{3}{4} \times 2 \times 4$
$2 \times 9 \times 11$	$1 \times 2 \times 4$	$1\frac{1}{2} \times 4 \times 8$

The size of the small and large grooves was arbitrarily set to approximate minor and major welds in the plates of the three different thicknesses. The

above grooves were then welded (without preheating) with electrode 7 ( $\frac{3}{16}$ -in. diameter, Table 2) which was previously found to give a satisfactory, sound weld in which the weld metal deposit met the minimum required physical properties of plain carbon steel. Three transverse sections were removed from each weld and were given the following post heat treatment:

Section No. 1—no treatment.

Section No. 2—1050 F, 2 hr, air cooled

Section No. 3—1650 F, 1 hr, air cooled

For additional information sections from the 2-in. minor and major welds were tempered at 1150 F, 1250 F, 1300 F and 1350 F. A hardness survey was then made across the weld and heat-affected zone and specimens were removed for microscopic investigation.

TABLE 4—BRINELL HARDNESS SURVEY OF ONE-PASS WELDS MADE IN PLAIN CARBON STEEL

	As Welded	1150 F, 1 hr, A.C.	1250 F, 2 hr, A.C.
$\frac{1}{2}$ -in. Plate			
Weld	222-230	228-240	228-234
H.A.Z.	216-220	176-185	159-162
P.M.	159-162	159-162	159-162
1-in. Plate			
Weld	222-240	240-252	234-246
H.A.Z.	228-234	180-190	165-169
P.M.	159-162	159-162	159-162
2-in. Plate			
Weld	256-270	234-240	228-240
H.A.Z.	235-250	190-195	169-176
P.M.	159-162	159-162	159-162

TABLE 3—BRINELL HARDNESS SURVEY OF WELDS MADE IN THE PLAIN CARBON STEEL

	As Welded	Normalized 1650 F, 2 hr, Air Cooled	1050 F, 1 hr, Air Cooled	1150 F, 1 hr, Air Cooled	1250 F, 1 hr, Air Cooled	1300 F, 1½ hr, Air Cooled	1350 F, 1 hr, Air Cooled
$\frac{1}{2}$ -in. Major							
Weld	176-228	135-141	180-216	176-216	169-205	162-205	165-200
H.A.Z.	185-210	156-153	180-216	169-176	159-169	153-162	165-169
P.M.	159-162	159-162	159-162	153-156	147-150	144-147	153-156
$\frac{1}{2}$ -in. Minor							
Weld	176-210	139-144	176-216	172-216	169-216	165-205	162-200
H.A.Z.	195-205	159-162	185-210	169-176	162-176	159-172	172-176
P.M.	159-162	159-162	156-165	153-156	147-150	144-147	153-156
1-in. Major							
Weld	159-195	137-153	159-195	.....	.....	.....	.....
H.A.Z.	169-185	165-150	165-185	.....	.....	.....	.....
P.M.	159-162	159-162	159-162	.....	.....	.....	.....
1-in. Minor							
Weld	210-216	150-159	204-222	.....	.....	.....	.....
H.A.Z.	205-210	153-162	195-205	.....	.....	.....	.....
P.M.	159-162	159-162	156-162	.....	.....	.....	.....
$\frac{1}{2}$ -in. Major							
Weld	195-228	156-162	195-234	.....	.....	.....	.....
H.A.Z.	195-205	156-162	176-190	.....	.....	.....	.....
P.M.	159-162	159-162	159-162	.....	.....	.....	.....
$\frac{1}{2}$ -in. Minor							
Weld	205-228	159-169	200-234	.....	.....	.....	.....
H.A.Z.	195-200	159-162	180-205	.....	.....	.....	.....
P.M.	159-162	159-162	159-162	.....	.....	.....	.....

H.A.Z.—Heat-affected zones.

P.M.—Unaffected parent metal.



## Experimental Results

## Hardness Survey

The results of the hardness survey are summarized in Tables 3 and 4. The hardness readings of all the one-pass welds have been combined to form the curves shown in Fig. 6. In the as-welded state, the maximum hardness of the weld and the heat affected zone increased with the section size of the plate, the hardness of the weld being 230 Brinell in the  $\frac{1}{2}$ -in. plate, 240 Brinell in the 1-in. plate and 270 Brinell in the 2-in. plate. The hardness of the heat-affected zone increased similarly from 200 Brinell to 230 and 250 Brinell. The increase in hardness with the section size is due to the increased cooling rate in the heavier sections.

As is shown in Fig. 6, tempering of the one-pass welds at 1150 F for 1 hr had little effect on the hardness of the welds, although the hardness of the heat-affected zone was reduced to 180 to 190 Brinell. Normalizing at 1650 F for 1 hr reduced the hardness of the weld and heat-affected zone to the hardness of the unaffected parent metal.

The effect of multi-pass welding is shown in Fig. 7. The hardness in the weld and heat-affected zone varies from the bottom to the top of the weld, being lowest at the bottom of the weld. The decrease in hardness at the bottom of the weld varies according to the number of passes in the weld. In the  $\frac{1}{2}$ -in. plate minor weld where only two passes were made, the bottom of the weld is 205 Brinell. The hardness

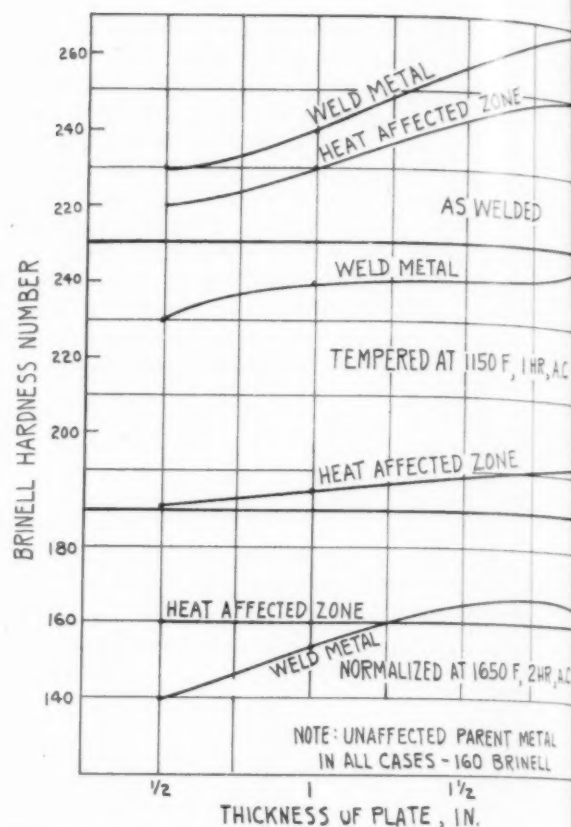


Fig. 6—Effect of section size on hardness of weld and heat affected zone of one-pass welds in Grade "B" steel.

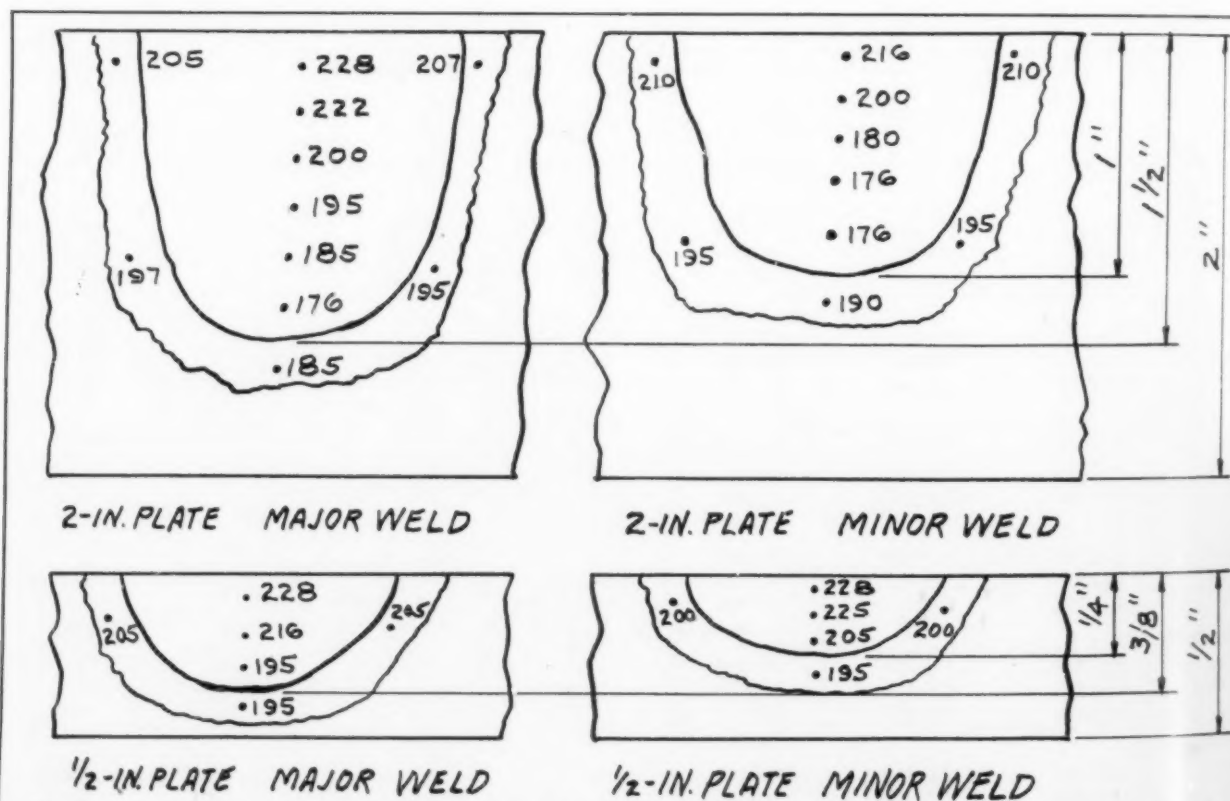


Fig. 7—Cross section hardness survey in as-welded condition of major and minor welds in  $\frac{1}{2}$ -in. and 2-in. plates, Grade "B" steel.

drops to 95 Brinell in the 1/2-in. plate major weld where three passes were made, and drops further to 76 Brinell in the 2-in. plate minor weld where eight passes were made. The hardness in the heat-affected zone followed the same pattern. This decrease in hardness in multi-pass welding is due to the renormalizing effect of succeeding passes. The top one or two passes in multi-pass welds, therefore, will have a higher hardness than the lower passes, and its hardness will approach that of the single-pass welds.

Physical properties of welds in the as-welded state, as reported by electrode manufacturers, are the properties of the weld in the location from which the test specimen is removed. The properties of the top of multi-pass welds and especially of single-pass welds will have higher tensile strength, lower ductility.

As in single-pass welds, tempering of multi-pass welds reduced the hardness of the weld metal only slightly while in the heat-affected zone the reduction of the hardness was more pronounced as shown in Table 5. Even after tempering at 1250 F there is a hardness differential of 60 to 70 Brinell points between the weld and parent metal. The reported physical properties of welds in the stress-relieved state (which would be equivalent to tempering at 1150 F to 1250 F) represent approximately the middle of the multi-pass weld. The tensile strength of the top of the multi-pass welds or single-pass welds in the stress-relieved or tempered state would be 25,000 to 30,000 psi higher, and the ductility would be correspondingly lower.

#### Microstructure Studies

Figure 8 illustrates the changes in microstructure of the heat affected zone brought about by welding. The weld metal is on the extreme left and adjacent to it is an area about 0.02 in. wide consisting of a very coarse pearlitic structure. This area must have reached temperatures in excess of 2100 F in order to form such coarse grains in an aluminum killed steel. The coarse structure is followed by an area of a very fine grain structure which was formed when the metal was

TABLE 5—EFFECT OF TEMPERING ON HARDNESS OF MULTI-PASS WELDS

	As Welded	1050 F. 1 hr, A.C.	1150 F. 1 hr, A.C.	1250 F. 1 hr, A.C.
Brinell Hardness Number Near Top of Weld				
2-in. Major Weld	228	216	216	210
2-in. Minor Weld	216	216	216	216
1-in. Major Weld	228	228	222	222
1-in. Minor Weld	228	228	228	222
Unaffected Parent Metal	162	162	156	150
Brinell Hardness Number Near Top of Heat-Affected Zone				
2-in. Major Weld	207	210	176	169
2-in. Minor Weld	210	210	176	176
1-in. Major Weld	205	200	170	170
1-in. Minor Weld	200	195	170	170
Unaffected Parent Metal	162	162	156	150

heated to temperatures just above the critical and followed by rapid cooling. Next is an area of spheroidized structure which formed at temperatures just below the lower critical or at approximately 1350 F. At the extreme right of Fig. 8 is the unaffected parent metal structure.

The above described pattern was found around the entire weld in the single-pass welds. In multi-pass welds, due to the grain refinement caused by the renormalizing effect of succeeding passes, the above described pattern will be found only adjacent to the top one or two passes. The initial passes and the adjoining heat-affected zone will be refined towards the bottom of the weld (Fig. 9).

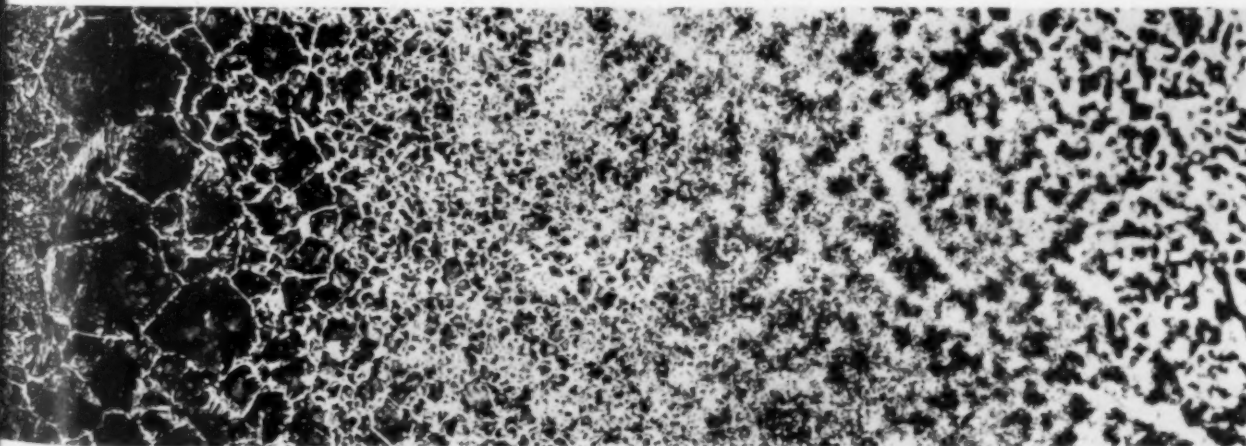
Tempering has no effect on the microstructure obtained by welding of the plain carbon cast steel. Normalizing, as is shown in Fig. 10, will refine the entire heat affected zone.

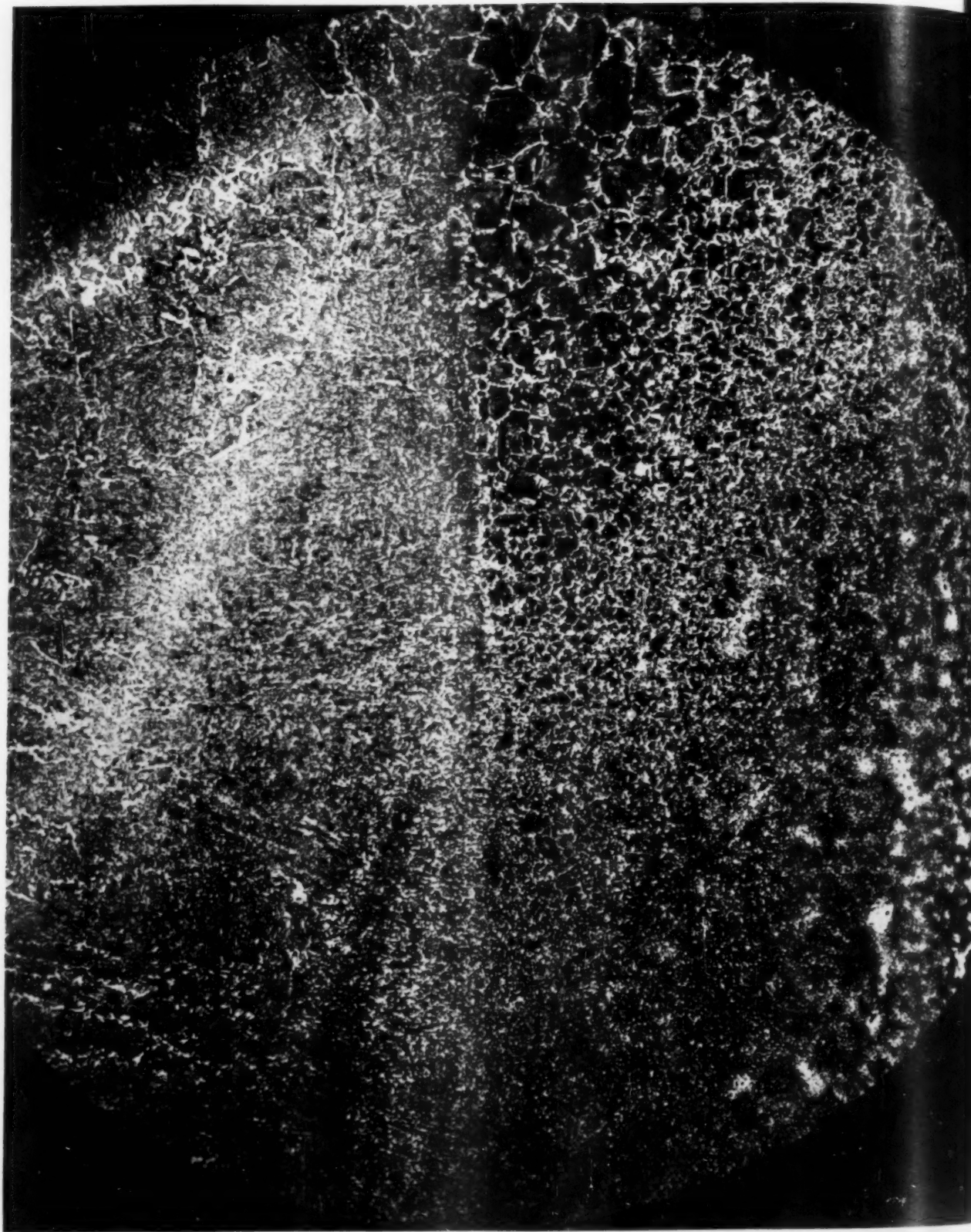
#### Significance of Results in Establishing Shop Procedure for Quality Welding

It has been shown that with proper electrodes sound welds can be made which will meet the minimum required physical properties of plain carbon steel.

The metallurgical study has shown that plain carbon cast steel can be welded without preheating up to at least the maximum section used in this study.

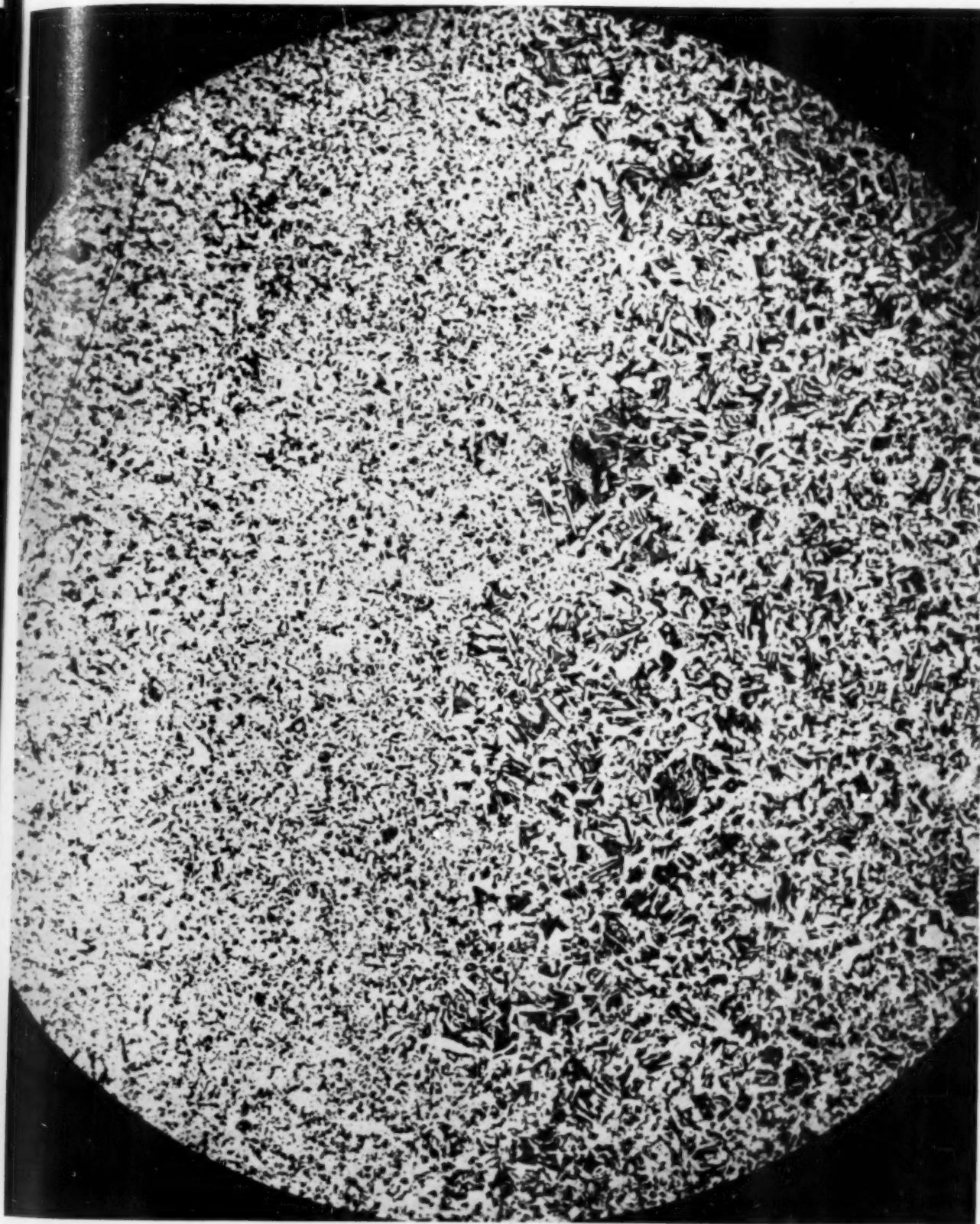
Fig. 8—Microstructure of heat affected zone of a one-pass weld made in plain carbon steel. The weld metal is on the extreme left. Mag. 100X, nital etch.





*Fig. 9—Grain refinement effect of multi-pass welding in plain carbon steel. Weld metal is on the left. Note the grain refinement in the initial passes shown at the bottom of the photomicrograph.*





*Fig. 10—Effect of normalizing on the heat-affected zone and weld metal of a multi-pass weld in plain carbon cast steel.*

Normalizing after welding restores the original structure in the heat-affected zone and reduces the hardness in the weld and heat-affected zone to the hardness level of the parent metal. Tempering up to 1250 F reduces the hardness of the welds only slightly. The hardness of the heat-affected zone begins to drop considerably at 1150 F.

Since tempering has little effect toward reducing the hardness of the weld metal, a hardness differential of up to 110 Brinell points exists between the weld metal and the parent metal (single-pass weld in 2-in. section). What effect will this hardness differential have on machining? In an experiment a one-pass weld was made in a slotted 1-in. diameter bar. A hardness differential of 84 Brinell points existed between the parent metal and weld, but no machining difficulties were encountered when the bar received the most drastic machining possible at the laboratory.

The results of this investigation show that in respect to hardness and microstructure, minor welds (especially one-pass welds) are more critical than major welds. Yet, at the present time many procedures for post heat treatment of welds are set up on the belief that major welds are much more critical than minor welds.

### Conclusions

Within the limits of this investigation the following conclusions may be drawn:

1. Organic coated electrodes are not suitable for repair welding of plain carbon steel castings because of their tendency to produce pinhole porosity.
2. Lime or mineral-coated electrodes generally produce welds free from pinhole porosity.
3. Electrodes have been developed which will meet the minimum required physical properties of plain carbon steel.
4. Tempering of welds in plain carbon steel has only a slight effect on the hardness of the weld metal or the structure of the heat-affected zone.
5. From the standpoint of hardness and microstructure, minor welds in plain carbon steel are just as critical or more critical than major welds.
6. Plain carbon steel can be safely welded without any preheating up to the maximum section used in this study.

### Acknowledgments

The authors gratefully acknowledge the cooperation of our Engineering Department and our various plants as well as some of the electrode manufacturers during some of the phases of this work. Acknowledgment is also due to G. A. Lillieqvist, Research Director, for his encouragement and suggestions during this work and to American Steel Foundries for their permission to publish these results.

### DISCUSSION

*Chairman:* J. F. RANDALL, Ford Motor Co., Dearborn, Mich.

*Co-Chairman:* D. C. ZUEGE, Sivyer Steel Castings Co., Milwaukee.

*H. C. CAMPBELL<sup>1</sup> (Written Discussion):* The authors are to be congratulated on a scholarly and well-documented paper. The welding fraternity will welcome this contribution to our knowledge of application of arc welding.

The authors' procedure is an excellent model for foundries and fabricators to follow. They first established specific welding procedures and specific heat treatments which were to be used

on all tests. They also predetermined the radiographic and tensile standards which all welds were to meet. Only then did they proceed to vary the composition of electrode coatings and weld metals in the search for satisfactory production welding electrodes.

Persons familiar with the work being carried on by the American Steel Foundries Research Laboratories will probably not be misled by the authors' terminology for the electrode coatings used. However, a cursory reading of the tables and text, in which the successful electrodes are called "lime-coated," and the authors' statement that "only electrodes that are lime or mineral-coated would comply with the proposed radiographic standard" might lead the reader to suppose that titania and lime-titania coatings were also tested and were found wanting. This is surely not the authors' intention. The approved designation for the electrode coatings now being applied to this low-alloy high-tensile type of electrode is "hydrogen-controlled." These coatings are mineral type, as the authors correctly report, but in practice they may be lime, lime-titania, or titania type. The important point is that they are low in hydrogen, not that they contain lime or titania. The title "hydrogen-controlled" is more inclusive and more informative, as it suggests the fundamental reason for the elimination of porosity and micro-fissures.

Attention is called to the fact that the authors do not place primary reliance for acceptance or rejection of an electrode on the transverse tensile tests, even though this test bar is included in their specification. This specimen gives valuable information as to the ultimate strength of the welded joint and the location of the failure, but attempts to draw any conclusions from the yield strength or ductility data should be discouraged. The yield strengths measured on transverse specimens are not reliable because different metallurgical structures and properties are being tested simultaneously and the true yield is impossible to determine. Likewise the ductility of a transverse specimen is dependent upon the relative yield and ultimate strengths of the plate and weld metal, and varies with the weld design, as the authors have shown. Messrs. LaGrelus and Wozny are using the transverse specimen in the only correct manner, by establishing their own norms and comparing all results with previously tested transverse specimens.

*Mr. WOZNY:* I want to thank Mr. Campbell for the compliments. In regard to terminology, we agree some improvement could be made. We agree with Mr. Campbell's suggestion that "hydrogen-controlled" would be a more descriptive term for this type of electrode.

*J. B. CAINE<sup>2</sup> (Written Discussion):* The authors are to be congratulated for their very thorough investigation of the effect of the welding rod on the quality of the weld. Although it is out of the scope of the paper, it should be mentioned that inorganic coated welding rods have other advantages than producing welds free from porosity.

In fact it is the writer's impression that this type of coating was developed for the higher carbon and alloy steels because of its ability to produce welds free from underbead cracking. Many of these high carbon and alloy steels can be welded without preheating using inorganic coated rods, whereas they must be preheated if an organic coating is present on the rod. The elimination of the preheating operation will, for these steels, more than offset the higher cost of the inorganic coated welding rods.

*Mr. WOZNY:* We are aware of that fact. We had done extensive work on our high-tensile Grade C steel where we had welded different type welds in various sections at room temperature, 40 F and 300 F. With hardness in the heat-affected zone as high as 500 BHN in 2-in. sections we did not run into any underbead cracking whatsoever.

*D. L. MATHIAS<sup>3</sup> (Written Discussion):* For many structures in the field of engineering, cast components have demonstrated their superiority over those produced by other methods of fabrication. Messrs. LaGrelus and Wozny undertook to solve a difficult assignment and are to be complimented on the results presented in their paper. So successful has been their work that little, if anything, can be added in discussion. While the repair of casting defects appears to be a simple job once the proper procedure has been worked out, many man-hours of painstaking investigation are required to develop that procedure to insure consistently satisfactory results under normal production conditions.

<sup>1</sup> Sawbrook Steel Castings Co., Lockland, Ohio.

<sup>2</sup> Metal & Thermit Corp., Rahway, N.J.

<sup>3</sup> Arcos Corp., Philadelphia.



It is interesting to note that the lime-ferritic or low-hydrogen electrode met the rigid requirements involved in the authors' investigation, whereas electrodes of the conventional types failed. From the data presented, it will be seen that the desired results were obtained by depositing weld metal of an analysis differing from that of the casting, the difference being due primarily to the advantages gained by keeping the carbon content of the weld down and compensating for this downward adjustment by small additions of chromium and molybdenum. In this manner, the physical properties of the weld deposits were made to match those of the casting or parent metal after the heat treatment normally used for castings of Grade B steel.

Thus, beyond the freedom from porosity, cracking and slag inclusion for which the lime-ferritic electrode is noted, it permits a wide range of weld composition to meet almost any requirement. An additional advantage of this type electrode is that it produces a slag capable of fluxing any sand which may be inadvertently experienced in the repair of defects or the use of castings in composite weldments.

D. C. SMITH<sup>4</sup> (Written Discussion): The authors of this paper should be commended on the thorough way in which they treated the subject.

Many institutions have set up qualification tests that must be met by the electrode producer which may have been copied from other specifications and which may or may not have any direct bearing on the use to which the electrode is made. This has been particularly true of the iron and steel casting industries which in the past have been compelled to adapt electrodes designed for fabrication for their casting repair.

Since the advent of the low-hydrogen type electrode, which has the inherent characteristic desired by the casting industry, we believe a test procedure requirement of the type described in this paper is timely and will enable them to take advantage of this new development and to focus further improvements and design for use on their own products.

Some characteristics of the low-hydrogen type electrode which make it desirable for casting repair are its:

1. Property of taking excessive high heats without deleterious effects on the weld metal such as cracking, excessive porosity and spatter.
2. Good clean weld metal free from slag, gas and other types of inclusion common with the cellulose type of electrode.

3. The property of adding alloys to the weld metal through the coating efficiently without excessive loss going through the arc.

During World War II it was this type of electrode which made it possible to effect large repairs in cast armor which otherwise would have resulted in costly scrap.

While the advantage of matching the chemical analysis of the base with the weld metal is apparent for uniformity of the finished product, other disadvantages such as the large number of different electrodes required to match the variations in casting make it impracticable. Since this method is based on performance rather than a chemical analysis makes it possible for the electrode manufacturer to design an electrode for the repair of several different casting analyses, the advantages of which to both consumer and producer are quite apparent.

CHAIRMAN RANDALL: It was mentioned that the so-called lime-type electrodes gave the greatest freedom from porosity. With the particular electrodes that were used for this investigation, was it found necessary to take any special precautions insofar as the moisture content of the electrodes was concerned? These particular types of coatings are somewhat hygroscopic and will pick up moisture readily. Many shops in the past have resorted to heating the electrodes prior to their use and I wondered whether any precautions were taken in connection with the electrodes in this test work.

R. D. THOMAS, JR.<sup>5</sup>: This question of moisture in a low-hydrogen, lime-type of electrode coating has been a problem facing the electrode manufacturers ever since they first came out. That dates back to 1934 when we started using this type of coating for austenitic stainless steels, later for stainless chrome irons, then for the low-chrome steels, during the war for the manganese-moly steels, and recently for the mild steels. So we have background on it, but we have not yet completely solved this moisture pick-up problem.

I would like to clear up several possible misunderstandings.

Yes, these electrodes are hygroscopic, but so also are any of the other electrodes that I have had occasion to test, the difference being that in the lime-type electrodes the introduction of moisture gives bad results from the standpoint of cracking, and sometimes porosity.

In the cellulosic-type electrodes which are never baked as dry in the first place, a change in the moisture content does not show much difference. To say that lime-type electrodes are more hygroscopic than the cellulosic type or the other types you have been used to is not necessarily true, but the absorption of small percentages of moisture by the lime type does change its properties.

A second point is the effect of that moisture pick-up on the different types of alloy steels that are deposited by the low-hydrogen type electrode. A small amount of moisture absorbed in an electrode which gives 70,000 psi ultimate strength, such as Mr. Wozny was using in most of his work, does not have nearly so severe an effect on either porosity or cracking as it would have on an electrode deposit which has 100,000 psi tensile strength. Wartime electrodes for cast armor had 110,000 to 120,000 psi tensile strength, whereas these new electrodes for the Grade B steel are for 70,000 psi tensile strength, and therefore the moisture pick-up is not so serious.

We electrode manufacturers, have made some progress which will help overcome some of the difficulties in moisture in electrode coatings. In the first place, we are doing a better job of packaging electrodes to prevent it. Our present packages are designed to prevent adsorption of excessive moisture in normal storage conditions over a period of 6 to 12 mon. However, it does not solve a more serious problem, which is the routine of storekeepers in various plants using the electrodes. In many cases, the storekeepers will take these electrodes out of the package, put them in bins and expect them to be useable after several months of exposure to humid atmospheres. In that case the moisture pick-up, unless the room is controlled to low humidity, is going to be serious and therefore all manufacturers will strongly recommend that in the low-hydrogen type of coating, you keep the electrodes in their original containers. However, sometimes you have to break that container and take 5 or 10 lb out of a 50-lb container, and the rewinding is sometimes more than the storekeeper wants to bother with. He is apt to leave the box open, and then the remainder of the electrodes will absorb moisture.

Our general recommendation is that in open cases, the storekeeper be provided with drying facilities so that he can pre-dry the electrodes before dispensing them especially if they have stood in open containers for any length of time. A 300-degree temperature for a period of 1 or 2 hr is usually adequate.

In regards to getting the electrodes too dry, that is true only under extreme conditions. If these electrodes have absorbed a great amount of moisture, their ceramic qualities have been changed. On re-drying, you get different welding characteristics, because the electrode coating does not have the same crucible action which the original electrodes had. It is not because the coating is too dry, but because of insufficient bonding of the ceramic that you occasionally discover your porosity after re-drying. In other words, it is not the absence of moisture that gives the porosity, but it is the absence of the shielding effect of the coating during the burning of the electrode.

CHAIRMAN RANDALL: We ought to bring out the question of moisture content because when people who are not familiar with this type of electrode start to use it for the first time, they will find they cannot handle it in exactly the same manner as they handled conventional electrodes and they will probably conclude it is not satisfactory.

F. P. HUBER<sup>5</sup>: Mr. Wozny stated that tempering did not materially reduce the surface hardness of the weld. If, following normal procedures, we welded our castings after the heat treatment and then just gave the casting a tempering treatment, we would have hard spots on the surface of the weld. From a machining point of view, would you recommend normalizing after the welding?

MR. WOZNY: I mentioned the experiment we made at the laboratory where a hardness differential of 84 Brinell points had no effect on machining. Of course, our machining was not of the high speed type that it would be in commercial machine shops. If there was no trouble in machining, I would not worry

<sup>4</sup> Harnischfeger Corp., Milwaukee.

<sup>5</sup> Berkley Machine & Foundry Works, Norfolk, Va.



too much about it, unless the use of the casting would be influenced somewhat by the hardness differential.

CHAIRMAN RANDALL: It has not been the general experience that tempering fails to reduce the hardness of welds and I think that in this particular case, that condition is associated with the particular materials being used and the choice of tempering times involved. You have a chrome alloy steel as the weld metal. I think your heat treatment consisted of one hour at temperature, which is probably an insufficient time to soften that material. It is not that the material will not soften on tempering, because the 5%-Cr alloy steels are regularly softened in weld metals on tempering by treatments below the critical. True, they do not go down to the minimum softness that you get on a full anneal, but it is perfectly possible to soften materials of that type by proper selection of tempering temperatures and times.

It has not been the general experience that tempering will fail to reduce the hardness of weld metals any more than of base metals, provided the tempering time is long enough, and the temperature is sufficiently high.

On a 2-in. thick casting, you would not normally consider that one hour was an adequate tempering time, would you?

Mr. WOZNY: That is debatable, but even in the 1/2-in. section there was no appreciable drop in hardness of the weld after

tempering. You will recall from Table 3 that in the 1/2-in. plate there was only a slight drop in the hardness of the weld after tempering at temperatures as high as 1350 F. Tempering at 1350 F for 1 hr is certainly more effective in reducing hardness than several hours at 1050 or 1150 F.

CHAIRMAN RANDALL: I thought that your specimens showing no reduction of hardness were held 1 hr and air cooled.

Mr. WOZNY: The tempering time was 1 hr.

CHAIRMAN RANDALL: In any case, I think you will recall that your experience was that the so-called manganese-molybdenum electrode did show a reduction in hardness on tempering.

C. W. BRIGGS<sup>6</sup>: On item No. 6 of the Conclusions, it was stated that plain carbon steel can be safely welded without any preheating up to the maximum section used in this study. Does that apply to 0.50, 0.60 or 0.70%-carbon steels?

Mr. WOZNY: No, the carbon should have been limited to the range of the steel investigated.

Mr. BRIGGS: This point is brought up since in the first part of the paper, you did not specify what the maximum carbon content of any of the steels you studied; you only gave an average which was 0.28 per cent carbon.

Mr. WOZNY: That should have been elaborated.

<sup>6</sup> Steel Founders' Society, Cleveland.

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# A SUGGESTED METHOD FOR THE DETERMINATION OF COKE REACTIVITY TO CARBON DIOXIDE AT COMBUSTION TEMPERATURES

By

H. Edward Flanders\*

## ABSTRACT

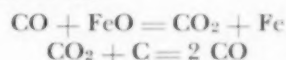
The generally poor operation of cupolas in recent years has been attributed to the improper preparation of cokes for foundry use. In an effort to determine the fundamental differences between good and bad cokes a method of determining the rate of reaction of coke to carbon dioxide at the temperatures existing in cupola combustion is proposed. It is based on the analysis of combustion gases as a column of coke burns as nearly adiabatically as possible. From the relation of composition to the height of the coke column the rate of the reaction can be approximately calculated and the coke reactivity to carbon dioxide estimated as a function of temperature.

## General Introduction

USE OF COKE AS FUEL in a blast furnace and in a cupola has arisen from the peculiar advantages of coke over the other fuels in these processes. The low density, irregular shape, and structural strength are conceded to be of benefit in the operation of these two units because of the contribution to stack porosity and the increased volume of wind which can be blown. Cavity is created as the coke burns, and this assists the decent of the change in the stack. Gaseous fuels would not give this advantage, and coal, by reason of its greater density, would only partially satisfy this condition. There would also be the waste of volatiles.

It is generally recognized that cokes from different coals, or from different coking treatments, have unequal properties when used in blast furnaces or cupolas. Some of these differences are made the subject of analysis in the determination of certain physical properties such as shatter, abrasion, density, etc. Other properties including structure, volatile matter, ash content, etc., are also determined and reported. One of the hidden properties, which is more or less ignored, is the rate of chemical reaction with oxygen, steam, or carbon dioxide. There is a feeling among many that the rates of reaction with these gases become the same for all cokes at the operating combustion temperatures, and that their study is of little importance. This may be more true in the operation of a blast furnace than in a cupola since in the former,

the reaction goes completely to carbon monoxide just above the combustion zone. However, it is of distinct advantage to have a blast furnace coke which is highly reactive to carbon dioxide in the reduction zone so that, in effect, direct reduction by carbon is promoted by the more complete removal of carbon dioxide. The pertinent reactions are:



The addition of these two reactions gives:



which is the reaction resulting from the removal of carbon dioxide by the coke and represents the basic equilibrium in the presence of coke.

In a cupola, the coke is not burdened with the function of ore reduction in addition to the requirements for heat. The elaborate processes of preheating the air necessary to attain sufficient temperature in a blast furnace are not usually a part of cupola air treatment simply due to the fact that the carbon dioxide is not reduced completely to carbon monoxide, as in the case of the blast furnace, and the corresponding temperature of the combustion gases are not lowered to such an extent.

## Differences in Cokes

In recent years, the rush for coke has markedly reduced the manufacture of hard burned coke for foundry use, and cupola operation has become more difficult. It may be that this is due simply to the lower temperature resulting from the more complete reduction of carbon dioxide to carbon monoxide when the more highly reactive coke of present manufacture is used. This handicap can be overcome by preheating and with this aid much less worry need be paid to coke quality. But such a course gives no knowledge about the basic reason for the differences in cokes as prepared currently in comparison with cokes prepared formerly and which are reputed to have been much better. It may well be that the coke reactivity, particularly to carbon dioxide, is of paramount importance, and that in tests for this property, cokes will be found to differ greatly.

\* Department of Metallurgical Engineering, University of Utah, Salt Lake City, Utah.

Methods of determining the reactivity of coke to carbon dioxide at moderate temperatures have been developed. Probably the best method is that of Jones, King and Sinnatt,<sup>1</sup> by which the reactivity at 1740 F (950 C) is studied. No literature has been found which gives a method for such a determination at combustion temperatures and these notes are written to suggest such a method.

### Suggested Method

The reaction between carbon and carbon dioxide is a typical mono-molecular reaction in which the concentration of only one reactant is changing. The velocity of such a reaction in terms of the rate of change of concentration with time is expressed by the standard formula

$$\frac{dC}{dt} = kC \quad (\text{Eq. 1})$$

where  $C$  is the concentration of carbon dioxide at any time  $t$  and  $k$  is a constant known as the specific reaction rate constant. There are three unknowns and the equation is solved when any two are determined. In the study of the above reaction, the rate  $dC/dt$  can be found by taking the slope of the line which gives the concentration of  $\text{CO}_2$  as a function of time. To do this it is necessary to analyze the process of combustion in a suitable manner so that the value of  $k$  can be determined.

To simplify the treatment let the following experimental conditions and assumptions be suggested.

A column of carefully sized coke, of such height that equilibrium will become fixed after combustion has proceeded for a time, is contained in a uniform, insulated cylinder of moderate dimensions. The refractory chosen is one capable of withstanding high temperatures with only a minimum of fusion so that the process can be considered as nearly adiabatic as possible. Along the side, parallel with the axis of the column, is a series of plugged holes, through which a prod may be pushed to measure the height of the coke as the coke burns. Gas analyses corresponding with the height of the coke column can be made by suitable continuous analyzers and by intermittent checks with Orsat or Burrell equipment. The air is carefully measured and blown at a constant rate. With the data thus obtained, the composition of the gas as a function of the height of the coke above some arbitrary reference, such as the tuyeres, can be easily shown.

### Assumptions Summarized

Several assumptions may be summarized:

1. Since the process is adiabatic, or nearly so, the temperature corresponding to any particular gas analysis can be easily calculated. It is assumed, in this procedure, that the composition of the gas coming off the coke is the same as would have been found had analyses been taken at that position when the coke column was much higher.

2. The free space between the pieces of coke through which the gas moves is essentially constant during combustion, or, for any particular size of coke, varies in a uniform manner, and can be taken as that

given by a bulk density determination. In the examples given below this free space is taken as 30 per cent.

3. Carbon enters combustion at the combustion temperature as does the ash since these are preheated during their descent in the column. The temperature of the coke, gas, and ash will be assumed to be identical at any position, although there is probably some lag in the attainment of temperature uniformity.

4. The air enters combustion cold, 75 F, and the carbon and ash enter the zone under consideration at about 3000 F (1650 C), and heat is used in further raising their temperature. The equations giving the heat contents of these substances are based on these references.

The quantities used in making the calculations have the following significance:

$a$  = cross sectional area of the coke column in in.<sup>2</sup>

$b$  = volume of air blown in cubic feet per minute

$f$  = free space (= 0.30 in the example)

$\Delta H_b$  = heat content of reactants and products due to heating

$\Delta H_c$  = heat of combustion

$T_r$  = temperature °R at reference = 492°R (0°C) (32°F)

$T$  = temperature °R

$N_v$  = fraction of carbon burned to  $\text{CO}_2$

$h_1$  = height above tuyeres in inches

$h$  = height above 21 per cent  $\text{CO}_2$  level

$x$  = per cent ash in coke

$y$  = per cent carbon in coke

$v_r$  = velocity of gases through coke at standard temperature in inches per second.

$v$  = velocity of gases through coke in inches per second

$t_1$  = time per inch in seconds at any position

$t$  = time between 21 per cent  $\text{CO}_2$  level and any particular position in seconds.

The velocity of the gases through the coke column may be calculated directly.

$$\frac{b \text{ ft}^3/\text{min} \times 1728 \text{ in.}^3/\text{ft}}{a \text{ in.}^2 \times f \times 60 \text{ sec/min}} = \frac{28.8 b}{af} = \text{in./sec}$$

As the combustion gases rise in the cupola there is a contraction due to the loss of temperature which is partially compensated by the reaction of carbon dioxide with carbon. The effect of temperature is given

directly by the ratio  $\frac{T}{T_r}$  and the volume change due

to reaction is given by the expression  $1.21 - 0.21 N_v$ . Under the conditions of perfect combustion to  $\text{CO}_2$  the volume of combustion gas is equal to that of the air blown when measured under similar conditions. The expression for velocity of the gas through the coke is then

$$v = \frac{28.8b}{af} \times \frac{T}{T_r} \times (1.21 - 0.21 N_v) \text{ in./sec.}$$

or

$$v = \frac{0.0585b}{af} \times T \times (1.21 - 0.21 N_v) \text{ in./sec.}$$

In two of the examples given later, the air blown was 750 cfm in a No. 1 Whiting cupola lined to 22-in.



Substituting in the above equation gives:

$$v = 0.385 T (1.21 - 0.21 N_v) \quad (\text{Eq. 2})$$

In a third test the air blown was 400 cfm, and the corresponding expression is

$$v = 0.205 T (1.21 - 0.21 N_v) \quad (\text{Eq. 3})$$

As mentioned above, the calculation of the temperature  $T$  can be accomplished by the usual procedure, since this temperature is fixed by the composition of the gaseous products of combustion. If carbon enters combustion at a temperature above 3000 F, its contribution to the heat of reaction is exactly compensated by the heat used in its preheating above this temperature. The air enters into combustion at 535° R (75 F or 25 C) and the ash has the same temperature as does

the carbon. Above 3000 F the ash is probably fused, and has a specific heat of 0.30 Btus per pound as given by Allison Butts<sup>2</sup> for acid slags. These conditions are included in the following equations which are based on dry air.

$$\Delta H_c = 52200 + 124200 N_v = \text{heat of combustion of carbon at } 535^\circ \text{R (75 F) or (25 C) in Btus.}$$

$$\Delta H_b = -1868 + 2.997T + 0.0004974T^2 = \text{heat content of carbon at } T^\circ \text{R in Btus.}$$

Summing

$$\Delta H_c = -50330 + 124200 N_v + 2.997T + 0.0004974T^2 = \text{heat of combustion at } T^\circ \text{R in Btus.}$$

The heat contents of the combustion gases in Btus may be calculated as follows from data given in another paper.<sup>3</sup>

$$\Delta H_b \text{ for } N_2 = (1.88 + 1.88 N_v) (-4050 + 7.094T + 0.0001976T^2)$$

$$\Delta H_b \text{ for } CO_2 = N_v (-6530 + 11.142T + 0.0004434T^2)$$

$$\Delta H_b \text{ for } CO = (1 - N_v) (-4135 + 7.210T + 0.0001907T^2)$$

$$\Delta H_b \text{ for } C \text{ above } 3460^\circ \text{R (1650 C)} = -16330 + 2.997T + 0.0004974T^2$$

$$\Delta H_b \text{ for ash above } 3460^\circ \text{R (1650 C)}$$

$$= \frac{x}{y} (3.6T - 12460)$$

Summing the above equations for the combustion products there results:

$$\Delta H_b \text{ gases and coke} = -28080 - 10010 N_v + (23.54 + 17.27 N_v) T + (0.001065 + 0.000623 N_v) T^2$$

$$+ \frac{x}{y} (3.6T - 12460)$$

At the theoretical combustion temperature this is equal to

$$-\Delta H_c = 50330 + 124200 N_v + 2.997T + 0.0004974T^2$$

and on combination there results

$$-78410 - 134210 N_v + (20.550 + 17.27 N_v) T$$

$$+ (0.000561 + 0.000623 N_v) T^2 + \frac{x}{y} (3.6T - 12460)$$

$$= 0$$

Assuming that the coke contains 8 per cent ash and 90 per cent carbon the following data were calculated and are plotted as a part of Fig. 1 from which the temperature for any  $N_v$  can be taken.

$N_v$	$T^\circ \text{R}$	$T^\circ \text{F}$	$T^\circ \text{C}$
0.00	3485	3025	1665
0.20	3935	3475	1915
0.40	4265	3805	2100
0.60	4515	4055	2235
0.80	4680	4220	2330
1.00	4840	4380	2420

The temperatures check closely with those reported previously.<sup>3</sup>

The relationship between the  $CO_2$  content of the combustion gases and  $N_v$  is also shown in Fig. 1.

By plotting the carbon dioxide concentration of the combustion gases as a function of the height above the tuyeres  $h_1$  the data may be extrapolated, either mathematically or graphically, to give the hypothetical position at which 21 per cent  $CO_2$  existed. It is this

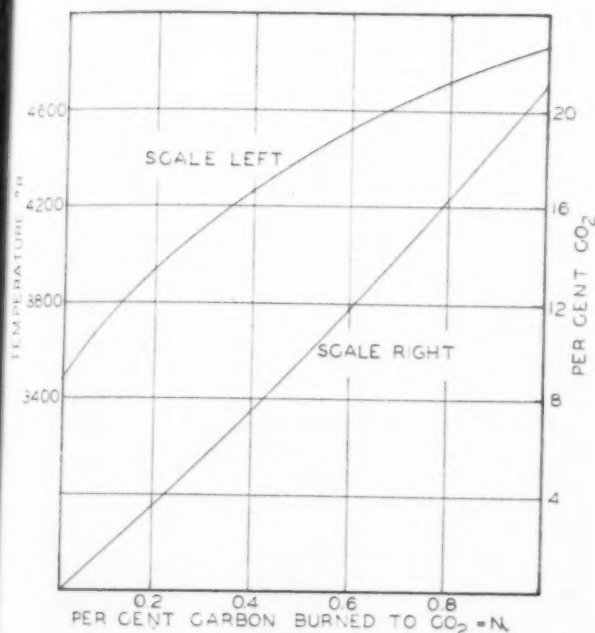


Fig. 1—Relation between per cent  $CO_2$  in combustion gases, fraction carbon burned to  $CO_2$ , and theoretical combustion temperatures.

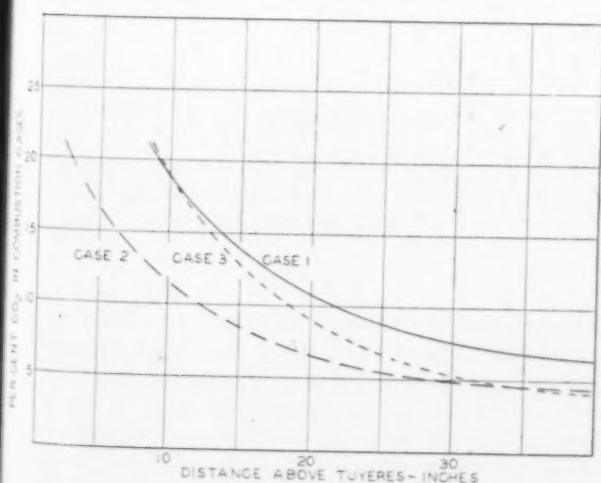


Fig. 2—Original data expressed as per cent  $CO_2$  in combustion gases at different levels of coke above the tuyeres.

new reference from which the height  $h$  is determined and the composition is easily plotted with reference to this level. The choice of 21 per cent  $\text{CO}_2$  as an initial concentration from which to work is in conformity with the ideal setting of the problem.

Since the composition of the combustion gases is assumed to fix the existing temperature, this may be found by reference to Fig. 1. By substituting these values for  $N_v$  and  $T$  in Eq. 2, the velocities of the gases can be readily calculated and plotted as a function of the height  $h$ . The reciprocal of the velocity is the time per inch  $t_i$  and this can be also plotted as a function of  $h$  and the algebraic equation found. When  $t_i$  is multiplied by  $h$  the time is evaluated

$$ht_i = t$$

so that the concentration can be now plotted as a function of time. This is the necessary condition for the solution of Eq. 1. But the corresponding values of  $C$  and  $t$  for this curve have different temperatures so that by evaluating  $k$  by taking slopes of the lines at different concentrations of carbon dioxide, also at different but determinable temperatures, the constant  $k$  can be plotted as a function of temperature.

As an example of this method, some data reported earlier<sup>3</sup> have been used although the tests were not made with more than a qualitative objective completely foreign to the present study. A more accurate experimental setting should be much more satisfactory. The examples are used primarily as an illustration of the method of treating the problem.

### Three Preliminary Tests

Data were selected from the lines representing the experimental results of three coke burning tests in which the  $\text{CO}_2$  content of the combustion gases were plotted as a function of the height of the coke above the tuyeres in inches. They are replotted in Fig. 2 and are also given in Table 1. The extrapolation of the curves, derived by the method of averages, to the 21 per cent  $\text{CO}_2$  level is shown in Fig. 2 and given in Table 1. Curves redrawn with reference to the 21 per cent  $\text{CO}_2$  level are shown in Fig. 3. The equations for these curves are the same as for the original curves except for the different origins. The gas velocities calculated from the original data are included in Table 1. The functional relationship between time per inch  $t_i$  and distance above the 21 per cent  $\text{CO}_2$  level  $h$  (Fig. 4) and the total time  $t$  are evaluated for each of the three cases. All the equations are included in Table 2. Figure 5 gives the data for the concentration of carbon dioxide as a function of time, and the derived slopes,  $dC/dt$  from which values of  $k$  and  $t$  are calculated, are summarized in Table 3. The values for  $k$ , the reaction velocity constant, are plotted as a function of temperature in Fig. 6. The curves suggest several important items:

1. The effect of temperature on  $k$ .
2. The effect of gas velocity on  $k$ .
3. The effect of coke size on  $k$ .

All these effects are in the expected direction. More complete and accurate data might define each of these relationships quantitatively. Perusal of the curves

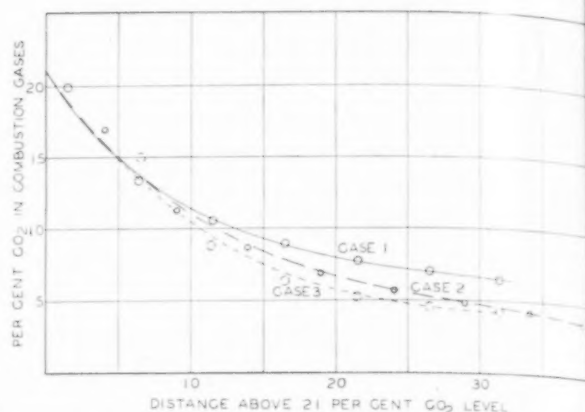


Fig. 3—Original data expressed as per cent  $\text{CO}_2$  in combustion gases at different levels of coke above the hypothetical 21 per cent  $\text{CO}_2$  level.

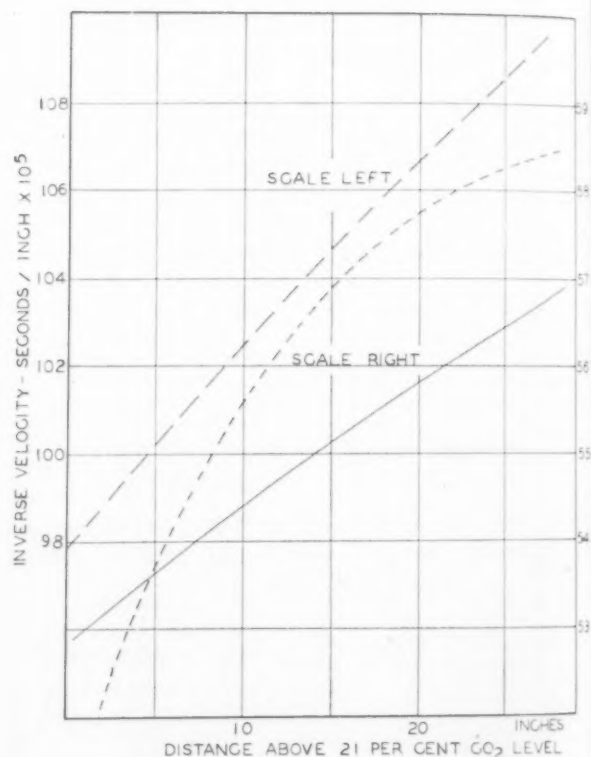


Fig. 4—Inverse velocity of combustion gases at different distances of top coke surface above the hypothetical 21 per cent  $\text{CO}_2$  level.

suggests that the effect of velocity on turbulence and on reaction rate may be quite similar to the effect of velocity on the rate of heat extraction from heated coils or straight pipes placed transverse to the direction of the flow; or to the effect of velocity on the reaction rate in towers packed with spherical balls. In both cases the rate of heat transfer is proportional to some fractional power of the velocity. Inconsistencies in the data are apparent but these may disappear when proper conditions are provided.

## Conclusion

In conclusion, it may be said that a method of determining the rate of reaction of coke with carbon dioxide under combustion conditions has been outlined. While the test may give values which are but approximately true, it is expected that the errors involved in the assumptions will be common under

standardized conditions of test, and the results will be of the proper relative order. The assumption that the properties of coke are constant under combustion conditions seems rather insecure. On the other hand, it is expected that cokes do differ in this respect and that this difference will give an excellent measure of the ability of the coke to give a high temperature at a suitably positioned melting zone in a cupola.

TABLE 1—SUMMARY OF SELECTED DATA FROM COKE BURNING TESTS WITH CALCULATED GAS VELOCITIES

Case No.	Distance Above		Carbon Dioxide		Temperature			Gas Velocity In./Sec.	Inverse Velocity Sec./In.
	Tuyeres	21% CO <sub>2</sub> Level	%	N <sub>2</sub>	°R	°F	°C		
1 Coke 2 in.-2½ in. Wind 750 cfm	0								
	5								
	10	1.5	20	0.92	4785	4325	2385	1879	0.000533
	15	6.5	13.3	0.51	4385	3925	2180	1862	0.000537
	20	11.5	10.7	0.38	4200	3740	2080	1828	0.000547
	25	16.5	9.1	0.31	4095	3635	2020	1805	0.000554
	30	21.5	7.9	0.26	4005	3545	1955	1790	0.000562
	35	26.5	7.1	0.23	3950	3490	1920	1767	0.000566
	40	31.5	6.5	0.21	3915	3455	1901	1758	0.000569
2 Coke 2 in.-2½ in. Wind 400 cfm	0								
	5	4	17.0	0.72	4630	4170	2300	1008	0.000992
	10	9	11.3	0.39	4230	3770	2080	976	0.001025
	15	14	8.8	0.29	4065	3605	1985	960	0.001042
	20	19	7.0	0.22	3940	3480	1815	942	0.001062
	25	24	5.9	0.18	3860	3400	1870	928	0.001079
	30	29	4.9	0.15	3800	3340	1840	920	0.001087
	35	34	4.3	0.13	3760	3300	1815	912	0.001096
	40	39	3.8	0.11	3720	3260	1795	907	0.001103
3 Coke 1½ in.-2 in. Wind 750 cfm	0								
	5								
	10								
	15	6.5	15.0	0.47	4335	3875	2135	1834	0.000544
	20	11.5	8.8	0.30	4070	3610	1990	1796	0.000557
	25	16.5	6.5	0.20	3895	3435	1890	1751	0.000571
	30	21.5	5.3	0.16	3820	3360	1850	1725	0.000580
	35	26.5	4.6	0.14	3780	3320	1830	1718	0.000582
	40	31.5	4.3	0.13	3760	3300	1815	1712	0.000584

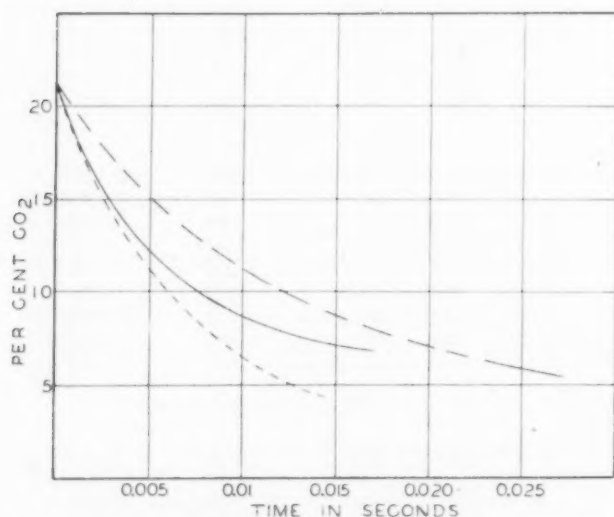
TABLE 2—CALCULATED EQUATIONS FOR EXPERIMENTAL CURVES, DERIVED CURVES REFERRED TO 21% CO<sub>2</sub> LEVEL, INVERSE TIME, TIME, AND CORRESPONDING DATA

Case No.	Equation	Height Above 21% CO <sub>2</sub>	% CO <sub>2</sub>	Total Time Above 21% CO <sub>2</sub> in sec.
1 Coke 2 in.-2½ in. Wind 750 cfm	$C = 34.64 - 2.17 h_1 + 0.0572 h_1^2 - 0.000525 h_1^3$	5	14.58	0.00268
	$C = 21.00 - 1.311 h + 0.0438 h^2 - 0.000525 h^3$	10	11.75	0.00544
	$t_1 = 0.000528 + 0.00000173 h - 0.00000001 h^2$	15	9.40	0.00828
	$t = 0.000528 h + 0.00000173 h^2 - 0.00000001 h^3$	20	8.00	0.01117
		25	7.40	0.01412
		30	6.92	0.01732
2 Coke 2 in.-2½ in. Wind 400 cfm	$C = 22.60 - 1.45 h_1 + 0.0415 h_1^2 - 0.000433 h_1^3$	5	15.31	0.00501
	$C = 21.19 - 1.364 h + 0.0402 h^2 - 0.000433 h^3$	10	11.04	0.01024
	$t_1 = 0.000978 + 0.00000481 h - 0.00000002 h^2$	15	8.31	0.01570
	$t = 0.000978 h + 0.00000481 h^2 - 0.00000002 h^3$	20	6.53	0.02132
		25	5.47	0.02715
		30	5.10	0.03313
3 Coke 1½ in.-2 in. Wind 750 cfm	$C = 19.24 - 1.169 h_1 + 0.02275 h_1^2$	5	15.34	0.00269
	$C = 20.95 - 1.237 h + 0.02275 h^2$	10	10.85	0.00556
	$t_1 = 0.000515 + 0.00000497 h - 0.00000009 h^2$	15	7.52	0.00853
	$t = 0.000515 h + 0.00000497 h^2 - 0.00000009 h^3$	20	5.31	0.01157
		25	4.27	0.01458
		30		



TABLE 3—TABULATION OF SLOPES (FROM FIG. 5) AND OTHER DATA

Case No.	Time in Sec.	C = % CO <sub>2</sub>	dC/dt	k	N <sub>2</sub>	Temperature			Velocity of Gas In. per Sec.
						°R	°F	°C	
1	0.0025	15.2	1505	99	0.75	4670	4210	2320	1890
	0.0050	12.1	944	79	0.61	4540	4080	2250	1890
	0.0075	10.1	635	63	0.52	4430	3970	2195	1876
	0.0100	8.75	437	50	0.46	4350	3890	2145	1863
2	0.0050	15.2	968	64	0.75	4670	4210	2320	1008
	0.0100	11.25	603	54	0.58	4500	4040	2225	998
	0.0150	8.75	392	50	0.46	4300	3840	2115	991
	0.0200	7.15	283	40	0.38	4230	3770	2075	976
3	0.0025	15.25	1760	115	0.75	4680	4220	2325	1895
	0.0050	11.4	1300	114	0.58	4500	4040	2225	1874
	0.0075	8.6	940	109	0.45	4340	3880	2140	1863
	0.0100	6.7	660	99	0.36	4210	3750	2065	1837

Fig. 5—Relationship between per cent CO<sub>2</sub> in combustion gases and time of travel above hypothetical 21 per cent CO<sub>2</sub> level.

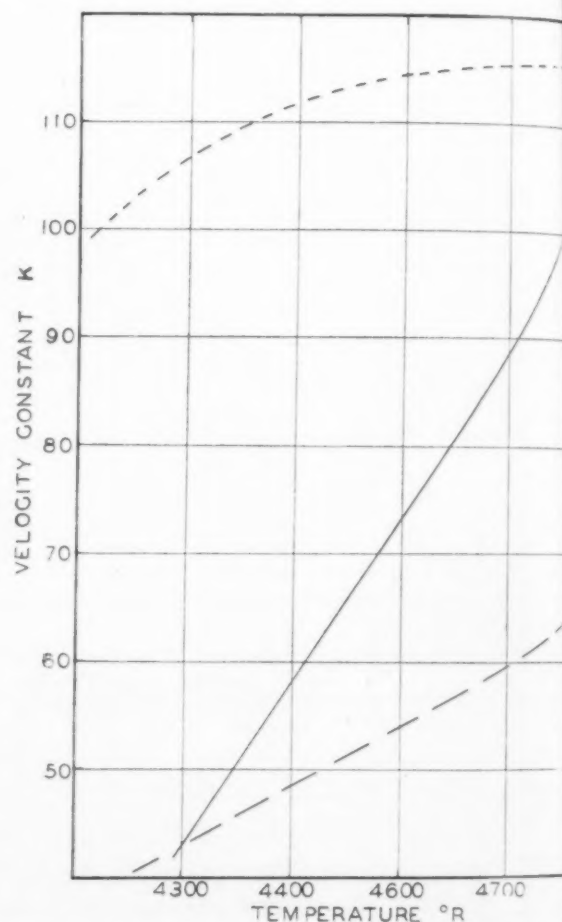
The effect of the several conditions of combustion such as moisture content in the air, oxygen enrichment, coke size, or wind velocity should be readily studied and evaluated by this method. While the above calculations are for dry air similar calculations can be made for moist air by incorporation of the principles given in this paper.

#### Acknowledgment

The author wishes to thank the University of Utah for its support and cooperation during the writing of this paper as a part of its research program. Thanks are also due to John R. Lewis, Head of the Department of Metallurgy for his assistance in this work.

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Fig. 6—Relationship between reaction velocity constant "k" for the reaction  $C \text{ (Coke)} + CO_2 = 2CO$  and the temperature.

#### DISCUSSION

Chairman: V. A. CROSBY, Climax Molybdenum Co., Detroit, Mich.  
Co-Chairman: D. E. KRAUSE, Gray Iron Research Institute, Columbus, Ohio

J. S. VANICK:<sup>1</sup> I find that in Mr. Flander's paper it is little difficult to square the velocity relationship in Table 3 with some of the things that happen in practical operation. For instance, we all know that in Table 1, as the results indicate

<sup>1</sup> International Nickel Co., Inc., New York

...the blast will lower the bed. We also know, and the  
...seem to indicate, that cutting down the coke size will  
...down the bed and Table 1 seems to back that up. Mr.  
...figures indicate that the velocity in the 750-ft blast and  
...1/2-in. coke (which has about a 10-cu in. volume) is the  
...is the velocity for the same 750 ft of air, in a smaller  
...that has about 5 cu in. of volume). It seems to me that  
...explanation is necessary. Either the smaller-sized coke  
...building up a pressure (no pressure values are expressed  
...which in turn might mean a greater quantity of air is  
...ing on that coke surface, or to get the same volume of air  
...ugh the small coke the velocity should have increased at  
...air-coke interface.

1890  
1890  
1876  
1863  
1068  
998  
991  
976  
1895  
1874  
1863  
1837

FLANDERS: I think the best way to answer the question  
...follows: If one takes a group of spheres 12 in. in diameter  
...another group of spheres 1 in. in diameter, with sufficient  
...of spheres to eliminate the effect of the container, it  
...be found that the unoccupied space (pore space) is the  
...in each instance. It is on this basis that the pore space  
...assumed constant at an estimated 30 per cent as a basis  
...calculation. The minimum pore space for packed spheres  
...per cent.

...the tests used as examples, the coke was very carefully  
...in all instances and unless the shape factor is different in  
...several cases, the pore space is the same. The pressure  
...ary to push air through the small coke is greater than  
...the larger coke but this is not due to a change in the  
...space.

...particular attempt was made to calculate the space factor  
...1 1/2-in. or a 2 1/2-in. cube as is suggested, but the simple  
...ate of 30 per cent was assumed.

...VANICK: Theoretically that answers the question, but  
...we try to get a higher bed we are very likely to increase  
...coke size in order to get the higher velocity blast.

...FLANDERS: Thank you for reminding me of this. It can  
...en on the curves that the rate of reaction of CO<sub>2</sub> with  
...was decidedly higher with the smaller coke. With the  
...er coke the bed would rise due to the reduced rate of  
...tion of CO<sub>2</sub> with coke higher in the bed. Thus the level  
...the hypothetical 21 per cent CO<sub>2</sub> content gas is raised as is  
...the bed height with larger coke.

...then these curves were plotted, comparisons with similar  
...ical conditions in which heat is extracted from closely  
...ed balls, or from staggered pipes by passing gas through  
...were attempted. In these cases the amount of heat ex-  
...ed (or, in the inverse case, put in) is an exponential power  
...than unity) of the gas velocity. Size is of definite im-  
...ance. While I was not satisfied with the relationships which  
...found as applied to the combustion of coke, there is  
...ely an effect of size, and this is shown in the paper.

...is the author's opinion that there are two unfavorable  
...tions under which to melt iron in a cupola. These are  
...a high CO<sub>2</sub> content gas which hinders superheating and  
...a low CO<sub>2</sub> content gas where the temperature may be too  
...These conditions are aggravated by small coke particu-  
...when lean coke ratios are used and the bed gets very low.  
...is based on the simple facts of higher reaction velocity  
...that the amounts of coke and air entering the bed must  
...ce with the amounts leaving. Per unit volume of air,  
...e coke leaves when carbon is in the form of CO.

...H. SPENCER: I wish to express appreciation to Dr.  
...nders for this paper on a subject which has usually in the  
...been considered rather thankless work. Probably there will  
...more work from time to time on the basic or fundamental  
...ies underlying cupola operation.

...then we say that a coke will do so and so, we mean that  
...coke will do that. There are quite a variety of cokes, not  
...in physical size but also in structure. There are certain  
...of which a large lump of coke will react faster than a  
...er lump of another type of coke. That needs to be kept  
...ind in all of these discussions.

...another point that struck me in passing is that the author  
...his presentation adhered entirely to the concentration of  
...gas as determining the reaction rate. The speed of reaction  
...pendent upon many other factors besides the concentration  
...the CO<sub>2</sub>. It is dependent upon the reactivity of the carbon  
...xygen and CO<sub>2</sub>. It is also dependent upon the pressures

and temperatures existing. It is dependent upon the tempera-  
...ture under which the reaction is taking place probably as  
...vitaly as it is upon the concentration. The speed of reaction  
...is usually directly variable with the square of the absolute  
...temperature, if I remember correctly. The temperatures inside  
...a cupola are variable from inch to inch. There is no level  
...throughout the cupola where you can say the concentration  
...of CO<sub>2</sub> is so much in the cupola. That is a point to remember  
...in making any calculations and trying to apply them to actual  
...practice.

These are not intended as criticisms but as suggestions for  
...somebody to work out these other variables and add them to  
...this information. I would like to suggest that this test can not  
...be applied directly as the one and only answer to your coke  
...problems or as a test for coke that you can make in a few  
...minutes.

DR. FLANDERS: I thank Mr. Spencer for his remarks. What  
...the test is expected to do is included in some of the items  
...mentioned. By properly choosing the conditions it should give  
...approximately the effects of coke size, wind velocity, and water  
...vapor on the rate of combustion of coke to CO<sub>2</sub>, and on the  
...rate of reaction of CO<sub>2</sub> with coke. It should be possible also to  
...study the effect of oxygen enrichment. However, the primary  
...purpose of this test is to differentiate cokes under a series of  
...arbitrarily chosen standard conditions which present as nearly  
...as possible the conditions prevalent during combustion in a  
...cupola.

I have been told by several men who work with coke that  
...the reactivities of all cokes become the same at combustion  
...temperatures. This, very likely, is not the case. However, it  
...can be made the subject of direct experimentation through the  
...use of the method outlined.

H. A. ROTH: A minor point in the interesting relations  
...that are being discussed is that the 30 per cent voids assumed  
...in the coke differs from the determinations reported by B. P.  
...Mulcahy in his paper "Foundry Coke: A Critical Study" in  
...TRANSACTIONS, A.F.A., vol. 52, pp. 819-845 (1944).

On p. 832 thereof, Table 4 gives the voids in nine sizes of  
...coke, based on the weight per cubic foot (in a box holding 8  
...cu ft) compared to weight per cubic foot of the "solid coke".  
...A No. 1 Foundry Coke of 5.1-in. average size had 57 per cent  
...voids; a Blast Furnace coke of 2.22-in. average size had 50.4  
...per cent voids, and a Pea Coke of 0.51-in. average size had  
...47.9 per cent voids.

A photograph, Fig. 5, in this paper shows a mixture of a  
...4-in. x 3-in. coke of 54.8 per cent voids which was "saturated"  
...in the box with Nut Coke of 1.2-in. average size and 48.9 per  
...cent voids. The large lumps appear to be in contact. The  
...voids were found to be 48.3 per cent, or very close to the 48.9  
...per cent of the smaller coke. With closer packing and a  
...slight difference in sizes it might be possible to have the void  
...space closer to 48 per cent of 55 per cent, or near the 30 per  
...cent used in the calculations.

DR. FLANDERS: The determination of the space factor in  
...coke can be done experimentally and the determined value be  
...put into the formulae for calculation of the gas velocity through  
...the coke. While this value could have been determined, my  
...guess made for the purpose of illustration may be only approxi-  
...mately correct. I do not remember B. P. Mulcahy's work and  
...whether he was talking about 50 per cent voids in the coke  
...heap or in the coke piece, which may be about 50 per cent  
...depending on methods of preparation.

There is a theory of combustion given by Haslam and Rus-  
...sell in their book entitled *Fuels and Their Combustion*. They  
...discuss the formation of a complex C.O. which forms on the  
...surface of the coke and which releases the combustion gases.  
...In the application of the test method presented, there may be  
...a correction necessary to compensate for the reduced size of  
...the coke piece as it burns to leave a cavity somewhat larger  
...than estimated. This point will require experimentation. A  
...compensating factor will be the filling of the cavities by the  
...movement of the coke pieces.

W. R. JAESCHKE: I should like to ask if that theory of 30  
...per cent or the theory of free space is alike for all sizes of coke.  
...Then theoretically the small coke should have no more resis-  
...tance against the cupola blower with the same air input.

<sup>3</sup> Rundle Manufacturing Co., Camden, N.J.

<sup>4</sup> Whiting Corp., Harvey, Ill.

DR. FLANDERS: That is not quite right. The resistance to gas flow is due, in part, to the friction between the gas and the coke surface. The greater this surface the greater this friction. One might consider it quite similar to lubrication where the lubricant next to the metal does not move, thus requiring more movement in the body of the lubricant. When the channels are narrow the friction at the sides is effective throughout a greater proportion of the gas.

MR. JÄSCHKE: In actual practice we do get more resistance when we are using small coke. I wonder whether it is all due to friction or whether the load-bearing qualities of coke at those high temperatures is not a factor. Then, of course we do not have perfect spheres. We have large pieces and small pieces. There is no degree of uniformity. When large and small pieces are mixed, the percentage of free space is far below the theoretical 30 per cent.

Then also with the small coke the reactivity rate is higher and when coke is oxidized by  $\text{CO}_2$ , some of it is converted into  $\text{CO}$ . We then produce a larger volume of gases, do we not? Probably twice as much, and that creates resistance to the flow of the air too.

DR. FLANDERS: That is right.

MR. JÄSCHKE: Also I would like to know if the 21 per cent of  $\text{CO}_2$  was measured at the cupola wall.

DR. FLANDERS: That was taken as hypothetical. The actual concentration curve was extrapolated back to the hypothetical position of 21 per cent  $\text{CO}_2$  and that would be above the tuyeres. The original reference was the tuyere.

MR. JÄSCHKE: The was the outside of the column of coke?

DR. FLANDERS: No, it is an average for the bed.

MR. JÄSCHKE: Do you think that percentage of  $\text{CO}_2$  is uniform across the area?

DR. FLANDERS: I do not think it would be.

MR. JÄSCHKE: I think in actual practice or in tests, if you go back to some of Belden's work in the old inverted cone theory you would find that the inverted cone gets quite deep with the smaller size coke, that we do more melting on the outside of the cupola and confine it to a narrow band around the lining, and the center of the cupola is nothing but a gas producer. If measurements are taken toward the center of the cupola, I think you will find that you get a deeper cone with small coke, and this explains why in Case 3 that 21 per cent cupola is higher than it should be because by all theories of combustion that point should be lower than in Case 1. That was the smallest coke you used.

DR. FLANDERS: I have forgotten why that is. I would have to look it up.

MR. JÄSCHKE: Twenty-one per cent  $\text{CO}_2$  is about it.

DR. FLANDERS: In Table 1, the position of the 21 per cent  $\text{CO}_2$  level is indicated. For Case 1, it is  $8\frac{1}{2}$  in. above the tuyeres and in Case 3 it is also  $8\frac{1}{2}$  in. above the tuyeres. But it must be recognized that the tests studied were not made in accordance with the procedure recommended in the paper. Then too, the data were limited and extrapolation to 21 per cent  $\text{CO}_2$  may be somewhat in error. This hypothetical position of 21 per cent  $\text{CO}_2$  was considered a suitable reference on which to base the calculations of the velocity constant.

MR. JÄSCHKE: I am concerned more with conditions across the whole area of the cupola than I am with an average at a

certain level in the cupola. I do not think we know enough about what is going on in the center of the cupola.

DR. FLANDERS: Cupola gases go up the sides because that is the easiest path to follow. The relatively straight wall gives at least one surface that is not responsible for changing the direction of flow of the gas. There is loss of pressure at the change of direction of movement and, with relatively large coke pieces, most of the pressure difference existing in the presence of small and large pieces is probably due to this factor, rather than to friction between the coke and gases mentioned earlier in the discussion.

The tendency for the gases to move fastest at the cupola lining results in a greater consumption of coke in this area and the formation of a cone of coke at the center. This condition is found in the extreme in a blast furnace where very little combustion takes place at the center and a high  $\text{CO}$  concentration exists. But these conditions exist in a cupola or in a furnace somewhat in proportion to diameter. The accumulation of a cone of coke in the middle of a test furnace 12 in. in diameter would not be expected. It is on this basis that the average analysis of the gases is thought to be the result of the property of the coke. A test should then be able to give a measure of the reactivity of the coke to  $\text{CO}_2$ , and in addition give the effect of water vapor, oxygen enrichment, wind volume per unit area, and of coke size. From the nature of the test all these can be determined as a function of temperature.

DR. FLANDERS (*Author's Closure*): Much of the discussion of this paper tended to interpret the results of the test in terms of conditions existing in a cupola. Two facts must be made clear with respect to the paper.

1. A method of determining a property of coke at combustion temperatures has been outlined.

2. The experiments chosen were used as illustration of a method of handling the data and not as obtained by experimentation with the method.

By knowing the property of a coke, even though this property be determined only in relative values between different cokes, the operator can correlate the performance of a cupola in terms of the property of the coke. It is quite likely that cokes differ in their high temperature reaction characteristics and these should be differentiated by the use of standardized tests.

The discussion of the pore space between the coke pieces need not be extended. Any determined value for this pore space can readily be inserted in the formula and calculation performed. However, some statements may be made with respect to the pore space between metrical shapes.

The maximum pore space between cubes, in the absence of bridging, is 50 per cent. The minimum pore space is zero. In the absence of bridging, cubes should pack so that the pore space is between zero and 50 per cent. The amount of pore space between the coke pieces is dependent on the method of putting them into a container, the extent of vibrations, the shape of the pieces. The high pore space determined by B. P. Mulcahy ["Foundry Coke: A Critical Study", *Transactions, A.F.A.*, vol. 52, pp. 819-845, (1944)] does not seem to be in agreement with his Fig. 4 on p. 834, where the pore space seems to be definitely less than 50 per cent. However, the pore space is not of uniform size.

I wish to thank those who participated in the discussion for their contributions.



# APPLICATIONS OF CORRELATION IN THE MALLEABLE IRON FOUNDRY

By

Robert G. Seidel \*

## ABSTRACT

*The author presents examples and a discussion of correlation techniques and their possibilities as a useful tool applied to foundry problems.*

IN ALMOST ANY FIELD we can think of, it is often desirable and sometimes necessary to observe the relationship which occurs between two or more variables. Correlation and regression analysis have been used in science from the time scientists first became aware of these variables and tried to determine their relationship. The techniques of correlation have been developed through the years to the point that the accuracy of some predictions is almost unbelievable.

The specific fields in which correlation has been successfully applied are far too numerous to mention. However, a few of them are: biological studies where heredity, environment, etc., are related to human development; agriculture in that crop yields depend on rainfall, soil condition, locality, etc.; the field in which we are all actively interested, economics. Here our economic condition depends on the supply of and the demand for goods and the factors that determine the supply and demand.

In the present day foundry, we are continually confronted with problems too complex for solution by observation or rule of thumb methods. A few problems of this nature might be: How should the chemical composition be changed to effect a certain change in tensile properties; or what factors cause scrap castings; or what effect does molding sand condition have on losses; and many more that could be enumerated. Too often these problems are solved by reason that a certain thing was done in the past so it should be done now. Frequently, the conditions are known to have changed but for lack of anything better, the same analysis is applied. In light of the success correlation techniques have had in so many fields, it seems within reason to apply it to foundry problems.

Before going into some applications in the industry, it might be advisable at this point to discuss what is correlation and what does it involve. If we take for an example one ladle of iron, analyze it for chemical

composition and temperature, then test it for some quality such as fluidity by spiral length, we can say: For a given temperature, carbon percentage, and silicon percentage, we have a spiral of a certain length. However, if we repeat our work at another time, it would be the rare case should the new data exactly fit our previous statement, regardless of the attempt to keep all conditions exactly the same as in the first test. This variation in results between the two ladles of iron is to be expected mainly for two reasons: One, the so-called chance errors, or changes from one test to the next in factors not measured or considered in the analysis; the other cause, the accidental errors of reading the calibrations of the apparatus. This does not mean the observer has made an actual error, but that he must estimate between calibration marks. For example, one observer might estimate the carbon analysis as 2.51 per cent and other observer as 2.52 per cent. By taking a number of observations, these variations will cancel out to some extent and we take the remainder into account as will be shown later.

In the statement above about fluidity being related to the temperature, carbon, and silicon, we can draw no conclusions as to the relative effect of varying any or all of these factors. If, however, we continue to test ladles of iron as before until we have a set of data, we can evaluate these relative effects. We are quite accustomed to thinking of two or more things as being related when one of them can be expressed as a mathematical function of the others. It is in finding this mathematical equation that correlation and regression analysis are useful. From the evaluated equation, it is possible to determine the relative effects of each of the independent variables on the dependent variables. Take, for example, the algebraic expression " $x = 4y + 2z$ "; holding  $z$  constant and changing  $y$  by one unit will effect a change of four units in  $x$ ; or holding  $y$  constant and changing  $z$  by one unit, will change  $x$  by two units. For the general case, the change in the independent variable multiplied by the coefficient of the independent variable gives the change in the dependent variable. Also, it is within limits of this equation that predictions can be made as to the fluidity of any ladle of iron, provided, of course, that the test conditions have remained

\* National Malleable and Steel Castings Co., Cleveland, Ohio.

the same except for the expected chance variations.

In the problems that arise with but two variables, we can plot the data on a graph and get some idea of the relationship just by observation. Also, with three factors the data might be plotted in some sort of a three-dimensional model. However, we cannot conceive of more than a three-dimensional system and the observation method breaks down beyond three variables. One distinct disadvantage of fitting in a line of data by observation is that no two persons will interpret the data the same way as to where the line should fall. By correlation and regression analysis, we can handle any number of variables, being limited only by the amount of calculation time deemed warranted in the investigation. Problems with as many as sixteen variables have been successfully handled. The number of factors to be used in our investigation can usually be limited by experience to three or four and thus keep the calculations from becoming too involved or lengthy.

Whether the problems encountered in the foundry have two or more variables requiring either simple or multiple correlation, they are, as likely as not, to be curvilinear, i.e., the dependent variable may vary with the independent variable according to some curved relationship. Often times it is better to sacrifice some accuracy by using a straight line rather than a curve to express the relationship as some of the usefulness of operating personnel is lost when the calculation of the mathematical function becomes burdensome. The actual techniques of analysis are beyond the scope of this paper, but there are a number of reference books available.<sup>1 and 2</sup>

To be of value the results must be tested for significance and put into a form readily understood by management and operating personnel. To the uninitiated, the technical language of correlation will only tend to confuse the issue, and the analysis is likely to be discarded as too complicated.

The aim of this paper is to present a few problems encountered in the foundry which lend themselves to correlation analysis in an effort to show the applicability of this technique.

The examples chosen are:

1. Effect of delays on combined losses.
2. Change in fluidity of iron from furnace to pouring station.
3. Effect of temperature, carbon and silicon on fluidity.

An attempt is made to show the existing relationships and value of the analysis, avoiding, as much as possible, all technical methods and terms.

#### Effect of Delays on Combined Losses

A problem of great importance to management is the influence of delayed or non-productive time on scrap losses. In this illustration, the delayed time is a daily average for the month of the hours lost in molding for all reasons (no cores, no iron, mechanical breakdown, etc.) and has the units of average daily molder hours. The combined loss is the monthly percentage of malleable iron production scrapped, both in the white iron and the annealed state.

A plot of the data (Fig. 1) shows a relationship between the variable to be quite possible, and from the distribution of the points, it is apparent that should a relationship exist, a straight line will fit as well as the simple curves (parabola, hyperbola, etc.).

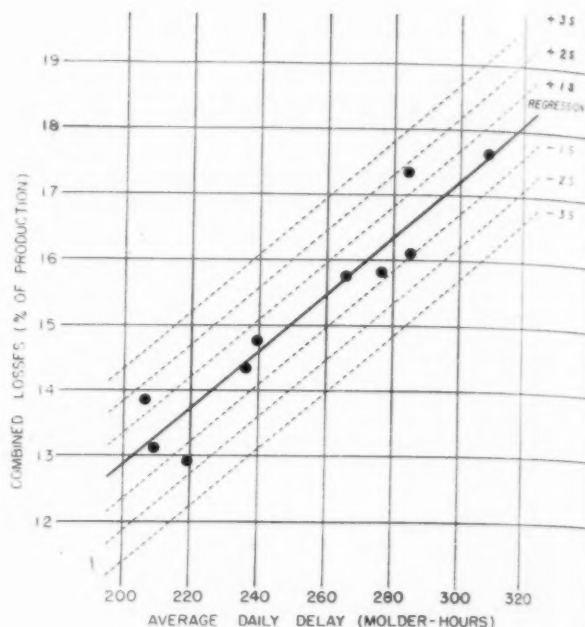


Fig. 1—Regression curve of combined losses as a function of delay time.

A correlation coefficient for the data of ten months was calculated and found to be 0.95. This in itself is of no value without an expression of its significance. It is found that in 95\* out of 100 times there is a definite correlation, and the coefficient will be at least as high as 0.85.<sup>1</sup> Both of these values, when tested for significance, show almost no chance whatsoever of the two variables not being related, or conversely for all practical purposes, they will be related 100 per cent of the time.

Knowing that the two variables are related gives us permission to evaluate a mathematical equation of the relationship, so that a practical use may be made of the analysis.

The line best fitting the data was calculated and superimposed on the graph of the data (Fig. 1). This line is known as the line of regression of the independent variable on the dependent variable. Should all our data fall exactly on the line, we could predict that for a certain delay time we can expect a certain loss. Upon observing the graph, this is found not to be the case, as variations are noticed and are to be expected from the reasoning given forth in the preceding material. It is possible, however, to calculate limits within which we can make predictions and expect to be right a certain percentage of the time. A value known as the standard error of estimate.

\* The 95 out of 100 times bears no relation to the correlation coefficient of 0.95. That the numerical values are identical is a coincidence.

relationship and from that, the standard error of estimate, was computed from the data, and found to be 0.50 per cent.

The meaning of this standard error might be best explained by using a hypothetical case on our graph. Should our average daily delays for the month be 220 molder-hours, we find the theoretical loss to be 13.75 per cent. Applying one standard error of estimate, both plus and minus the 13.75 per cent, we get values of 13.25 per cent and 14.25 per cent. Sixty-eight per cent of the time for a delay time of 220 molder-hours, we can expect to have losses between 13.25 per cent and 14.25 per cent. This, however, does not appear so practical, as we wonder what is going to happen to the other 32 per cent of the time. If we take two standard errors on each side of the line, this will include 95 per cent of all cases, and three standard errors will include 99.7 per cent, or practically all. Thus for 220 molder-hours delay, there is almost no chance of the loss not being between 12.25 per cent and 15.25 per cent. These limit lines are the dashed lines labeled "1S," "2S," and "3S." For an increase of 10 molder-hours in the average delay, we can expect an increase of 0.45 per cent in losses.

There is much yet to be done in the way of a complete analysis of this problem. For example, we might wish to know which defects making up the losses are influenced by delays and to what extent they are influenced. Or, we might investigate the relative effect of the losses by the different causes of delays. This is not meant to be a complete investigation of the problem, but merely an example of the usefulness of correlation techniques.

#### Change in Fluidity from Furnace to Pouring Station

The question arose as to what happened to the fluidity of the iron, from the time it was run into the ladle at the furnace spout, until it was poured into the molds at the pouring station. There was reason to believe that furnace conditions were not wholly responsible for occasional runs of low fluidity iron, the result of which led to this investigation. Spirals were poured at the furnace spout, and also at the pouring station, into the same type of standard core. After forty such tests were accumulated, the correlation coefficient was calculated and found to be 0.58. Significance tests show a definite correlation in 95 out of 100 times, with the coefficient being at least as high as 0.37. Further testing shows practically no chance of the data not being correlated when the coefficient is as high as 0.58, and only about 2 out of 100 times will they not be correlated when the coefficient is as low as 0.37.

Computing the line of regression gives the equation: Length at Pouring Station =  $-1.72 + 0.68 \times$  Length at Furnace Spout. Thus for an increase of one inch in the furnace spiral, we can expect an increase of 0.68 inches at the pouring station.

The standard error of estimate of the pouring station spiral is calculated to be 1.70 inches, which means, to include practically all the cases we must add and subtract from the regression line, a value of three standard errors or 5.10 inches. This fact, along

with the observation of the plotted data (Fig. 2) indicates a wide variation or spread to be expected in the results. The numbers adjacent to the points in Fig. 2 are the number of samples having the coordinates or values of the plotted points. For example, there were eight samples having a furnace spiral of 16 inches and a pouring station spiral of 10 inches.

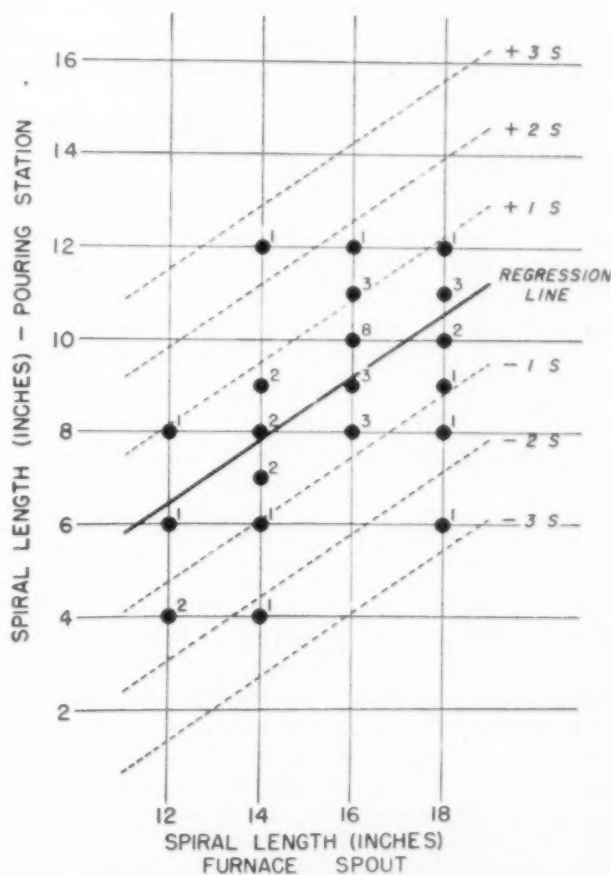


Fig. 2—Regression curve of pouring station spiral as a function of furnace spout spiral.

In this problem, as in any other problem of this nature, some engineering judgment must be applied to the results. Statistics alone will never provide a complete solution to the situation. In this case, upon investigating the method used in pouring the spirals, several things were found to exist that might introduce chance errors. The one felt to be most important was that the observers were not too consistent in holding a constant time for the moving of the iron from the furnace to the pouring station. Thus there was a possible chance error, in that the test on the pouring floor was not of the same iron that was tested at the furnace. Also, the conditions of the ladle were not taken into account, and such things as one ladle being colder than the others would bring in variations. Had there been a great many more than the 40 observations made, this chance error would have had more of an opportunity to cancel itself out and reduce the variation. It is quite possible to compute the number of variations required to reduce this variation to a desired value; however, even with this com-



putation we must use judgment in deciding between the results wanted and the economics of taking a great many observations. A laboratory analyst often takes but a very few samples and works with a high degree of precision, while someone like an insurance company statistician might take upwards of a million people to be sure his results will be representative of the population. In general foundry practice we are not able to work with high precision, nor are we able to take thousands of samples. We, therefore, must decide on a medium that will give acceptable accuracy with a minimum of samples.

The conclusions to this investigation must be that we are definitely sure of a correlation between furnace spiral and ladle spiral, but the variation is of an order and magnitude larger than desired. The analysis should be repeated in an attempt to reduce the variation and make the results more practical. In the repeat analysis, it would be wise to take more samples than 40; possibly as many as 100, and to more closely supervise the observation in order to obtain a higher degree of precision.

#### Effect of Temperature, Carbon and Silicon on Spiral Length

This investigation was run as a repeat analysis to determine whether the melting process had remained substantially the same over a certain period of time. At the time of the original analysis, it was found that fluidity was definitely dependent on temperature and carbon, but not too dependent on silicon over the range of the process. For this study, 21 spiral samples were taken at the furnace spout, an optical pyrometer temperature measurement being taken at the time of each sample. The samples also went to the chemistry laboratory for quantitative analysis as to carbon and silicon content. Assembly of the data and computation of a correlation coefficient gave a value of 0.245. Testing this value for significance shows it to be in the region of uncertainty; i.e., we cannot be sure that there is or is not a correlation. The test of the coefficient revealed that there would not be a correlation 7 times out of 100. A value of 5 out of 100 has been in common practice; but since this result was so close, it is not safe to draw a conclusion on the basis of the significance test. The fact that this analysis had been run in the past and a correlation found was enough of an indication to conclude that a correlation existed in the present data.

Upon evaluating the equation, however, radically different results were obtained in the effect of temperature and carbon on spiral length or fluidity. The results of the first investigation showed 0.01 per cent of carbon to produce the same effect as 1° F change in temperature, while the present analysis required 22° F to equalize the effect of 0.01 per cent change in carbon. On the strength of this difference, it was advisable to discount the results of the second investigation.

No definite conclusion can be drawn as to the reason for the great difference in results between the two analyses, except for two facts worth noting. First, the spiral in the first investigation was poured into a green sand mold, while a dry sand mold was used in the

second case. Also in the first investigation 100 samples were taken with but 40 in the second. It has been shown that the number of sample has a considerable effect on the variation; the variation decreasing with an increase in sample size. As to whether the difference in these two cases was due entirely to too few samples, or entirely to the data not being good enough, or to a combination of both, we cannot say. We can say, however, that a correlation exists; but the evaluated relationship does not warrant its practical use and the investigation should be repeated with an effort toward reducing the chance errors.

#### Conclusion

In closing, it might be said that neither the examples nor the discussion is meant to be a complete treatise on correlation techniques applied to a foundry, but merely an effort to introduce some of the possibilities.

The examples were chosen to show that correlation does not give a definite yes or no answer to a problem but provides us with a tool that may be profitably used in judging a situation.

The reader can well imagine many more places in his own plant where such applications might exist.

The author wishes to thank Dr. H. A. Schwartz and W. K. Bock for their assistance in reading the manuscript, and in furnishing some of the material presented.

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#### DISCUSSION

**Chairman:** C. F. JOSEPH, Central Foundry Div., General Motors Corp., Saginaw, Mich.

**Co-Chairman:** ERIC WELANDER, Union Malleable Iron Works, Deere & Co., East Moline, Ill.

**CHAIRMAN JOSEPH:** We have done work along this line in the past. We probably did not plot the same curves that the author has, but similar work has been carried on. We started some work a few years ago on the effect of the moisture content of the air going into the cupola. We plotted a number of curves showing the effect on scrap losses as the moisture content of the blast increased from 4 gr per cu ft up to 10 gr per cu ft. We also plotted curves from 4 gr per cu ft down to 1 gr per cu ft, and this was done over the period of a year or more. All this information was plotted and it definitely showed that the scrap losses were higher as far as shrinks and cracks were concerned when the moisture content of the air was from 4 gr per cu ft up to, say, 10 gr per cu ft. We hold it at 4.

This matter of correlation is important, to be able to trace losses in the foundry. Another application where it is used every day is in the plotting of various types of scrap losses which occur with the changing of certain raw materials.

**HARRY GRAVLIN:**<sup>1</sup> We are able to confirm the same results on absolute humidity vs porosity on camshaft castings. Our results were obtained at a moisture content of less than 3½ gr per cu ft. It is interesting to note that the author has done the same type of work. On the basis of our experimental record over a period of a year, we are installing one unit for humidity control in our cupola blast. This unit is designed to handle a blast of 11,000 cu ft at a minimum moisture content of 1 gr per cu ft. We hope that it will confirm experimental results as to the correlation between scrap and moisture in blast.

<sup>1</sup> Ford Motor Co., Detroit

# STEP AGING OF A MAGNESIUM-BASE CASTING ALLOY

By

E. J. Vargo<sup>1</sup> and G. Sachs<sup>2</sup>

## ABSTRACT

The tensile strength of the magnesium-base sand casting alloy AZ63 in the solution heat treated and aged condition, HTA, has been found in commercial practice to be slightly lower than that in the solution heat treated condition, HT, in contrast to the Specification AN-M-36. This discrepancy can be explained on the basis of the stress-strain curves for these two conditions, considering the fact that the commercial quality is slightly inferior to that obtainable in laboratory tests.

In order to improve the properties or to reduce the time required for aging, the properties of test bars were determined (a) after aging at various temperatures and (b) after "step-aging" at two different temperatures, first at 250 F and then at 350 or 400 F. Both laboratory tests and trial production runs revealed that by replacing regular aging by step aging, (a) the strength and elongation can be raised distinctly, for a given aging time, or (b) the aging time can be reduced significantly for a given combination of properties.

## Introduction

THE PRESENT INVESTIGATION originated in two observations made in the magnesium foundry of the Wellman Bronze and Aluminum Co., Cleveland, Ohio, over a period of several years. They relate to the most common heat treated magnesium casting alloy, containing six per cent aluminum, three per cent zinc, and 0.2 per cent manganese (AZ63). This alloy is covered by the Army-Navy Aeronautical Specification AN-M-36 which refers to this alloy as composition "A" and specifies the properties assembled in Table 1 as minimum values for two conditions of heat treatment, the heat treated (HT) and the aged (HTA) conditions.\*

The first plant observation relates to the respective properties of these two conditions. In commercial practice, little difficulty is encountered in meeting the properties specified in Table 1 for either condition. Also, the increase in yield strength and the decrease in elongation by the generally used aging

TABLE 1—AN-M-36 SPECIFICATION PROPERTIES OF ALLOY STUDIED

Condition	Tensile Strength, psi	Yield Strength, psi	Elongation in 2 In., Per Cent
HT	32,000	10,000	7
HTA	34,000	16,000	3

treatment at a temperature of 350 F for 16 to 18 hr corresponds to that derived from the specifications. However, the difference of 2000 psi in the tensile strength values of the aged and solution heat treated conditions respectively, expected from Table 1, was not verified in the foundry. On the contrary, as illustrated by frequency curves for a large number of tests, Fig. 1 to 3, the tensile strength of the aged condition was found to be, on the average, more than 500 psi lower than that of the heat treated condition. This tensile strength was rather close to 39,000 psi, or slightly lower than the value of 40,000 psi reported by Nelson,<sup>1</sup> for both conditions.

The second plant observation was that, as the result of some difficulties with temperature control, certain heats were subjected to "step aging," consisting of aging at 250 F for three hours, followed by aging at 350 F for 13 hr. The strength values for such specimens appeared to exceed those previously encountered. As a consequence of this observation, the entire production was subjected tentatively to step aging. The tensile strength of test bars in this step aged condition was found to be decidedly higher than that of normally aged test bars, and also of those in the heat treated conditions, Fig. 1 to 3. A comparison of the production for consecutive time intervals consisting of one month each, in which the aging treatment was different, also showed that the elongation was considerably increased by this step aging, while the yield strength remained practically the same. The test bar properties prior to aging were, according to Fig. 1, identical in these time intervals. Therefore, the observed changes in the aged condition were caused by the change in the aging treatment.

In order to substantiate these observations, extensive

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<sup>2</sup> Director, Research Laboratory for Mechanical Metallurgy and Professor of Physical Metallurgy, Case Institute of Technology, Cleveland, Ohio.

\* The designation "heat treated" used throughout this paper applies to the solid solution heat treated condition, and the designation "aged" to the solution heat treated and artificially aged condition.

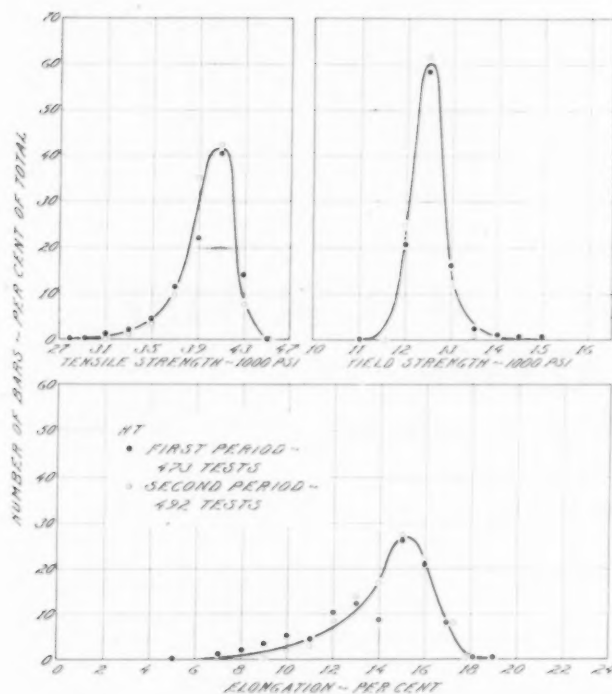


Fig. 1—Frequency curves for solution treated (HT) test bars of a magnesium casting alloy during two consecutive time periods (each test represents a different melt).

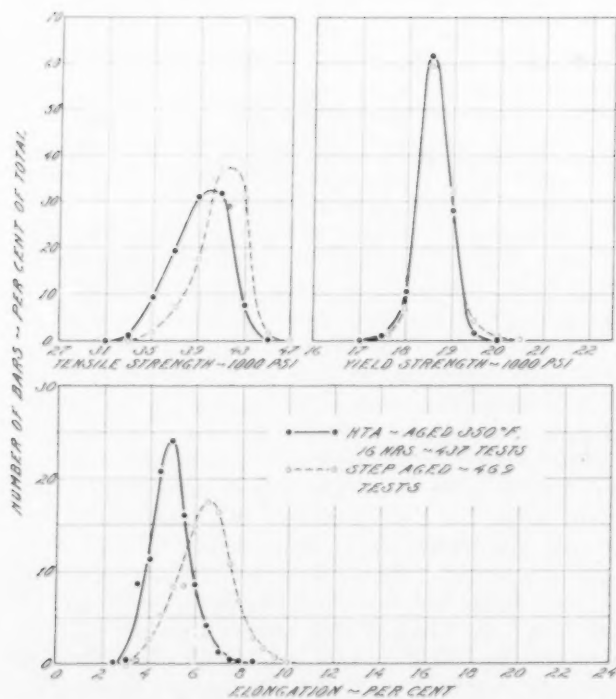


Fig. 2—Frequency curves for conventionally aged (HTA) and step aged (250 F, 3 hr; 350 F, 13 hr). Test bars of a magnesium casting alloy (each test represents a different melt).

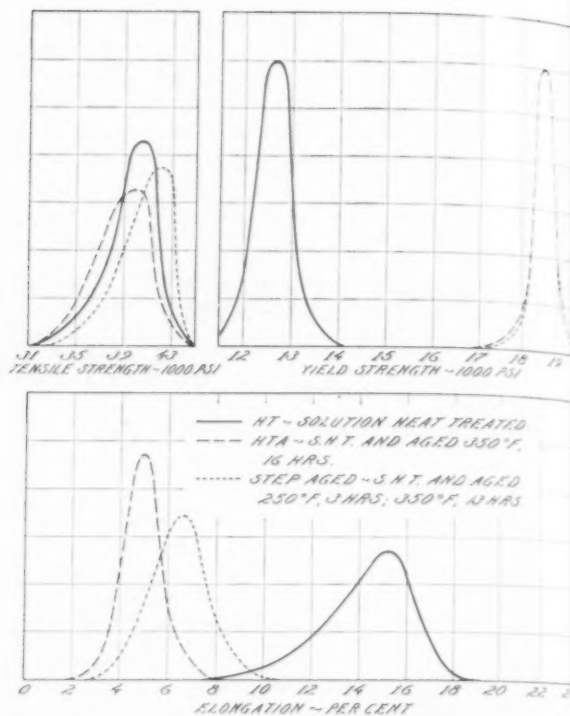


Fig. 3—Comparison of the frequency curves for a magnesium casting alloy (AZ63) in different heat treated conditions.

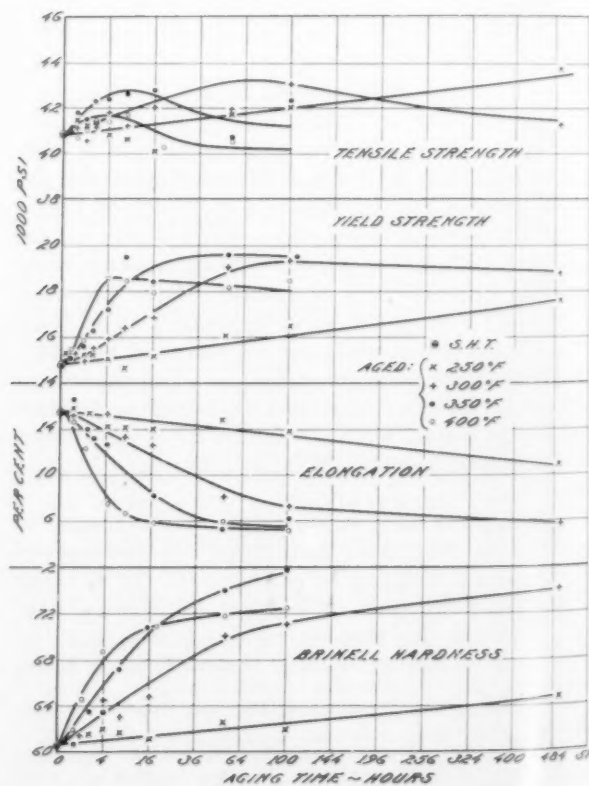


Fig. 4—Effects of aging at various temperatures on the properties of a magnesium casting alloy (6% Al, 3% Zn, 0.2% Mn).



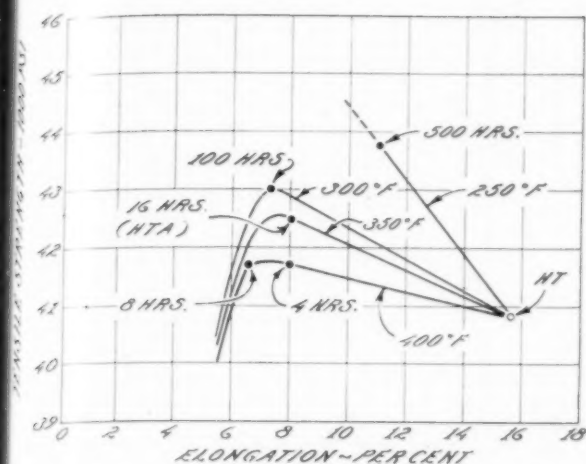


Fig. 5—Combinations of tensile strength and elongation obtainable on aging at various temperatures for a magnesium base casting alloy.

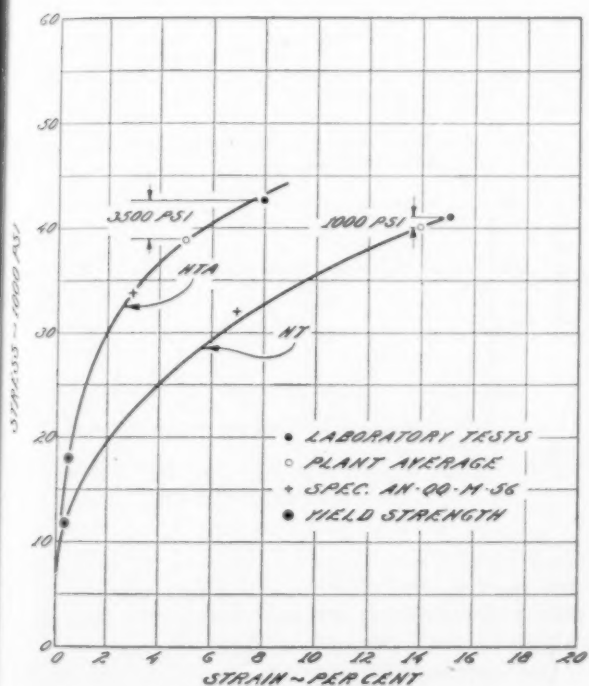


Fig. 6—Stress-strain curves showing various combinations of strength and elongations for the heat treated (HT) and the aged (HTA) conditions of a magnesium base casting alloy.

laboratory tests were initiated. The object of this investigation was two-fold. First, the actual changes in tensile strength on aging the magnesium casting alloy were to be determined, and the discrepancy reported above was to be explained. Second, the type and magnitude of the benefit which may be derived from a step aging treatment was to be determined.

#### Material and Procedure

Various heats of a magnesium casting alloy containing 6 per cent aluminum, 3 per cent zinc, and 0.2 per

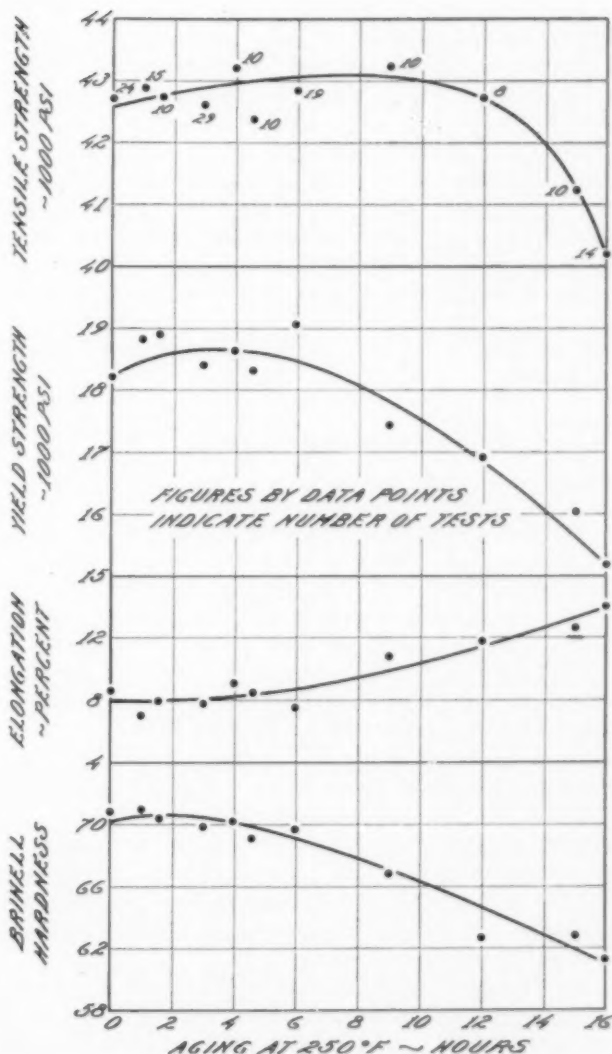


Fig. 7—Effects of two step aging at 250 and 350 F for a total time of 16 hr.

cent manganese were used for this investigation. This alloy is covered by the specifications AN-M-36. It carries the ASTM designation AZ63, and is also known in industry as Dow H, or AM265 alloy.

Test bars were poured in the foundry of the Wellman Bronze and Aluminum Co. under controlled commercial conditions. The metal was refined at 1350 F with Dow 310 flux by vigorous stirring for 2 min. It was then superheated to 1650 F for 5 min, allowed to cool to 1480 F and, at this temperature, eight tensile test bar molds were poured off from one crucible. The molds were of the two-bar type, and were made of green sand.

The solid solution heat treatment consisted of inserting the test bars into an air convection type furnace at 500 F, holding at 640, 670, 690 and 710 F for one-half hour each, and at 730 F for 12 hr. The test bars were then removed and quenched in an air blast.

For the aging treatment, the test bars were inserted into the furnace at the aging temperature. In the case of two step aging, they were left in the furnace while the temperature was raised.

Test bars which contained visible porosity or inclusions, which exhibited a large grain, or the microstructure of which revealed the presence of either beta phase, either in the form of massive white areas or in the form of lamellae were rejected.

Stress and strain values reported here are (with very few exceptions) averages of at least 10, and up to 24 individual tests. In a preliminary test it was found that the average tensile strength value obtained from four test bars conformed to the expected ideal value within  $\pm 500$  to 1000 psi. Even if a considerably large number of test bars was used to obtain average values, the accuracy of the test values appeared only slightly improved.

### Results and Discussion

#### Effect of Regular Aging

To determine the basic effects of regular aging treatments, four series of tests were made. Solution heat treated test bars were aged for various times up to 500 hr, at temperatures of 250, 300, 350 and 400 F.

The effects of aging temperature and time are illustrated in Fig. 4. The change in properties for this alloy follow the well known general pattern. In agreement with many other alloys, slightly better combinations of tensile strength and elongation are obtained, the lower the aging temperature.<sup>2</sup> However, this relation cannot be commercially utilized to any large extent, because very long aging times are required at temperatures sufficiently low to yield an advantage. Within a practically feasible aging time of less than 24 hours, the lowest temperature to produce satisfactory aging is 350 F.

To illustrate the differences in the effects of various aging temperatures, the test data assembled in Fig. 4 are replotted in Fig. 5, using the elongation as abscissa and the tensile strength as ordinate. Because of the small magnitude of these differences, which do not always exceed the variations in the test data, the accurate position of the various curves in Fig. 5 is uncertain. However, this graph clearly illustrates two facts. First, as discussed above, the highest strength attainable at a certain aging temperature decreases with increasing temperature. Second, increasing the time (going to left in Fig. 5) beyond that required for aging to maximum strength (i.e., overaging) causes considerably less favorable combinations of strength and elongation than shorter aging (i.e., underaging).

According to the results of the laboratory tests in Fig. 4, an increase in tensile strength of at least 1500 psi over that of solution heat treated test bars should be obtained on aging at a temperature of 350 F for 16 hr. These tests yielded an average tensile strength of approximately 41,000 psi and an average elongation of 15 to 16 per cent for the heat treated condition, while the tensile strength was approximately 42,500 psi, and the elongation 8 per cent for the bars aged for 16 hr at 350 F. However, it must be considered that the test bars which yielded these results were cast under close control. On the contrary, the less rigidly controlled casting of test bars in ordinary commercial practice resulted, according to Fig. 3, in an average tensile strength of approximately 40,000 psi for the

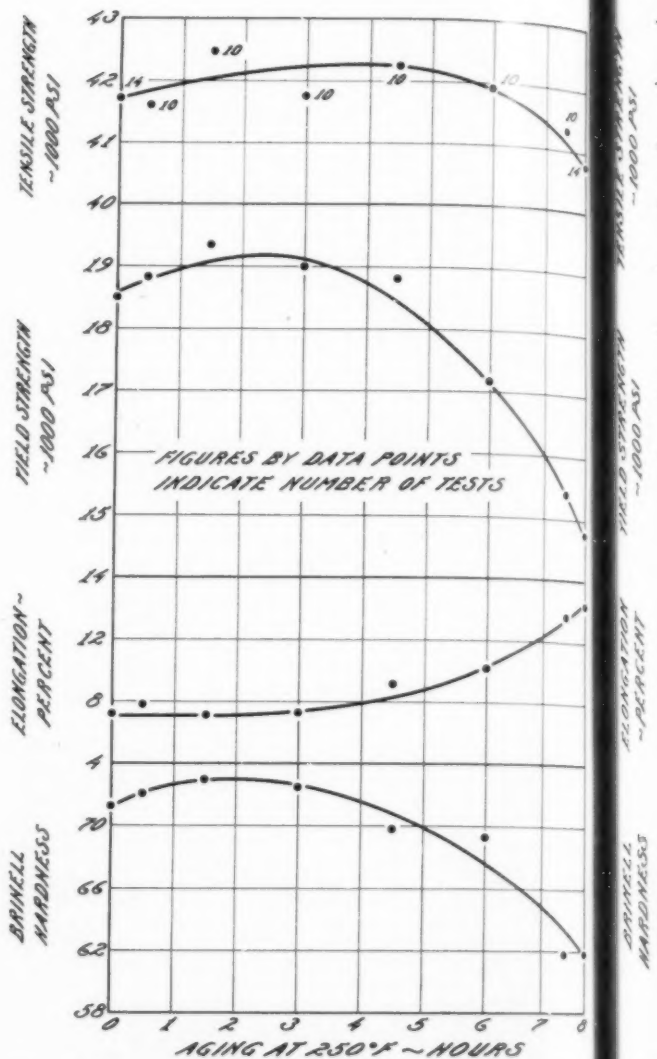


Fig. 8—Effects of two step aging at 250 F and 400 F for a total time of 8 hr.

heat treated, and 39,000 psi for the aged condition, with corresponding average elongations of approximately 14 and 5 per cent, respectively.

The apparent discrepancy between these results is readily explained by the respective differences in elongation. Each of the two heat treatments is characterized by a well defined stress-strain curve. These are shown in Fig. 6, as derived from a number of tests. The cast quality of a test bar does not affect the position and shape of its stress-strain curve, but only determines its terminal point. The coordinates of this terminal point are the elongation and the tensile strength.\* For the heat treated condition, the elongation of 15 per cent observed in the laboratory tests thus corresponds to the observed tensile strength of 41,000 psi. The plant average elongation of 14 per cent then determines a tensile strength of approximately 40,000 psi, which is only 1000 psi lower than the laboratory value. For the aged condition, the elongation of 8 per cent observed in the laboratory

\* This applied only to metals which do not exhibit necking in tensile tests.

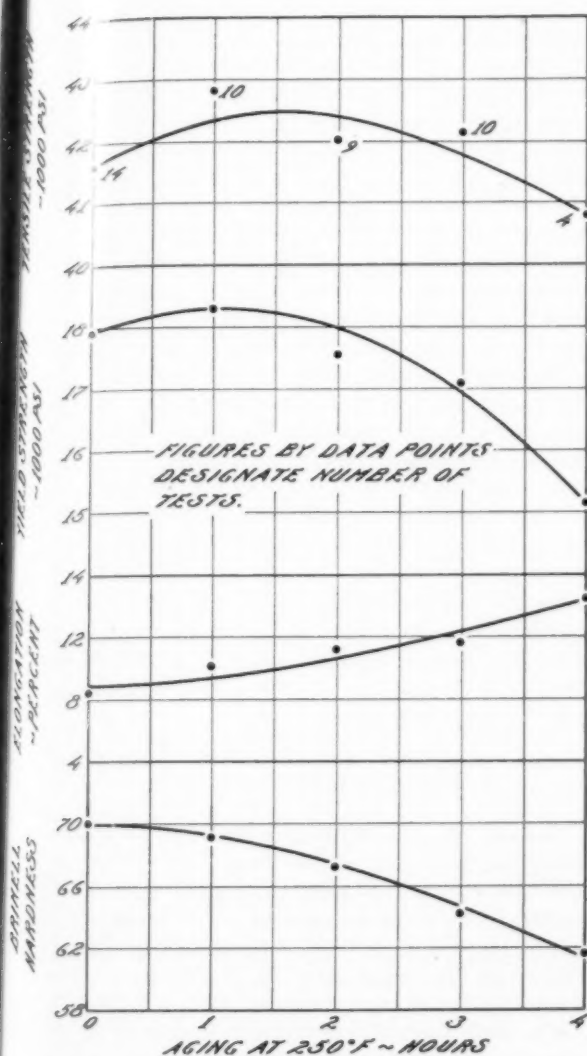


Fig. 9—Effects of two step aging at 250 F and 400 F for a total time of 4 hr.

### Effects of Step Aging

In order to investigate the effects of step aging and to develop optimum treatments, the following three series of tests were made:

Series I—Total aging time 16 hr, first at 250 F and then at 350 F.

Series II—Total aging time 8 hr, first at 250 F and then at 400 F.

Series III—Total aging time 4 hr, first at 250 F and then at 400 F.

In each series, the first aging time, at 250 F, was varied between zero hours and the total aging time.

The results of the tests are assembled in Fig. 7 to 9. The first (left) point on each curve corresponds to a possible regular aging treatment. In Fig. 7, it corresponds to aging at 350 F for 16 hr, in Fig. 8, to aging at 400 F for 8 hr, and in Fig. 9 to aging at 400 F for 4 hr. In each series, a better combination of properties was obtained by a certain step aging treatments, in which a fraction of the total aging time was consumed at 250 F, than with aging at a single temperature. This is further illustrated in Fig. 10, which conforms to the previously discussed Fig. 5.

In order to evaluate the benefit which may be obtained from step aging, it must be first considered that a substantial reduction in yield strength cannot be tolerated. This restricts the range of treatments, the results of which are shown in Fig. 7 to 9, such that the aging time at 250 F should not exceed one-half of the total time.

Various step aging treatments conforming to this requirement of maintaining the yield strength cannot be derived from the tests conducted thus far. In general, for given aging temperatures and total aging time, step aging appears to result, according to Fig. 7 to 9, in an almost constant or slightly increased yield strength and tensile strength, while the elongation may be raised up to 1 or 2 per cent above the values obtained on aging for the full time at the higher temperature.

Because of their small magnitude, these effects cannot be established quantitatively, in spite of the considerable number of test bars which supplied the average properties, plotted in Fig. 7 to 9. Consequently,

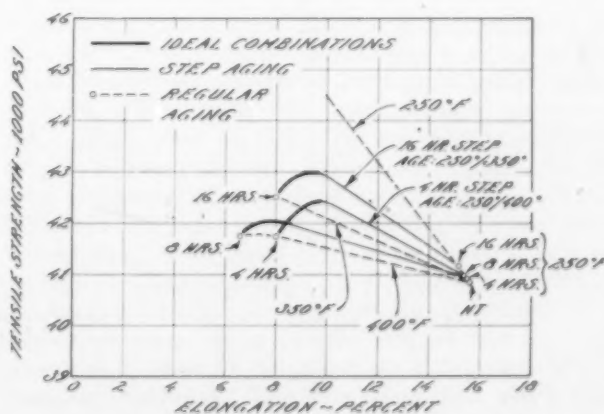


Fig. 10—Combinations of tensile strength and elongation obtainable on step aging for a magnesium base casting alloy.

establishes a tensile strength of approximately 42,500 psi. The plant average elongation of 5 per cent then corresponds to a tensile strength of 39,000 psi, or 3500 psi lower than the laboratory value.

The above mentioned observed differences in elongation between laboratory and plant tests are probably not very accurate, since their magnitudes are rather small. However, the discussed relations would also apply if the elongation of both conditions of heat treatment decreased by an equal amount because of imperfections in the cast structure.

Thus, a decrease in the cast quality reduces the tensile strength of the aged condition considerably more than that of the heat treated condition. The combinations of minimum strength and elongation from Table I are also added to Fig. 6. This graph clearly illustrates that the allowance made by the specifications for an interior casting quality is considerably larger than for the heat treated (HT) than for the aged (HTA) condition. Consequently, specifications should allow for the minimum tensile strength of the aged condition a smaller rather than a larger value than that of the heat treated condition.



conclusions drawn from the results of the laboratory tests contain a considerable factor of uncertainty. Such statements, therefore, must be considered primarily as suggestions for further work on a more elaborate basis than was possible here. It appears that the test results will yield the following relations, with the above mentioned restrictions:

1. Of the three series investigated, step aging at 250 and 350 F (Series I), Fig. 7, yields the best combination of properties. However, if for a given series and improvement achieved by step aging, over the properties obtained on aging at the higher of the two temperatures (for the same total aging time) is considered, step aging at 250 and 400 F for a total aging time of 4 hr (Series III), Fig. 9, appears to yield the largest effect. On the other hand, it appears that aging at 250 and 400 F for a total time of 8 hr (Series II), Fig. 8, yields properties only slightly higher than those obtained on aging at 400 F for 8 hr. This must be attributed to overaging.

2. It must be mentioned that the values determined in production differ from those of the laboratory tests in one significant respect. The commercial step aging treatment which was selected on the basis of a few plant observations consists of aging at 250 F for 3 hr and at 350 F for 13 hr. It yielded a considerable improvement in elongation (Fig. 2) which exceeds that expected from the laboratory tests (Fig. 7). No explanation can be given for this discrepancy.

#### Microstructures

In order to reveal the structural reasons for the difference in properties between regularly and step aged material, a microscopic investigation of a number of specimens was made. The heat treated condition consists mostly of uniform solid solution crystals, with some heterogeneous inclusions ( $Mg_2Si$ , Mn), both at the grain boundaries and within the grain. The microstructure of aged metal always shows a certain percentage of apparently retained solid solution crystals, while the remainder consists of precipitated areas similar in appearance to pearlite in steel. This grain structure is in agreement with X-ray investigations<sup>3</sup> which indicate that, during aging, an increasing portion of the solid solution decomposes (almost) to its equilibrium condition, while the remainder is subject to a very slow precipitation process.

The aging treatments which are considered most interesting commercially, are those which develop maximum tensile strength at any given aging temperature, e.g., 16 hr at 350 F and 4 hr at 400 F (Fig. 4 and 5). The microstructure of such conditions generally exhibits precipitated areas only over 25 to 50 per cent of the cross section. Metal which is subjected to aging for longer times than necessary to develop a maximum tensile strength is called "overaged." Such excessively precipitated metal apparently has minor commercial significance,\* because it possesses both lower tensile strength and elongation (possibly

associated with a higher yield strength) than the intermediate stages.

This microstructure of the partially (and completely) precipitated conditions after regular aging was found to be distinctly different for different aging temperatures. The lamellae in the precipitated areas were thinner and closer together the lower the aging temperature. At lower aging temperatures, the aging time appeared to be of little influence in these respects. At the higher aging temperature, long aging times resulted in some breaking up and coagulation of the lamellae. Furthermore, according to some microscopic observations by Hess and George,<sup>4</sup> low temperature aging causes a more uniform precipitation throughout the grains, or "continuous precipitation,"<sup>5</sup> rather than the gradual replacement of the suppressed solid solution by completely precipitated areas, or "discontinuous precipitation," described above.

The microstructure of specimens subjected to step aging exhibited distinctly finer lamellae than that of the material aged at the higher of the two temperatures used in step aging. Step aging for a given time resulted in a structure and in properties similar to those obtained at a temperature intermediate between the two temperatures of step aging after a considerably longer aging time.

#### Further Benefits from Step Aging

Step aging of magnesium alloys has already been found commercially interesting because of the improvement in both tensile strength and elongation which can be obtained for a given (or slightly reduced) aging time. The benefit derived from such aging is a decrease in rejections. In a properly controlled plant, this benefit is of comparatively little significance.

A considerably larger benefit would result from a substantial reduction in the total aging time. Certain conclusions in this respect can be drawn from a comparison of Fig. 7 to 9. As discussed previously, the time required for complete aging decreases with increasing aging temperature. Simultaneously, the properties of the fully aged material decrease. Consequently, the improvement derived from step aging may permit obtaining the same properties as in conventional aging (350 F for 16 hr) within a much shorter total aging time. From the tests discussed so far, for instance, it appears that on aging at 250 F for 1½ hr and at 400 F for 2½ hr, a tensile strength almost equal to that of commercially aged (HTA) metal, associated with a markedly increased elongation, may be obtained. The yield strength appears to be slightly lower than after aging at 350 F; however, this reduction is insignificant and can be tolerated.

In order to confirm this conclusion, a plant test was carried out in which 101 solution heat treated test bars of regular commercial production are step aged for 1 hr at 250 F and for 3 hr at 400 F. The time required to bring the furnace up from 250 to 400 F was 20 min, rendering the total aging time approximately 4½ hr. The properties obtained with this step aging treatment, represented by the frequency curves in Fig. 11, were found to be not only

\* This condition is used only if stability at elevated temperatures is of primary importance.

equal but distinctly superior to those of the commercially aged (HTA) production. For bars which were step aged, the tensile strength was the same as that for bars given the regular commercial aging, while both the yield strength and elongation were slightly higher. Thus a benefit in properties was derived from step aging in spite of a 75 per cent reduction in total aging time.

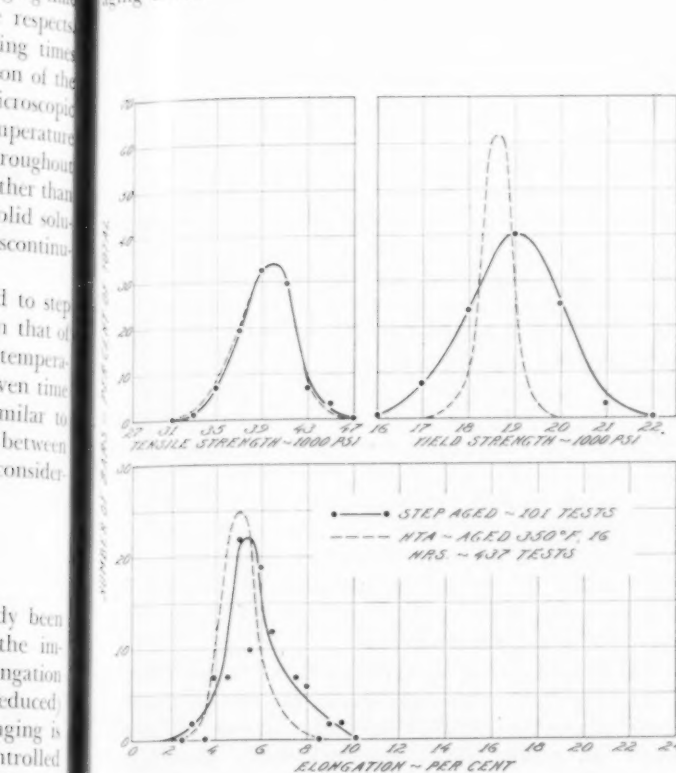


Fig. 11—Frequency curves for step aged (250 F, 2 hr; 400 F, 2 hr) test bars of a magnesium casting alloy.

It appears possible that some intermediate step aging can be developed which yields both a general improvement over the present practice and a saving in aging time. According to the results of this investigation other promising combinations of aging temperature may be 250 and 375 F, 275 and 375 F, or 225 and 400 F.

#### Conclusions

The following conclusions can be drawn from this investigation of the magnesium casting alloy AZ63:

1. Long time low temperature aging of this alloy produces a better combination of properties than that obtained with a short time high temperature aging treatment.
2. By means of a step aging treatment, a higher tensile strength and greater elongation can be obtained than those resulting from the standard commercial treatment (HTA) of the same length of time.

3. It appears possible with a short time step age to obtain better elongation and the same tensile strength as compared with the standard treatment, without loss in yield strength. A plant test along these lines yielded better properties than those predicted from the laboratory test.

4. The superior physical properties obtained with low temperature aging are attributed to the reduced size and spacing of the lamellar precipitate. Step aging results in a microstructure similar to that obtained in a much longer time at some temperature between the two temperatures of the step age.

#### Acknowledgments

The authors are indebted to the Wellman Bronze and Aluminum Co. for the support of this investigation. They also acknowledge the assistance and criticism by Mr. L. J. Ebert in preparing the manuscript.

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#### DISCUSSION

Chairman: L. Brown, Magnesium Fabricators Div., Bohn Aluminum and Brass Corp., Adrian, Mich.

Co-Chairman: A. T. Ruppe, Bendix Products Div., Bendix Aviation Corp., South Bend, Ind.

W. E. SCHA: I wonder whether the authors have given sufficient consideration to the over-aging effect that apparently is produced by treatment for 16 hr at 350 F. It appears from the curve in Fig. 4 that this treatment may be reducing the tensile strength and elongation below peak values. There is the possibility that this factor is playing an important part in some of the results that the author obtained with the step aging treatment. Evidence of the merits of the step aging treatment would have been more convincing if the authors had included some curves showing the effect of aging at 400 F without the preliminary 250 F aging. The question remains as to whether practically all of the aging was imparted by the 400 F portion of the treatment.

Let us consider the last treatment of 1 hr at 250 F plus 3 hr at 400 F in Fig. 4. What were the properties obtained with just 3 hr at 400 F? Did the authors gain anything by that preliminary hour at the lower temperature?

MR. VARGO: I believe that careful study of Fig. 4 and 9 will point out the benefit obtained by the preliminary hour at 250 F in the step age. In an investigation of this type the differences in results are rather small. Unless you test a great number of bars, it is extremely difficult to be absolutely sure that it is not just the natural difference in pouring one particular heat of test bars. We feel that with the large number of test bars tested over this period of the investigation we did show as accurately as possible the effect of step aging.

<sup>1</sup> Aluminum Co. of America, Cleveland

# DESIGNING STRAINER CORES

By

H. L. Campbell \*

## ABSTRACT

The author proposes seven standard shapes and sizes of strainer cores for the foundry industry. Strainer cores are employed to control discharge of metal from pouring basins or to regulate flow of metal in gating systems. The size and number of holes in a strainer core determine the rate at which metal will pass through the core under a given head of metal. Selection of a strainer core should be based on the required delivery rate of metal.

THE A.F.A. CORE TEST COMMITTEE has been making a study of the uses of strainer cores, with the ultimate purpose of proposing certain standard shapes and sizes for the foundry industry. When definite designs of strainer cores are generally adopted, the equipment needed to make these cores will be simplified. Pattern shops can then provide prints of the established dimensions on all pattern equipment. At jobbing foundries where large numbers of different patterns are received, strainer cores of standard sizes may be taken directly from stock and used without

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delay in production.

Strainer cores are employed to control the discharge of metal from pouring basins or to regulate the flow of metal in gating systems. If the stream of metal from a pouring ladle is not restricted, all non-metallic material is necessarily washed directly into the casting. By providing a restricted area in the gating system, any material lighter than the metal will be floated out and will not enter the cavity for the casting.

In the making of small molds, a strainer core is often placed near the bottom of each sprue. During the pouring of these molds the sprue is kept filled with metal so that any slag delivered with the metal will remain in the sprue. To aid in the pouring of large molds, a pouring basin is usually provided on each mold. By restricting the discharge of metal from the pouring basin, any slag from the ladle will be retained in the pouring basin.

Many designs of strainer cores are used regularly in the foundry. Although attempts have been made at some plants to simplify and standardize on strainer cores, no united effort has been made previously to

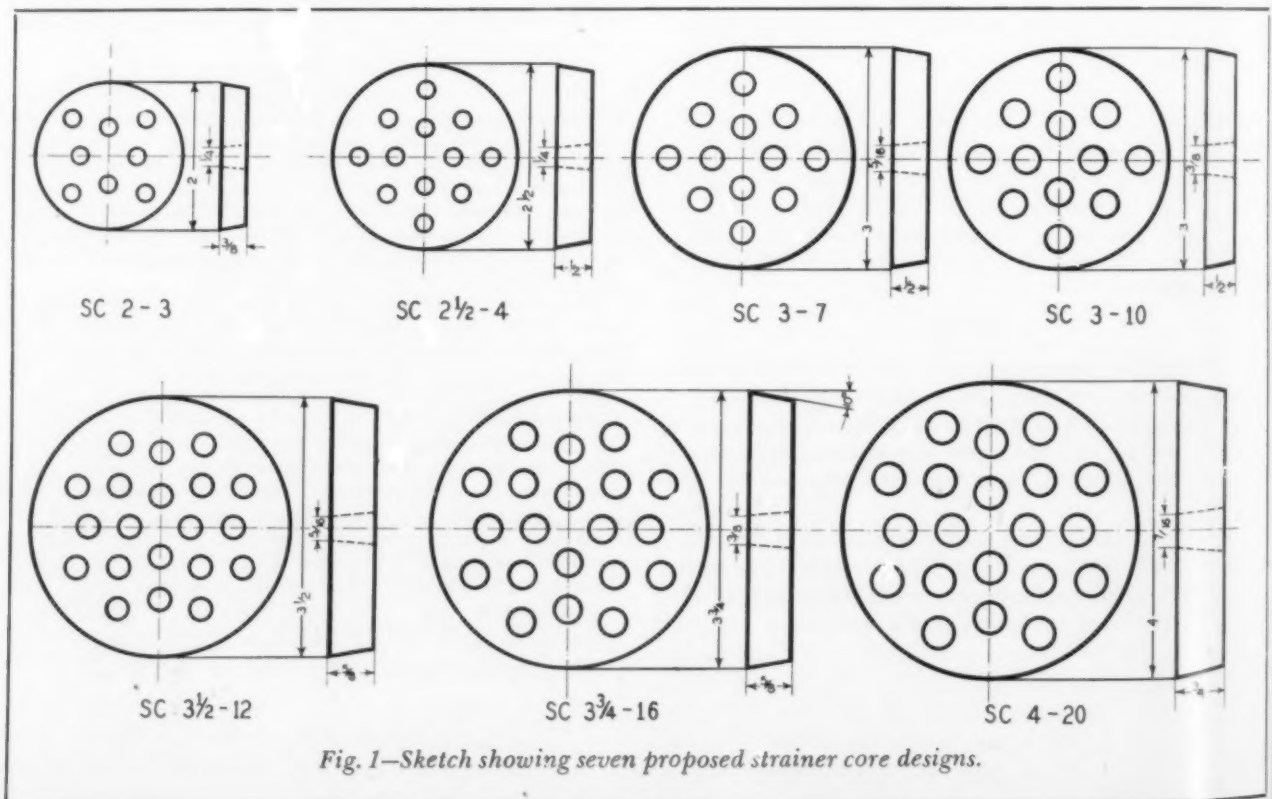


Fig. 1—Sketch showing seven proposed strainer core designs.



TABLE 1—STRAINER CORE DIMENSIONS

Designation	O.D., In.	Thickness, In.	Diameter Single Hole, In.	Number of Holes	Total Area of Holes, Sq. In.	Delivery*, Lb/Sec
SC 2-3	2	$\frac{3}{8}$	$\frac{1}{4}$	8	0.39	3
SC 2 $\frac{1}{2}$ -4	2 $\frac{1}{2}$	$\frac{1}{2}$	$\frac{1}{4}$	12	0.59	4
SC 3-7	3	$\frac{1}{2}$	$\frac{5}{16}$	12	0.92	7
SC 3-10	3	$\frac{1}{2}$	$\frac{3}{8}$	12	1.32	10
SC 3 $\frac{1}{2}$ -12	3 $\frac{1}{2}$	$\frac{5}{8}$	$\frac{5}{16}$	20	1.53	12
SC 3 $\frac{3}{4}$ -16	3 $\frac{3}{4}$	$\frac{5}{8}$	$\frac{3}{8}$	20	2.21	16
SC 4-20	4	$\frac{3}{4}$	$\frac{7}{16}$	20	3.01	20

\* Delivery rate determined by trial with cast iron at 2700 F under a head of 2 in.

establish standard designs of strainer cores for the benefit of the entire industry.

When liquid metal comes in contact with an oil-sand core, the bonding material is destroyed by the heat from the metal, which causes a gradual disintegration of the core. As the conductivity of heat through the core is relatively low, this destructive effect progresses slowly. To withstand this effect of the heat, the sections of a strainer core must be of sufficient thickness to hold together until the required amount of metal has passed through it.

After examining many designs of strainer cores now in use, it was concluded that  $\frac{1}{4}$  in. of core mixture between the holes was adequate for most applications. In other words, a strainer core should serve its purpose until  $\frac{1}{8}$  in. of the material around each hole is destroyed by the heat from the metal. Furthermore, the thickness of strainer cores should be based on the forces to which the cores are subjected. The larger the diameter, the greater should be the thickness.

The size and number of holes in a strainer core determine the rate at which metal will pass through the core under a given head of metal. Openings smaller than  $\frac{1}{4}$  in. in diameter are usually not necessary for strainer cores used in molds for gray iron castings. The holes may be straight or tapered depending on the equipment used in making the cores.

### Strainer Core Designs

Seven designs of strainer cores, as shown in Fig. 1, are proposed for the gray iron industry. The dimensions and delivery of metal through the cores are given in Table 1. Each design of core is designated by its maximum diameter in inches and the delivery of metal in pounds per second. In the development of this series of sizes of strainer cores, the plan was to provide a sufficient number of steps in the delivery rate to satisfy all requirements. At least  $\frac{1}{4}$  in. of core material lies between any two holes in each design. Also, a sufficient width of core is provided for the print. The thickness of strainer cores was fixed on the basis of designs now used satisfactorily.

As the purpose of a strainer core is chiefly to control the amount of metal which passes through it, the selection of a strainer core should be based on the required delivery rate of the metal. If it is desired to pour 200 lb of cast iron in 20 sec, a strainer which will deliver 10 lb of metal per second will be needed.

## DISCUSSION

Chairman: J. A. RASSENFOSS, American Steel Foundries, East Chicago, Ind.

Co-Chairman: E. C. ZIRZOW, National Malleable & Steel Castings Co., Cleveland

CO-CHAIRMAN ZIRZOW: We are all more or less familiar with strainer cores. They are used in gray iron and malleable shops primarily to control the flow of metal, but also to prevent dirt and foreign material from entering the casting.

In the beginning of this investigation we tried to get an idea of how many different types of strainer cores were now in use. We found that there were as many different types of strainer cores as there were foundries contacted. Some foundries used as many as 15 or 20 different types.

MEMBER: It is stated in the paper that the holes may be straight or tapered, depending upon the equipment used in making the cores. The tapered hole will pass more iron than a straight one. It depends upon which way you set the taper of the strainer core in the basin. In the upside down position, it will pass more iron than in the right side up position. The molder will discover that quickly and will turn the core upside down and get the iron through.

MR. CAMPBELL: In our experiments on delivery rate, we had a plain straight-edged orifice. There are no doubt, some factors involved in the shape and sharpness of the edges as well as in the thickness of the core that would affect delivery rate to some extent. This would be relatively minor as far as we are concerned. We cannot possibly get extreme accuracy because there are many variables involved in the delivery from orifices.

The purpose of tapered holes in cores is to help in the removal of the cores from the core box. As has been stated, it is possible that you would get different deliveries with different angles in the shape of the orifice. This is not serious from our general point of view of the delivery from these standard shapes and designs of strainer cores.

CO-CHAIRMAN ZIRZOW: Reference was made to the case of a molder just turning the core upside down, so he would increase the flow of the metal. That is a problem in supervision—to restrain the molder from doing that. They evidently do that because they are running into a misrun or into a strain. I have seen that happen many times in our own plant where the molders will switch from a 7-hole strainer core to an 8 or 9-hole strainer core because they are getting misruns. I do not think we can get any standardization if we are going to let the molders change strainer cores at will.

W. R. MOGGIDGE:<sup>1</sup> At the conclusion of the paper the author suggested that the Committee wished to obtain information that would aid in the standardization of strainer cores. Would it be of value if data on various types of core sand mixes were included?

MR. CAMPBELL: The matter of composition of strainer cores was not part of this paper. The real purpose was to bring to the attention of foundrymen an attempt to standardize on a few designs of strainer cores. There are many branches of this subject of strainer cores involving compositions, method of preparation and use of them, etc., that runs into a long story. I think it would not be well to take the time right now to go into the matter of composition of the different mixtures used.

In the investigation of strainer cores used in industry, we were surprised at the number of designs of strainer cores. We believe that is unnecessary. The opening is simply an orifice to control the rate of flow of the metal. A round orifice is probably as efficient as any. Also, we were quite surprised to see cores that were so similar, with very little change in the size of the openings, the total opening area, used by some of the larger companies. The differences were so small that it hardly seems practicable to use the many different designs. It would seem that we could get some standardization of sizes. We would like to have you say how many designs you think we should have in the foundry industry. Perhaps seven designs would be enough to cover this delivery range. We need the help of all the operators to aid us in standardization on the different designs.

CO-CHAIRMAN ZIRZOW: We purposely avoided going into sand mixtures which is quite a problem.

<sup>1</sup> Ford Motor Co. of Canada, Windsor, Ontario, Canada

# INGOT METAL vs. VIRGIN METAL

By

Fred L. Wolf \*

THE RELATIVE MERITS of ingot metal versus virgin metal have been a subject of long and heated discussions in brass foundries for many years. When a foundry has decided to use ingot metal in preference to virgin metal, the question arises whether or not the advantages and savings gained by the use of ingot metal are offset by an increase in furnace, foundry and test bench losses; also, whether or not there is a decrease in the mechanical and physical properties of the casting. These questions usually are asked when trouble occurs in the foundry and losses suddenly grow beyond normal expectancy. The practical foundryman, the man in the shop, then attempts to locate the source of trouble. Experience has shown that many factors may be responsible, but the one most frequently blamed is the metal, especially if a new shipment of ingot was used at that time.

During the pioneering days of copper-base ingot production, scrap was generally purchased from the scrap metal dealers. This was roughly sorted by hand, with only a crude guess as to its composition. It was then melted and cast into ingot with little or no attempt to refine or control the impurities. There is no doubt that foundrymen in those days had some justification for their pessimistic attitude toward the use of ingot metal.

## Ingot Metal Refining

During World War I and the period immediately following, virgin metals were scarce and the demand for reliable ingot metal increased. This resulted in the development of a highly specialized refining industry in which large reverberatory furnaces replaced most of the crucible and small furnace units. Since that time many firms in the industry have reached a high plane of efficiency. By the use of strict technical control, they now produce a uniform product which meets the rigid standard specifications of the American Society for Testing Materials. Non-ferrous ingot metal necessarily contains iron and nickel in limited quantities, due to their presence in the original scrap. These metals within the limits of the specifications are not injurious. Gardner and Saeger<sup>1</sup> found iron to be advantageous in cast red brass (85 Cu, 5 Sn, 5 Pb, 5 Zn) if the amount was less than 0.5 per cent. The

value of nickel is given recognition by A.S.T.M. specifications (1947) which have been revised to consider copper content to be met, if the copper plus nickel (0.75 per cent Max.) comes within the copper range.

Several years previous to World War II, a company operating a large non-ferrous foundry in the middle west, decided to make an extensive production test in order to settle, within its own organization, this controversial question regarding the advantages and disadvantages of ingot metal versus virgin metal. This firm manufactured brass valves, and used an alloy, the composition of which classified it as "Leaded Semi-Red Brass," meeting A.S.T.M. Specification B30-45T Class 5A. The foundry used for this test was modern, fully mechanized, and equipped with a sand conditioning and distribution system. It had modern melting, molding and pyrometric equipment and was operated under strict technical control.

In order to obtain a just comparison over an extended period the following plan was adopted: "The foundry was operated for a period of two months during which only virgin metal plus returned gates and sprues from the resulting castings were used. The foundry was then operated for the following two months in a similar manner with ingot metal plus the resulting gates and sprues." During this test all known factors were supervised and followed by the technical staff. All furnace charges were made up under the supervision of a metallurgist, and followed through the operations of melting, casting, cleaning, inspecting, machining, assembling and final hydraulic testing. All metal was poured under pyrometric control and accurate records were kept throughout the test.

At the time this test was conducted, there was a decided price differential in favor of ingot metal. While this varies from time to time, there has generally been some difference in favor of composition ingot.

Some factors are difficult to evaluate in dollars and cents. Considering such things as: amount of storage room used, time consumed in weighing, in charging furnaces, in breaking up white metals into small pieces and in properly mixing the molten metals, it was apparent that the use of ingot had quite an advantage over that of virgin metals.

During the test an attempt was made to judge the character of the metal. At the beginning, a casting was

\* Mansfield, Ohio.

TABLE 1—MELTING RECORD DURING 2-MON. PERIOD

Type Metal Used	Weight of Metal Melted, Lb			Total Melted
	Low-Frequency Induction Furnace	Indirect-Arc Furnace	Gas-Fired Crucible Furnace	
Virgin	383,215	403,981	56,275	843,471
Ingot	396,197	357,503	68,160	821,860
Total	779,412	761,484	124,435	1,665,331

TABLE 2—MELTING LOSS DURING 2-MON. PERIOD

Type Metal Used	Loss, Lb	Loss, Per Cent
Virgin	15,773	1.87
Ingot	15,579	1.89
Total	31,352	1.88

TABLE 3—CASTING LOSSES DURING 2-MON. PERIOD

Type of Metal	Per Cent Loss Dirty Castings	Per Cent Loss, Misruns	Per Cent Other Losses	Per Cent Loss, Total
Virgin	3.70	1.0	0.55	5.25
Ingot	2.40	0.80	0.60	3.80

TABLE 4—RESULTS OF TEST BENCH INSPECTION

Metal	Valves Tested	Per Cent			Per Cent Good
		Body Losses	Center Piece Losses	Other Losses*	
Virgin	96924	1.60	0.90	1.40	96.1
Ingot	79728	1.50	0.40	2.00	96.1
Total	176652	1.55	0.65	1.70	96.1

\*Defective machining, defective assembling.

selected from each heat. This was then sectioned, polished and given the deep acid etch. The test was unreliable, as it was found that the appearance of the etched specimen was in many cases misleading. Such factors as gating, and thickness and shape of the casting seemed to have more effect than the quality of the metal. Etched sections of valve parts usually showed results as follows:

- Stems—Always good dense metal
- Bodies—Good, with exception of center end
- Tail Pieces—Poor along parting line at gate
- Union Nuts—Good
- Very tail center pieces—Always poor
- Normal center pieces—Good

It was found in some instances that good and poor etched sections could be obtained from the same heat or from the same ladle. The above described procedure required considerable time and effort, and since it failed to furnish information of much value, it was discontinued at the end of the first month.

#### New Test Procedure

After discarding the above method a new procedure was developed. A test piece was cast by pouring the metal into an impression made in a bucket of mold-

ing sand, using a wooden cylinder 1 in. in diameter as a pattern. A test specimen was poured from each heat, and then fractured and examined visually. This test had considerable value, but could not be considered foolproof. In a few cases, heats that produced poor fractures on the specimen, produced castings that showed a good etched section and a good fracture. Using this test the percentage of good dense fractures obtained was higher on heats poured from ingot than from virgin metal.

#### Melting Furnaces

Three types of furnaces were used in melting the metal during this test. They included a battery of four low-frequency induction furnaces, a battery of three indirect-arc furnaces, and a battery of gas-fired crucible furnaces. An accurate record was kept of the weight melted in each furnace during the respective two-month periods and is shown in Table 1.

The melting loss during these same periods is shown in Table 2.

Considering the total weight of metal melted during the test, it appears that the slight difference of 0.02 per cent between the losses on the two types of metal is negligible.

An accurate record was kept of the foundry losses and all scrap castings were examined before they were returned to the metal repository. An examination of the foundry loss chart showed that for the 18 months preceding this test, the loss amounted to 3.74 per cent. During this 18-month period most of the metal melted was ingot although there was a fair amount of virgin metal used, and the heats were mixed. During the period of the test, the losses were broken down as shown in Table 3.

After the above four months of test, the foundry returned to the use of both virgin and ingot metal, and the loss figure returned to about the pre-test figure, 3.74 per cent.

In this test, the castings were followed through the cleaning, machining, assembling and final testing. After final assembly all valves were subjected to a hydraulic pressure test to weed out defective valves. The record of the test bench inspection is shown in Table 4.

Table 4 shows a small margin in favor of ingot metal in the elimination of leaker losses and indicates that in this respect no benefits are to be derived by the use of virgin metal.

#### Tensile Strength and Elongation

Failure to meet mechanical properties has frequently been attributed to the metal. During the four-month test period, tensile strength never dropped below 31,000 psi and elongation never below 28 per cent. A.S.T.M. specification for this alloy, B30-45T Class 5A, set a minimum requirement of 29,000 psi tensile and 18 per cent elongation.

The finish of an assembled article such as a brass or bronze valve has considerable effect when selling to the trade. Castings are not uniform in color or appearance as they come from the foundry and this factor is difficult to control. Some foundries overcome this difficulty by giving the casting a bright dip. The general treatment in obtaining uniform color is:



1. Sand or shot blast after removal of gates and sprues.
2. Immerse for 1 min in a solution of
  - 1 part 36° Bé nitric acid
  - 2 parts water
3. Rinse thoroughly
4. Dip quickly in a solution of the following mixture
  - (a) 2 gal—36° Bé nitric acid
  - (b) 8 gal—60° Bé sulphuric acid
  - (c) 1/2 oz hydrochloric acid
5. Rinse thoroughly and quickly dip into following solution
  - (a) 1 1/2 oz potassium permanganate
  - (b) 16 oz dilute sulphuric acid (1:50 Water)
  - (c) 10 gal water
6. Rinse thoroughly and dry on a steam table.

Using the above procedure, it was shown that castings made from ingot metal showed a richer and deeper color than those made from virgin metal.

### Conclusions

This thorough test on the production of valve castings, conducted on a strictly impartial basis, proved conclusively that the use of refined copper-base ingot metal of A.S.T.M. specification B30-45T Class 5A, offered many advantages over the use of virgin metal:

1. Metal cost of ingot was considerably less than that of virgin metal at the time this test was conducted.
2. The use of ingot showed a saving in storage room, in time consumed in making up charges, in preparation of metals for weighing and in alloying time.
3. Regardless of the furnace used in melting, fractures of test plugs made from ingot metal were superior to those made from virgin metal.
4. Melting loss was about the same on ingot as on virgin metal.
5. Foundry and test bench losses were lower on castings from ingot than on those from virgin metals.
6. Ingot easily met A.S.T.M. specification requirements for tensile strength and elongation.
7. After chemical treatment, castings from ingot metal had a slight advantage over virgin metal as to appearance.
8. There was no evidence that the advantages obtained by using ingot metal in preference to virgin metal were offset by any disadvantages.

### References

1. H. B. Gardner and C. M. Saeger, Jr., "The Effect of Sulfur and Iron on the Physical Properties of Cast Red Brass" (85 Cu, 5 Sn, 5 Zn, 5 Pb), *Proceedings, A.S.T.M.*, vol. 33, part II, pp. 448-61 (1933).

### DISCUSSION

Chairman: H. M. ST. JOHN, Crane Co., Chicago

Co-Chairman: J. J. CURRAN, Walworth, Greensburg, Pa.

WM. ROMANOFF:<sup>1</sup> I am in accord with Mr. Wolf's findings that ingot metal can be and is being used in preference to virgin metal. Whether the results at all times are as good as Mr. Wolf's is questionable. I am not going to tell you that by the use of ingot metal you will obtain higher physical properties or less foundry losses than by the use of virgin metals. I have never seen castings made of virgin metal that could not be duplicated with ingot, or vice versa. All things being equal, there are some advantages in the use of ingot metal as against virgin metal.

<sup>1</sup> H. Kramer & Co., Chicago

1. There is the item of cost, particularly today. There are some alloys that can be bought for as much as four cents per pound below the cost of making them out of virgin metal.

2. In weighing and melting a heat of alloyed ingot, as compared to virgin metal, the former is more foolproof. As an illustration, if we were making an 85-5-5-5 casting from virgin metal, four separate weighings would be necessary, the copper, tin, lead and zinc respectively. A mistake in either weighing would naturally throw off the analysis. This would not be true in the use of ingot. If 110 lb were weighed instead of 100 lb, or 90 lb instead of 100 lb, there would merely be more or less metal, but the analysis would be the same.

3. An alloyed ingot will melt at a lower temperature than the pure copper, which must be melted before adding the zinc, lead, and tin, or in whichever order you intend adding these latter elements, with an accompanying saving in melting time and fuel cost.

4. Another advantage which may only be temporary, is the fact that at the present time most of the virgin metals such as copper, tin, lead and zinc are difficult to obtain, while alloyed ingots do not offer this problem.

H. J. ROAST:<sup>2</sup> As a metal user I agree with the author's statements in the first place and with Mr. Romanoff's statements in the second. I have had several cases where the only thing that I had to do to solve the major problems of the foundry was to substitute ingot for virgin metal. The question of volume comes in. Frequently, people that use virgin metal do not mean only ingot copper, for example. They may take other forms of so-called virgin metal but these take a greater volume, and when you have greater volume, particularly in crucible melting you increase the hazard of oxidation or of hydrogen absorption. In my opinion there are many things to be said in favor of ingot as compared with the pure metal. There are perhaps some exceptions where you have an extremely fine question at issue. These then become almost self-evident.

During the war, many of us in the States and in Canada had to make non-magnetic bronze. Some of us felt we could make it easily because when we applied a magnet to a piece of this bronze and it did not lift or adhere to it, we said it was non-magnetic. We were soon to find that the Navy and the Army, but particularly the Navy, did not agree with that. You had to have it so that the finest possible turning, with the strongest possible magnet, would not make any movement in the said turning.

C. S. COLE:<sup>3</sup> On that last point—the non-magnetic bronze—I would like to call your attention to a paper published in the *ASTM Bulletin*, probably in August (1948). This paper was written by Prof. Butts of Lehigh University who did the experimental work on magnetic testing of copper-base alloys for Frankford Arsenal during the war. It is a very interesting paper.

There is a very interesting sidelight in Prof. Butts' work which has not yet been emphasized. By a solution heat treatment it appears possible to bring alloys with the standard tolerance for iron into the non-magnetic class so that they can be used and you do not have to resort to metal of special high purity. This applies to several standard alloys, and the three casting alloys that look most favorable in Prof. Butts' work are the 85-5-5-5 (Alloy 4A), the 81-3-7-9 (Alloy 5A) and commercial yellow brass (Alloy 6B). These three alloys can be heat-treated so that they can be used for applications requiring non-magnetic properties when made to the standard specification limits and not to any special limits, with very low iron.

MR. ROMANOFF: I want to comment on an experience we had with magnetism in a nickel-silver alloy containing iron. We found the magnetic property rather fleeting. Some castings made from the same composition showed magnetic properties, while others did not. In fact, proper heat treatment could eliminate the magnetism.

CHAIRMAN ST. JOHN: I would like to add one point which would be considered a disadvantage in the use of ingot. I am in favor of using ingot and do use it almost exclusively, but there is one thing that needs to be watched for. You will find sometimes a difference in fluidity between one lot of ingot and another which cannot be predicted from the composition or from the usual analysis of impurities. If it is good ingot, the difference will not be great, and only in cases where fluidity is critical need it be considered. Occasionally, you will find it necessary to change the optimum pouring temperature from one

<sup>2</sup> Canadian Bronze Co., Ltd., Montreal, Que., Canada

<sup>3</sup> American Society for Testing Materials, Philadelphia

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lot of ingot to another, where fluidity and pouring temperature are critical. It is true that in cases of that kind, it simplifies the problem some to use a small percentage of virgin metal with the ingot and the foundry returns.

G. P. HALLIWELL:<sup>1</sup> A few years ago we installed a spectrograph in our laboratories for closer control of our smelting and refining processes. As there were no standards available for the establishment of working curves, it was necessary that we make our own. To do this we used virgin metals and added the required amounts of minor metals necessary to give us the required alloys. Chemical analysis of several samples taken from each heat failed to show agreement. It was found necessary to remelt these heats at least three times before consistent chemical analyses could be obtained. This might be explained on the basis that repeated melting produced an increased number of nuclei from which solidification could start simultaneously and thereby prevent segregation.

CHAIRMAN ST. JOHN: I think very few of us would like to make castings from all virgin metal.

S. W. CHAPPELL:<sup>2</sup> I would rather make high-pressure application castings from all virgin metal on the second melting than from ingot.

CHAIRMAN ST. JOHN: That is, you mean to start with all virgin and melt it the second time?

Mr. CHAPPELL: We have found that high-pressure castings, operating under 300 to 5,000 lb pressure, cannot be made satisfactorily from ingot metal.

Mr. ROMANOFF: What is your reason?

Mr. CHAPPELL: I have no reason.

Mr. ROMANOFF: Your statement is difficult to answer without more detailed information as to your foundry practice and the reasons why the castings did not withstand pressure when made from ingot, but were satisfactory when made from virgin metal. It is a known fact among foundrymen and metallurgists that when the cause of the defect is known, the remedy is usually obvious. You give no cause. It might have been improper gating, wrong pouring temperatures, furnace atmosphere or any one or more of a number of reasons. Of course, you might have used poor quality ingot. As you know, there is ingot and ingot. Any good ingot from a reputable manufacturer should have been satisfactory. During the war, there were all types of castings made to withstand pressures, probably as high as those which you mentioned. Virgin metals, even if desired, could not be obtained by most foundries, so they made the castings out of alloyed ingot successfully.

CHAIRMAN ST. JOHN: In view of what I stated before, I think it is only fair to add that during the war and up to the present time we have made many castings required to stand up to 2,000 lb or more pressure and have had no difficulty in getting them with ingot metal using 88-10-2 alloy.

Mr. CHAPPELL: Where physicals were concerned, ingot metal gave us as good or better results than virgin metal did.

Mr. ROMANOFF: This seems like an inconsistency. If the physical properties of ingot heats were as good if not better than the virgin metal heats, why did not the castings made from the ingot heats withstand pressure? Apparently it was not a metal defect, but some other foundry defect.

Mr. CHAPPELL: They just would not hold water under pressure. The same gating method was used and other conditions likewise were the same.

Mr. ROMANOFF: It could hardly have been all the same. There must have been something that was done differently, which can hardly be discussed without more detailed foundry information.

Mr. CHAPPELL: That was the result. We ultimately had to go back to using virgin metal for all high-pressure valves, manifolds, etc.

D. C. CAUDRON:<sup>3</sup> We made a number of these castings from Composition "G" and the majority of them seemed to be satisfactory under 5,000 lb test pressure. They were tested at Mare Island Navy Yard under pressure for 1 or 2 hr. One night they left the pressure on one casting that tested O. K. and by the next morning it developed a weeping that caused its rejection. From then on I do not think we ever made a casting that was pressure-tight, according to Navy specifications, because if it were left under pressure for an extended period of time, it would show weeping somewhere.

Mr. ROMANOFF: Were the fractures the same?

Mr. CHAPPELL: The fractures were perfectly satisfactory. There were no visible signs of any difference.

Mr. ROMANOFF: If you were able to get over 50,000 psi tensile without difficulty and show a perfect fracture, why should 5,000 lb pressure cause a leakage? Why should it even cause weeping if you have a good fracture? If you had a shrinkage condition, I could understand it, but if you had a solid fracture I cannot imagine 5,000 lb pressure passing through an alloy that would have a tensile strength of a minimum of 50,000 psi. I am not doubting or disputing your word.

Mr. CHAPPELL: We have not been able to understand it yet, but they did.

Mr. ROMANOFF: You probably are blaming the wrong thing.

Mr. CHAPPELL: I do not think so, because we switched overnight from virgin metal to ingot, using the same methods, the same mold, the same gatings, the same risers, and the same melting conditions. They went from fairly good to very bad. We do not save all the castings made from virgin metal, but our losses were so great with ingot that we had to switch back to virgin metal entirely.

B. A. MILLER:<sup>4</sup> What will smelters do when copper-base scrap metal becomes non-existent? How would they make satisfactory ingots?

Mr. ROMANOFF: We would make it out of virgin metal. We made it during the war. We used a big percentage of virgin metal during the war to supplement our scrap, because we could not get all the scrap that we required. The demand was too great. That is one reason I would not try to leave the impression that an ingot metal will give you better results.

W. B. SCOTT:<sup>5</sup> We had a condition in our plant where the use of virgin metal was the answer to many problems on special aluminum bronzes. At that time the volume of those bronzes was not sufficient to interest any ingot producer. When the volume reached the point where they could devote time and attention to it, they have been able to duplicate from scrap everything that we have been able to do from virgin metal. I do want to say that in those alloys, analysis did not show the difference. The difference was a matter of hot shortness. It became the standard practice to purchase metal ingot and the only way we could use that ingot was to remelt it, not once, but several times, to make it usable. It was simpler in our own shop to make our own ingot and start from that basis. It is a matter of understanding what the problems are that are involved.

Mr. ROAST: This discussion seems to be largely on the value of merits of virgin vs. ingot metal on the basis of pressure castings. I understand that Mr. Chappell was using pressure of 300 to 5,000 lb.

We have a pressure test which we subject metal to. We have a cylinder about  $3\frac{1}{2}$  in. in diam and about 7 in. long, with one end open and one end closed. It is 1 in. in cross section. We machine off one-third of the OD, one-third of the ID. We have tested it frequently to 6,000 psi for the remaining inside metal without any leaks, and that has always been made from ingot metal. I think that should be borne in mind. Incidentally, however, I should never choose 88-10-2 for a pressure casting but rather a composition ingot 75 Cu, 15 Pb, 7 Sn, 3 Zn. Even the British Navy finally allowed us to substitute the leaded bronze for the gun metal for 6,000-lb pressure work. There is one limiting factor. You do not want to heat it over 400 F or you might have trouble, although actually at 500 F you frequently do not. But it will from ingot make satisfactory castings at 6,000 lb pressure.

Mr. CHAPPELL: The snorkel exhaust valve, a jacketed valve, is a good example of the type casting I am talking about. I think it becomes heated over 400 to 500 F. That is the main exhaust outlet from the engine and I think in service it exceeds 400 to 500 F in temperature.

H. E. YOUNGER:<sup>6</sup> I want to tell of an experience we had during the war with CO<sub>2</sub> fire extinguisher valve castings. We made 175,000 of these castings from 85-5-5-5 alloy. They were tested at 2800 lb pressure for 30 days before the Air Corps would accept the extinguisher.

<sup>1</sup> Electric Boat Co., Groton, Conn.

<sup>2</sup> Pacific Brass Foundry, San Francisco

<sup>4</sup> Baldwin Locomotive Works, Philadelphia

<sup>5</sup> Aurora Metal Co., Aurora, Ill.

<sup>6</sup> Chicago Faucet Co., Chicago

# PRODUCTION HARDNESS TESTING IN A MALLEABLE SHOP

By

C. Schneider and L. Ulsenheimer\*

## ABSTRACT

The authors discuss two methods for measuring Brinell hardness of pearlitic malleable castings. The first method is the method of applying the Brinell load of 3000 kg through a 10 mm ball manually or automatically. The diameter of the impression made in the casting is measured by means of a Brinell microscope and this value is converted to Brinell hardness number (Bhn). In the second method the hardness is determined by means of a direct reading machine first by applying a 1500-kg pre-load, resetting the dial to zero and then applying the full 3000-kg load. The dial on the machine then indicates the hardness of the casting. Constant checking of the direct reading machine with frequent calibration is necessary. The authors point out that in this, as in many other operations, there are opportunities for error.

TRADITIONALLY, the foundry has checked the hardness of malleable castings by means of a test lug. This method has been and still is considered the most acceptable way for determining this characteristic. Hardness as determined by the Brinell method is not considered a too reliable measure of machinability in the range of malleable iron.

With the introduction of pearlitic malleable the use of the Brinell method of hardness testing became

necessary. Entering this field early, the Cleveland Works of the National Malleable and Steel Castings Company has met many problems in the production handling of hardness testing by this method. As the tonnage of this material increased and the flow of the product through the shop accelerated, improved techniques were required. These techniques were developed as necessity and experience dictated. The primary requirement was, of course, to keep pace with the volume demand. Today's operations represent the experience acquired in keeping step.

Since the test lug is a necessary part of equipment, we have adopted definite sizes and tried to plan the position on the casting as well as the number used.

ASTM standards specify that the lug should be of a size that is proportional to the thickness of the casting but should not exceed  $\frac{5}{8}$  in. x  $\frac{3}{4}$  in. in cross-section. For our use we have decided on four sizes. These are rectangular and are dimensioned as indicated in Fig. 1. The usual taper is  $\frac{1}{32}$  in. per in. on each side and the height equals the longest diameter of the base.

When the lug is being placed, consideration is given

Type	A	B	C
1	$\frac{9}{16}$	$\frac{5}{16}$	$\frac{11}{16}$
2	$\frac{5}{8}$	$\frac{3}{8}$	$\frac{3}{4}$
3	$\frac{11}{16}$	$\frac{1}{4}$	$\frac{13}{16}$
4	$\frac{3}{4}$	$\frac{1}{2}$	$\frac{7}{8}$

Fig. 1—Showing sizes of test lugs used in authors' plant for malleable castings.



Fig. 2—Casting showing test lug.

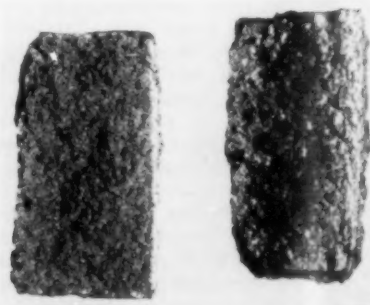


Fig. 3—Fractures of test lugs.

\* National Malleable and Steel Castings Co., Cleveland, Ohio.



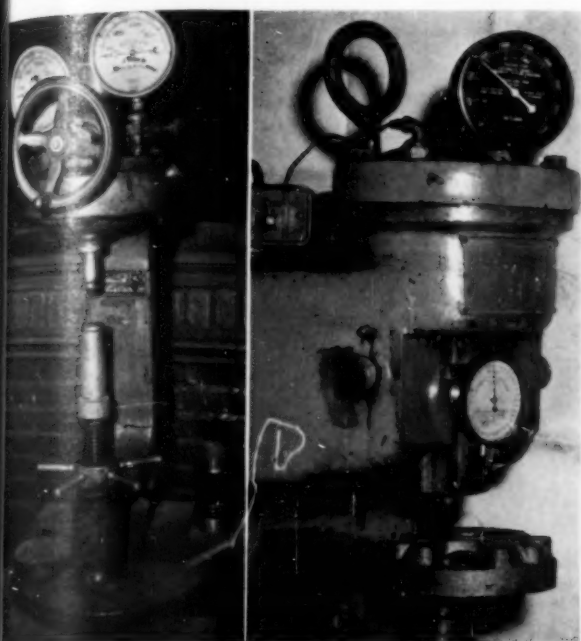


Fig. 4 (left)—Motor driven Brinell hardness testing machine. Fig. 5 (right)—Direct reading hardness testing machine with pre-load applied to casting and indicator dial reset to zero reading.

load and (2) determination by the direct reading machine.

The evolution of a method suitable for the handling of a large number of castings per hour has included several steps. Initially, all hardness determinations were taken by preparing a surface, making the impression, and measuring the diameter. One, two or three operators were required depending upon the desired number of castings per hour. This is the operation we use today on a small part of our production.

#### Process Simplified

There have been changes over the years which have simplified this process. They have come largely in the manner of applying the load since the method used to prepare the surface is essentially the same. On the earlier machines the load was applied by a hand operated pump or by the application of weights. Motor driven units to provide the load have replaced the manually applied load but the operation is much the same (Fig. 4).

In the attempt to simplify the process most effort has been directed toward eliminating the polishing of a surface and the glass reading. In addition to the time required for the measuring of the impression by a microscopic glass there was also the not inconsiderable element of operator fatigue.

It was reasoned that since the depth of penetration

to the possibility of breaking it off by rough handling when the casting is packed for annealing. We have also found it advisable to anticipate possible interference with other operations such as grinding or chipping. If the lug is on a surface that must be flat and smooth, removal may be a problem. With these points in mind the lug is placed at or near a heavy section and, if possible, in the drag.

The next point we consider is the number of lugs to be used. On small castings weighing less than 5 lb. the usual practice is to use no lug and a casting is broken for examination. On castings weighing up to 20 lb. either one or two lugs may be used but on some smaller castings two lugs have proven mandatory. If the casting weighs more than 20 lb., it is our practice always to use two test lugs (Fig. 2). When we have used two lugs and the casting has been rejected by the first, the remaining one is not removed but is utilized to check after the re-anneal.

In this type of hardness testing the visible evidence given by the test lug fracture provides the basis for acceptance or rejection. The fracture in Fig. 3 offers proof of the metal quality as well as of the malleability. To secure a good fracture, the lug is marked well on the pattern so that it can be broken by hammering in one direction. One indication of the condition of the metal is found in the ease with which the lug is separated from the casting.

When handling pearlitic malleable, the problem becomes one of segregating those castings which are outside the specified Brinell hardness number range. The test lug in this instance does not serve the purpose and is not used. The authors' plant is using two ways to check this hardness: (1) microscope reading of the impression of a 10 mm ball under a 3000 kg

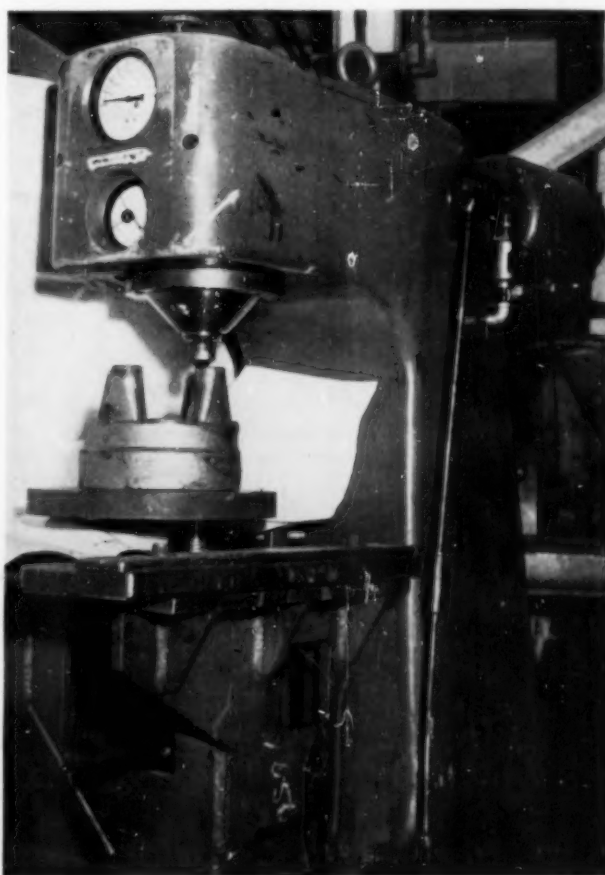


Fig. 6—Casting being hardness tested in direct reading hardness testing machine.

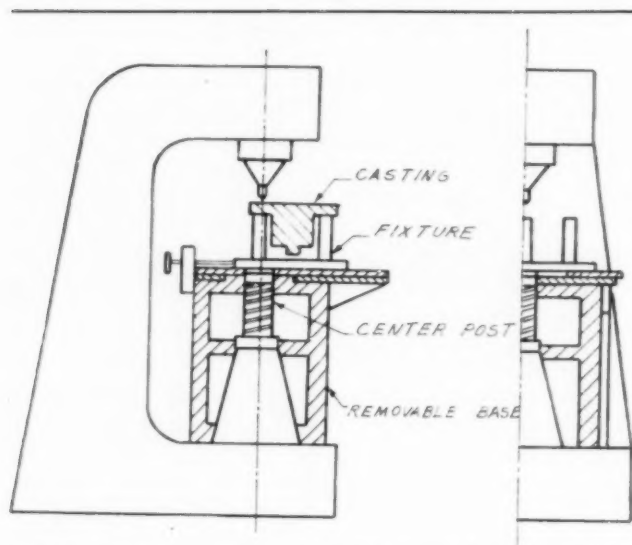


Fig. 7—Showing removable base around center post for resting fixture.

of the 10 mm ball was a function of the diameter, an accurate measure of this penetration should give an indication of the hardness of the casting. One of the problems which required an answer was the question of the point at which the start of the penetration should be measured. We needed to have the ball seated firmly on the piece to be tested yet not penetrate it to any significant depth. By experimentation and trial we found that if the ball were seated with a 1500 kg load the penetration would seat the casting satisfactorily.

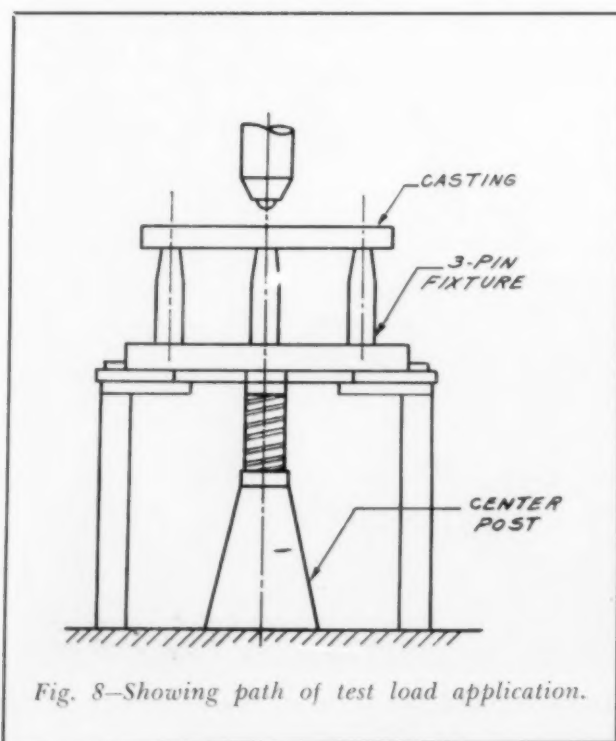


Fig. 8—Showing path of test load application.

The first arrangements for this type of checking had a dial indicator mounted on the machine so that the movement of the head containing the 10 mm ball could be observed. After the 1500 kg pre-load was applied the indicator was returned to zero (Fig. 5). Then the full 3000 kg load was applied and the indicator reading noted. On the basis of this reading the casting was found to be hard, soft or within the required range.

This machine could be used for either glass reading or the method of direct reading. To set up the limits on the dial indicator comparison was established. A casting was prepared and read in the usual manner and the Bhn noted. Immediately adjacent to the spot checked the depth of penetration was noted on the dial indicator. By comparison between the Bhn as determined by glass and penetration as measured by dial, limits were established for acceptable hardness. These limits, established by comparison, were approximate and no particular dial reading could be said to equal a definite Bhn.

The direct reading machines now being used automatically apply the pre-load and measure the penetration after the application of this load. The operator's hands are free to handle the casting since the load is applied by means of a foot pedal.

This change still required the operator to seat the casting on the center post and adjust its height. To secure an accurate measure, correct alignment of the ball, casting and center post was necessary. Now a means of placing the piece quickly and accurately on the center post needed to be devised. Since the parts to be tested varied in design and thickness, a fixture was required which could be readily changed, could be easily moved into the correct position, and would give the proper suspension to the casting.

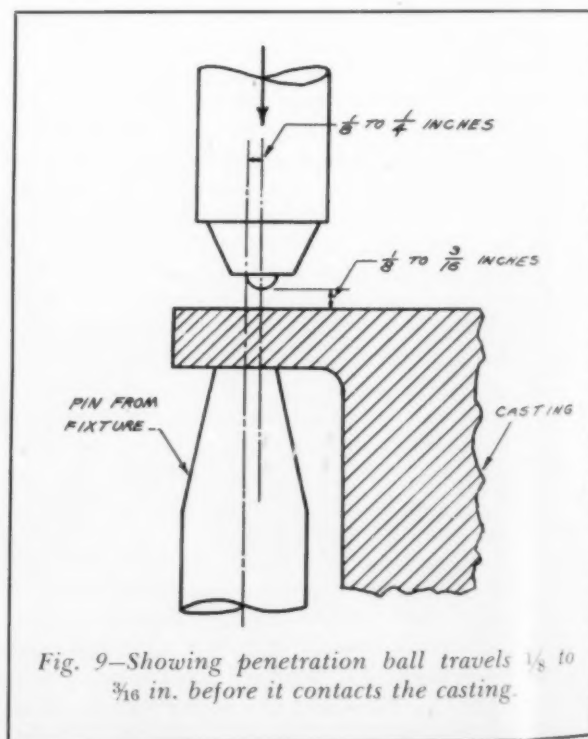


Fig. 9—Showing penetration ball travels  $\frac{1}{8}$  to  $\frac{3}{16}$  in. before it contacts the casting.

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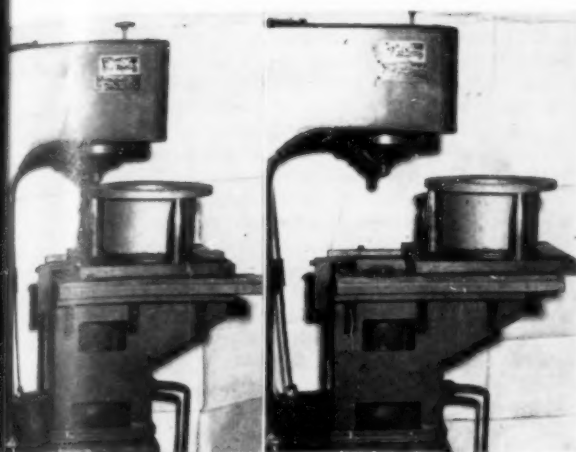


Fig. 10 (left)—Test fixture rests on removable base and is slid out from under the penetration ball. Fig. 11 (right)—Showing adjustable stop to control location of fixture under the ball.

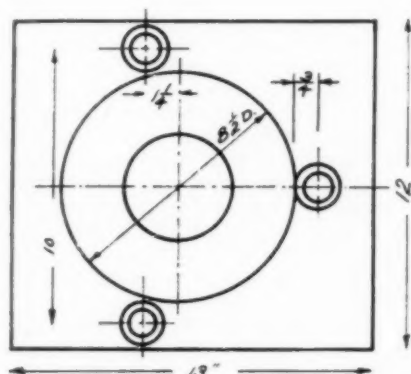


Fig. 12—Hardness testing fixture.

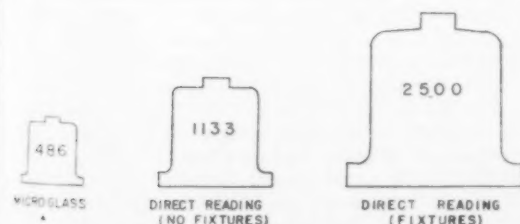
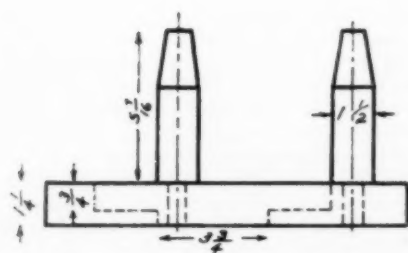


Fig. 13—Average number of castings hardness tested per day at authors' plant.

At first the authors mounted the fixture on the center post (Fig. 6). This showed that a broader base was desirable. Consequently, a removable base was built around the center post with an opening through which the post could rise and another through which it could be adjusted (Fig. 7). This base provides the surface on which the fixture rests. We have illustrated the path for the transmission of the load (Fig. 8).

The fixture which rests on the base was designed to permit rapid loading of the machine. In designing

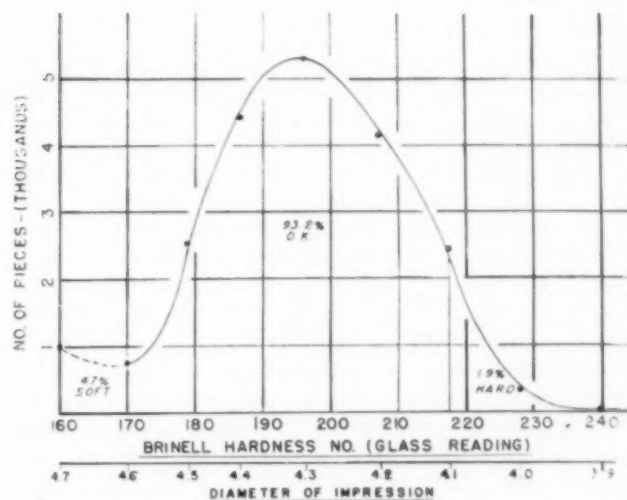


Fig. 14—Plot of 21,141 hardness tests showing distribution of percentages of rejected and accepted castings.

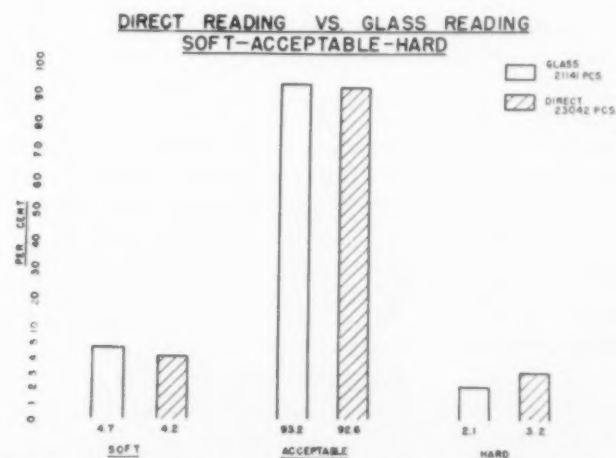


Fig. 15—Showing percentages of accepted and rejected castings by the two hardness testing methods.

the fixture we found that the best results were obtained when two dimensions were maintained. First, the center line of the pin on which the casting rests should be  $\frac{1}{8}$  or  $\frac{1}{4}$  in. off the center line of the transmitted load. This prevents the piece from tilting thereby introducing an error. The second dimension which we try to maintain is the distance the ball must move before it contacts the work. Be results have been obtained when the ball travels from  $\frac{1}{8}$  to  $\frac{3}{16}$  in. These are shown in Fig. 9.



A 1 or 1¼-in. gray iron plate is the most satisfactory for the fixture base. On this base, hardened pins are mounted to support the casting and provide the carriage. This fixture or carriage rests on the removable base and is slid out from under the ball (Fig. 10). The casting is inserted; the carriage is slid back into place, and the reading taken. At the back of the base we have placed an adjustable stop (Fig. 11) to control the location of the fixture under the ball and provide the first required dimension mentioned above. If the casting is smaller, it is frequently possible to leave the carriage in position and place the piece directly on it.

In the illustration of the fixture (Fig. 12), an opening in the bottom of the plate may be observed. This provides clearance for the trunnion in this instance. Such arrangements are sometimes required by the overall dimensions of the casting. The use of these fixtures has made possible an increased output. We have undertaken to illustrate this in Fig. 13 which shows the average number of castings handled per day.

The question will naturally arise, "How accurate is this procedure?" Let it again be stated that the use of the direct reading machine to determine hardness is only as accurate as the comparison set up between a casting of known Bhn and the indicated depth penetration on the dial. Uniformity of suspension and load application are essential to an accurate comparison. We have, however, taken figures from our regular production which illustrate the process as it is operating. The castings in the lot were made in the same period and went through the same treatments. A part of the production is normally read by glass because of short runs and other limiting factors. A distribution according to diameters measured was plotted on 21,141 pieces and is illustrated in Fig. 14. Figure 15 shows the overall picture of the percentages rejected and accepted by the two methods.

The authors realize that the desirability of testing hardness by the direct reading machine may be debatable. Constant checking of the machine with frequent calibration is necessary. The operators must be impressed with the importance of watching the operation and checking the calibration of their machine. In this as in many other operations there are opportunities for error.

### DISCUSSION

*Chairman:* C. F. JOSEPH, Central Foundry Div., General Motors, Corp., Saginaw, Mich.

*Co-Chairman:* ERIC WELANDER, Union Malleable Iron Works, Deere & Co., E. Moline, Ill.

*Co-Chairman WELANDER:* It interested me to note that the authors have eliminated preparation of the casting surface in practically all their castings before the hardness tests have been made. In some shops that preparation of the surface by spot grinding consumes much time. According to the paper the authors have been able to eliminate that almost entirely except for the checking period prior to setting up their limits.

*CHAIRMAN JOSEPH:* I should like to discuss Brinelling of pearlitic malleable castings. We spot grind all castings that are hardness tested on the Gogan or Detroit hardness testing machines which are the same machines used by the authors. We find it absolutely necessary to get a smooth spot ground on the casting to be sure that the readings are correct. I am surprised that the authors have found that they do not have to spot grind their castings. We have gone through this procedure with thousands of castings and definitely have checked our results. We find that we must spot grind every casting. Prac-

tically every casting which we produce for pearlitic malleable is Brinelled. We produce rocker arms weighing about one third pound each. We check about 1500 castings per hour with either the Detroit or the Gogan machines.

Recently we had one of our electrical men work on a magnetic retentivity test. He was able to perfect this electrical test without spot grinding, he is able to check with this machine 3600 rocker arms per hour. This test is not in actual production but it will be shortly after it gets customer approval.

We have been making castings for over 20 years without test lugs and we feel that the use of test lugs is not necessary. It requires time to break off the lug and 99 out of 100 test lugs are not examined and the fracture interpreted. In many cases the test lugs have to be ground after the lug is broken off. We believe that it is much better to put the extra effort on improved control, if that is possible.

One good control of the iron is the shearing of the gate. It will give you some indication of whether the casting is machinable. Where the casting is ground or the gates are ground that is not possible, but the use of test lugs can be overdone.

*E. M. STRICK:*<sup>1</sup> With reference to not grinding castings before hardness testing them on the aforementioned machines, if the standard was set for castings not ground would Mr. Joseph find variations that would throw the Brinell hardness off?

*CHAIRMAN JOSEPH:* We of course would like to Brinell castings without spot grinding them because it requires extra work but we have definitely found that we were not able to hold the castings to close enough control in case they go outside the range. The Brinell impression you obtain on a rough casting is influenced greatly by the possibility of the casting having a decarburized rim. If you do not grind the surface of the casting off, how can you expect to get an accurate reading on the casting? All pearlitic malleable has a slight skin, and in order to get a true Brinell hardness of the casting, we have always felt that it is necessary to take off possibly 0.015 in. In fact, a lot of our castings have a raised pad on the casting that is ground flush with the casting. That makes it possible to grind the casting without getting below the machined surface. We do not do that on all of our castings, but a lot of them where it is necessary to have accurate machining, there is a raised Brinell pad on the casting. This was not the case with the rocker arm because the rocker arm had enough metal which was machined off on the boss end.

*MR. STRICK:* There is another question I would like to ask of Mr. Joseph. On the shearing of pearlitic malleable castings, how is it possible to tell whether the casting is soft or hard when you are shearing it?

*CHAIRMAN JOSEPH:* I did not mean that we could tell from the shearing of the pearlitic malleable iron. I meant this for regular malleable iron. We use that as a gage very often. If there is any trouble with the iron, the operator could tell from the shearing of the malleable iron casting whether it was hard. We do not shear any pearlitic malleable castings. We did shear one job during the war, a machine gun part that was machined at the point where it was sheared. If it did break in a little the machining operation would take care of it.

*MR. STRICK:* That is exactly what I found. In shearing pearlitic iron, the gates have a tendency to tear out and they are tough on shearing tools so we had to confine it entirely to malleable iron.

*CHAIRMAN JOSEPH:* I do think it is possible to shear pearlitic malleable castings if you want to spend some time on the proper shearing knives. In fact, we have thought about that many times because it is cheaper to shear than to grind. There is no reason, if the gate is not too heavy, why you cannot shear a casting by the proper design of the shear knives.

*MILTON TILLEY:*<sup>2</sup> We found that this preload of 1500 lb which you do not use for measuring the hardness takes care of that pearlitic or decarburized edge. So it does not have the effect you might think it does on the final load.

*L. J. WISE:*<sup>3</sup> The author stated that the Brinell hardness test is not a good indicator of machinability of normal malleable iron. Do you mean, for example, that you could get a diameter of 5 mm or better with a 3000 kg load and have it not easily machinable? Conversely, could you get a diameter under 4.7 mm and have it readily machinable?

<sup>1</sup> Erie Malleable Iron Co., Erie, Pa.

<sup>2</sup> National Malleable & Steel Castings Co., Cleveland

<sup>3</sup> Chicago Malleable Castings Co., Chicago

Mr. SCHNEIDER: Machinability is often influenced by factors not necessarily indicated by Brinell hardness. Variations in alloy content may affect it, the pearlite may vary and decarburization may also affect the machinability. It is not likely, however, that hardness under 4.7 will be machinable as malleable iron.

L. N. SCHUMAN:<sup>2</sup> Our experience has been that there are many factors other than hardness that influence tool life or machinability. A decarburized edge, with the tool not ground at the proper angle and rake, will ruin your cutting edge rapidly. You could change the angle of the tool and get somewhat better results. Naturally, a little burned-in sand will cause trouble.

A thin pearlitic rim with a thin light cut would give far more trouble in machining than the same pearlitic rim if you were taking 3/32 in. off. That is why I think, generally speaking, that Brinell hardness is not a measure of machinability.

CHAIRMAN JOSEPH: I would like to have an expression of opinion on putting Brinell hardness in malleable iron specifications. We have some malleable iron customer specifications that definitely call for Brinell hardness in the specification. We do not like it, not that we cannot meet it, but there is always a question that if the customer should have trouble with the castings, he will insist that you Brinell these castings 100 per cent.

HYMAN BORNSTEIN:<sup>4</sup> A Brinell hardness range may be a desirable thing in a specification. We are thinking about it in gray iron castings and I believe that if a malleable casting is annealed properly, it will fall within a relatively close range of Brinell hardness.

As previously mentioned, how are you going to take that Brinell test? If you grind off the surface so as to get a measure of what is back of the surface are you getting a Brinell reading on the whole casting? Obviously not, and it may be that the surface is giving trouble in machining.

CHAIRMAN JOSEPH: Does your company put the Brinell hardness maximum in their specifications?

MR. BORNSTEIN: No, we do not. However, if there is a complaint, one of the first things we do is to make a Brinell test on the casting.

J. H. LANSING:<sup>5</sup> I wonder if we are not talking about two different things, one, the necessity of Brinell readings on standard malleable castings and the other, the necessity of Brinell readings on pearlitic malleable. There certainly is a greater necessity in the latter case than there is in the former. In fact, there is a question if there is any necessity in the case of standard malleable, although the taking of Brinell readings is often desirable in the case of pearlitic malleable. So I wonder if we should not clearly distinguish between which we are talking about.

CHAIRMAN JOSEPH: We were talking in the latter case about standard malleable iron. It is true in pearlitic malleable. I am positive you must have a Brinell range, but I referred to regular malleable iron. Quite often the question comes up, "Why do you not have a maximum Brinell hardness that the malleable iron casting must pass?" In most cases I think 133 Brinell is about as high as a lot of people produce malleable. A lower carbon iron would be somewhat harder, about 143.

MR. LANSING: That is my point, that is, no Brinell specification is necessary for standard malleable iron. It was stated that the Handbook says "that the Brinell reading is not necessarily a good indication or a necessary indication of machinability". This statement refers to a standard malleable iron and not pearlitic malleable iron.

W. D. McMILLAN:<sup>6</sup> I am glad Mr. Lansing spoke before I did regarding the standard malleable iron because I do not believe that a Brinell hardness should be part of the specifications for malleable iron. As a salvage proposition it offers a very practical method of sorting castings, some of which may be under-annealed. Normally, we consider 140 Brinell and under as fully annealed. If the Brinell is higher, a microscopic examination is made to determine if the higher hardness is due to cementite or pearlite or to both. If the Brinell is about 170 and the structure is pearlitic, usually the castings are acceptable, providing they do not require severe straightening.

If the higher hardness is due to cementite, which would definitely affect the machining, the castings are re-annealed or disposed of by whatever means is necessary. We have a specification of 140 Brinell which is used as indicated above as a basis for using or discarding certain castings.

With regard to test lugs, they are of value provided they are large enough to represent the cross section of the casting. If they are too large, they are difficult to remove after annealing, and there is a tendency to knock them off in the hard iron casting. A small test lug may be fully annealed, while the heavier sections of the castings may show some cementite.

With regard to the direct reading Brinell, we built a hydraulic machine which we used as a means of sorting castings. We used a 100-lb minor load and a 300-lb major load and the reading is made on a dial indicator in much the same manner as on the Rockwell machine. Limits are established by correlating with a standard Brinell impression.

With regard to the oscilloscope, we have used this type of equipment to determine relative degree of anneal and relative hardness which may be correlated with the magnetic permeability. We have used this equipment in sorting small castings. It is necessary to set up a pair of coils for each particular casting to be inspected. The castings must be inserted accurately and in the same position in order to obtain accurate results. Once the set-up has been made, this provides a rapid means of establishing a relative hardness of castings.

MR. TILLEY: It is well to remember that standard malleable iron normally runs from 80 to 130 Brinell. You can have a normal malleable with a Brinell of 130, containing high phosphorous and high silicon. It is also possible to produce a 90 Brinell malleable so that it would have enough pearlite in it to produce a hardness of 130 Brinell and that would affect machinability. So you have a problem of taking care of the analysis of low carbon, high silicon, and high phosphorous. We have found that 50 points of phosphorous will raise the Brinell hardness number 20 points.

MR. McMILLAN: Is the presence of pearlite of sufficient importance to affect the machinability to any extent? You certainly machine pearlitic iron without much difficulty. I do not think a trace of pearlite, up to 0.3 per cent, as estimated from the microstructure, would make enough difference, as a matter of salvaging.

W. K. BOCK:<sup>2</sup> I do not like the idea of inserting in a specification a lot of numbers unless we have a definite reason for it. Malleable castings are purchased on properties such as strength, machinability, shock resistance, and in some instances on fatigue resistance. If you tried to correlate any one of these properties with the possible exception of strength with the Brinell number, that correlation would probably not be very good. The Brinell number then would be only a rough check on the property required in that casting. In that case, putting hardness requirements in that specification would probably either get you castings that you were not sure had the required properties, or else the manufacturer would be caused to throw out a lot of castings that were acceptable simply because the Brinell number which is not correlated too well anyway, showed that the casting was outside the limits.

It seems to me that Brinell specification is going to be a hindrance to the malleable foundry as well as to the castings purchaser who wants his castings in a hurry. The buyer will have to waive that hardness specification from time to time to get castings that he can use. The tendency is to narrow down the hardness range as a safety factor. In so doing you simply throw out many perfectly good castings. I am not sure that Brinell hardness is a measure of properties we want to get at, anyway.

MR. McMILLAN: Our specifications are divided into two divisions. One is the specification giving mechanical properties for the engineer and the other is more or less a guide to the foundry. It is a control practice rather than a specification for the engineer.

MR. BOCK: The idea of using Brinell hardness simply as a means of sorting castings is one thing, but to write it into a specification and sign a contract on that basis is an entirely different matter.

<sup>2</sup> John Deere & Co., Moline, Ill.

<sup>3</sup> Malleable Founders' Society, Cleveland

<sup>6</sup> McCormick Works, International Harvester Co., Chicago

# OBSERVATIONS ON KNOCK-OFF RISERS AS APPLIED TO STEEL CASTINGS

By

S. W. Brinson\* and Joseph A. Duma\*\*

## ABSTRACT

*Knock-off risers are differentiated from necked-down risers. Some quantitative information is presented showing important relationship among the variables which govern the successful use of knock-off risers in cast steel applications.*

## Introduction

NECKED-DOWN RISERS are risers integrally attached to a casting by a short neck whose cross-sectional area is less than that of the riser, a condition which may or may not make the riser severable from the casting by flogging with hand sledges. The term, knock-off risers, as herein used, specifically refers only to risers which can be easily severed from the casting with manually powered hammers. The maximum round neck size for knock-off risers in B steel (Navy Department Specification 49S1), using a 12-lb. maul, is approximately  $2\frac{3}{8}$  in. In the absence of a fillet, and in the presence of a hot tear in the neck, larger diameter necks have been seen broken off. In general, necks exceeding this size manifest high resistance to fracturing. Hence, arbitrarily defined, knock-off risers are those whose neck diameters (or equivalent cross sectional areas) are necessarily under  $2\frac{3}{8}$ -in. (4.4 sq. in.). Because of this, application of knock-off risers is restricted to a comparatively narrow range of small casting sizes, while that of necked-down risers is not restricted to any particular range of casting sizes.

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The energy in foot-pounds required to rupture various neck sizes is shown in Fig. 2. Data for the graph was obtained by dropping weights from different heights onto the mid-point of the unsupported, projecting half of rigidly clamped, necked-down, steel test specimens. These specimens were of 0.20-0.30 per cent carbon cast steel in the as-cast condition, machined to have notches similar in design to the knock-off necks in Washburn risers, Fig. 1. Each point in Fig. 2 is the average of two values. Points designated with "X" are values obtained on a few production castings. Figure 2 integrates only one set of conditions, and should not be construed as being representative of the resistance offered by knock-off risers to flogging. There are many variables affecting notch sensitivity which have not been herein considered. From the data presented, it appears that the impact delivered by a manually powered 12-lb. maul is inadequate to fracture round necks in excess of  $2\frac{3}{8}$  in. in diameter.

The use of both knock-off and necked-down risers by the iron and steel industries dates back to 1908, the year Edwin C. Washburn obtained patent No. 900970 covering the use of necked-down risers. Since that time many foundries have used and discarded the necked-down principle. Relegation of the principle from general application to the proverbial dusty shelf of disuse was inevitable in the face of many reports of inconsistency of results. This is not surprising, for, until recent times, there was a lack of quantitative knowledge on critical relationships governing successful use of the necked-down principle.

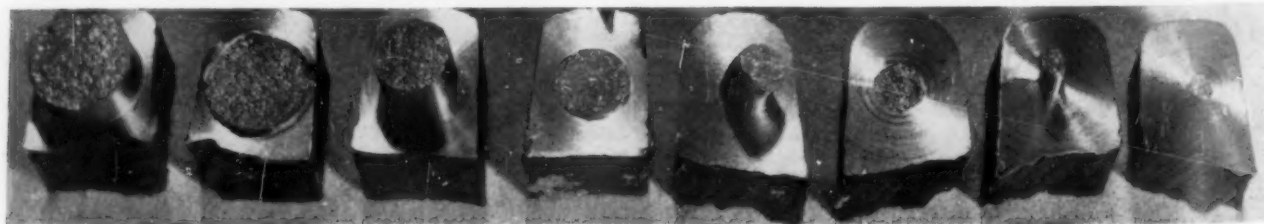


Fig. 1—Fractured "B" Steel Coupons with Machined Necks Simulating Washburn Necks. Metal in As-Cast Condition.

\* Master Molder. \*\* Foundry Metallurgist, Norfolk Naval Shipyard, Portsmouth, Va.



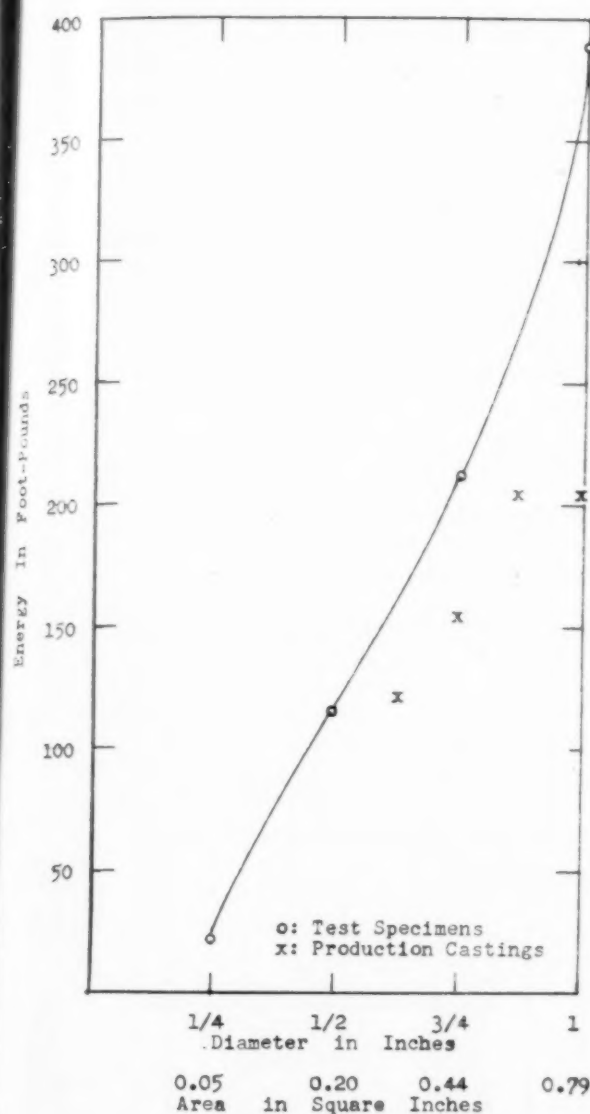


Fig. 2—Graph of Impact Strength of "B" Steel Shown in Fig. 1.

Of late, much work has been done on evaluating the pertinent variable relationships (see bibliography), in consequence whereof more consistent results are being obtained. The Steel Founders' Society of America has done much to clarify the confusion which surrounded the working of this principle in practice. The object of the subject report is to supplement the work of previous investigators and to throw further light on the workings and limitations of the principle, particularly with respect to the influence of depth of casting on the determination of the critical dimensions for knock-off risers.

By way of review, the successful application of knock-off risers depends upon a thorough understanding of the interrelationship of several pertinent variables. They are core thickness or height of riser neck, cross sectional area and shape of the neck, cross sectional size and shape of riser (determined by the cross sectional area and shape of section to be fed), and depth of casting section to be fed. It will be

TABLE 1—COMPOSITION OF SOME WASHBURN CORE MIXES IN CURRENT USE

Used by (See bibliography)	Washed and Dried Base Sand	Silica Flour	Bentonite	Cereal Binder	Core Oil	Other Binder	Moisture Content
(1)	Silica			1.3	2.2		
(2)	Silica			1.1	1.0	1.9a	2.5
(4)	Ottawa	10.0		1.5	1.5		4.6
(6)	Silica	8.0		Yes	Yes		
(9)	A.F.A. No. 56	30.0				3.0b	5.0
(9)	A.F.A. No. 50	17.0		1.0	1.0		5.0
Authors' A.F.A. No. 1	No. 60	30.0	0.6	3.0b		0.9c	6.0
Authors' A.F.A. No. 2	No. 100	2.0		1.4	2.6	0.3c	6.5

a Iron oxide

b Canary dextrin

c Resinous binder

TABLE 2—PHYSICAL PROPERTIES OF AUTHOR'S WASHBURN CORE MIXES

	Room Temperature Properties			
	Green		Baked 450 F—4 hr	
	No. 1	No. 2	No. 1	No. 2
Permeability, A.F.A. units	35	29	45	55
Compressive Strength, psi	3.1	1.4	586	342
Shear Strength, psi	0.8	Under 1	124	100
Tensile Strength, psi	5.5	5.2	126	161

High Temperature Properties (Own Atmosphere)			
Exposure Time Min.	Temp., °F	Compressive Strength, psi	
		No. 1	No. 2
20	1350	295	15
12	2000	440	60
8	2500	558	22

Gas evolved after exposure to 2000F for 1 min.			
No. 2.....	8.2 cc/g.		
No. 1.....	5.8 cc/g.		

noted that previous research has not evaluated the influence of the last variable (depth of casting) on the feeding efficiency of knock-off risers. Not that previous research overlooked this variable, but because it was found desirable to keep casting design constant and not run into centerline shrinkage which is very likely to occur in castings with height-to-width relationship greater than 1 to 1.

#### Test Methods and Procedure

The interrelation of the aforementioned variables on the feeding efficiency of knock-off risers was studied by making cylindrical billet castings in green sand, using circular design neck openings (referred to in the literature<sup>9</sup> as type A neck) gated and risered as illustrated in Fig. 3. Billet castings ranged in size from 2 to 5 in. in diameter, and varied in depth from a height-to-width ratio of 1 to 3. Billets were open-risered with top feed heads whose diameter exceeded that of the billet by 25 per cent. Steel for the billets was melted in a basic-lined, Moore "Lectromelt" arc furnace, and it was poured at approximately 2850 to

2900 F from a 2-ton bottom-pour ladle. A minimum of one-half inch of anti-piping compound was applied as a covering on all risers.

After shot blasting, the green castings and risers were sectioned longitudinally by sawing for examination of the solidity of interior metal. Castings which appeared solid after sawing were further checked radiographically for cavities.

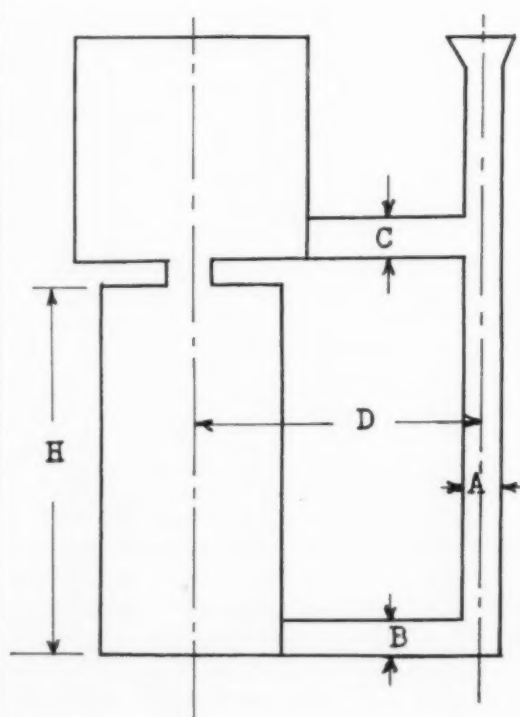
Table 1 gives the composition of knock-off core mixes which have been reported in the literature cited in the bibliography. In Table 1 are included the two mixes used by the authors. Mix No. 2, because of its lower gas content (Table 2), is used to make all thin cores, from  $1/8$  to  $1/2$  in. inclusive; while

mix No. 1, because of its higher strength is reserved exclusively for thick cores,  $5/8$  to  $3/4$  in., the venting of which is entirely dispensed with. In order to enhance the strength further, the cores are wired internally as shown in Fig. 5.

#### Discussion of Results Obtained

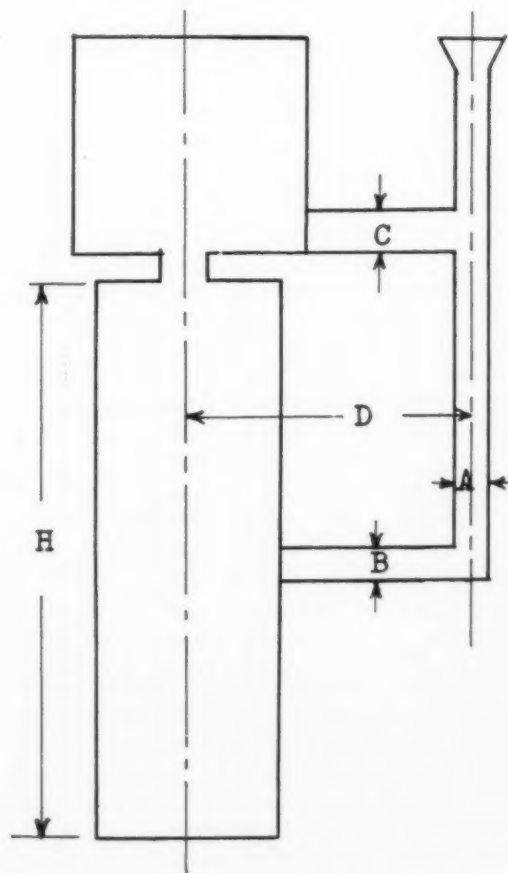
Results of the various tests performed on knock-off risers are discussed under several headings, corresponding with the several major variables influencing the effective functioning of knock-off risers in application. These are core thickness or neck height, extent and geometry of neck across section, casting and riser size, casting depth.

Fig. 3—Gating Arrangement and Dimensions Used in Making Castings.



Gating: (a)

Casting Dia. In.	Type Gate
2	(a)
4	(a), (b) for H=12 in.
6	(a), (b) for H=12 in.
8	(a)
10	(a)



Gating: (b)

A	B	C	D
Inches			
1	$3/4$	1	5
$1-1/4$	1	$1-1/4$	6
$1-3/4$	$1-1/2$	$1-3/4$	7
2	$1-3/4$	2	$8-1/2$
$2-1/2$	2	$2-1/4$	10

## Core Thickness

The knock-off riser core must possess certain properties, among these are low heat capacity to absorb minimum heat in reaching incandescence, thinness for rapid heat saturation, sufficient physical strength to resist breakage when hot metal surges through it during the pouring of the casting, and last but not least, chemical inertness with respect to steel.

Surprising as it may seem, steel discs would make good knock-off cores if they did not weld themselves to the casting (Fig. 12). Steel has three times the heat capacity of sand per unit of volume, and therefore absorbs three times as much heat per cubic inch to increase its temperature one degree. However, when used in a sufficiently thin disc whose total weight is less than one-third the weight of the sand core it replaces, it will absorb less heat for every degree rise in temperature than a sand core of the same diameter weighing more than three times as much. Because of the tendency to weld itself to the hot metal surrounding it—despite the use of oxide coatings and refractory washes to prevent such welding,—use of steel discs for knock-off cores is considered impractical. Carbon and graphite discs also would be suitable materials if they did not react chemically with the steel.

Concerning the thickness of sand cores, just what is meant by the axiom, "use as sufficiently a thin a

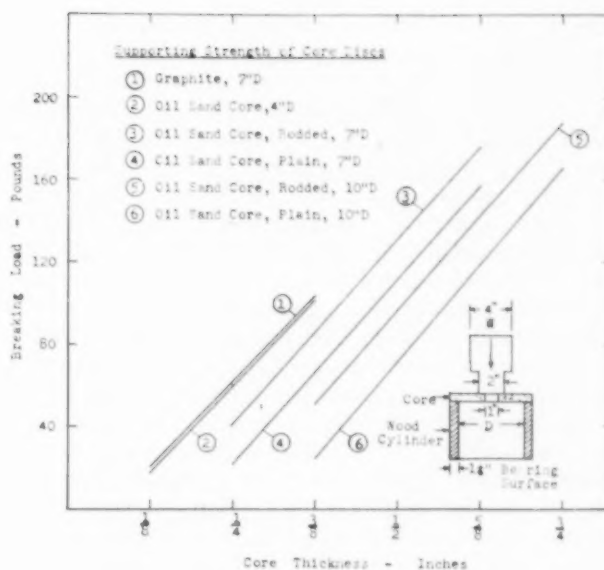
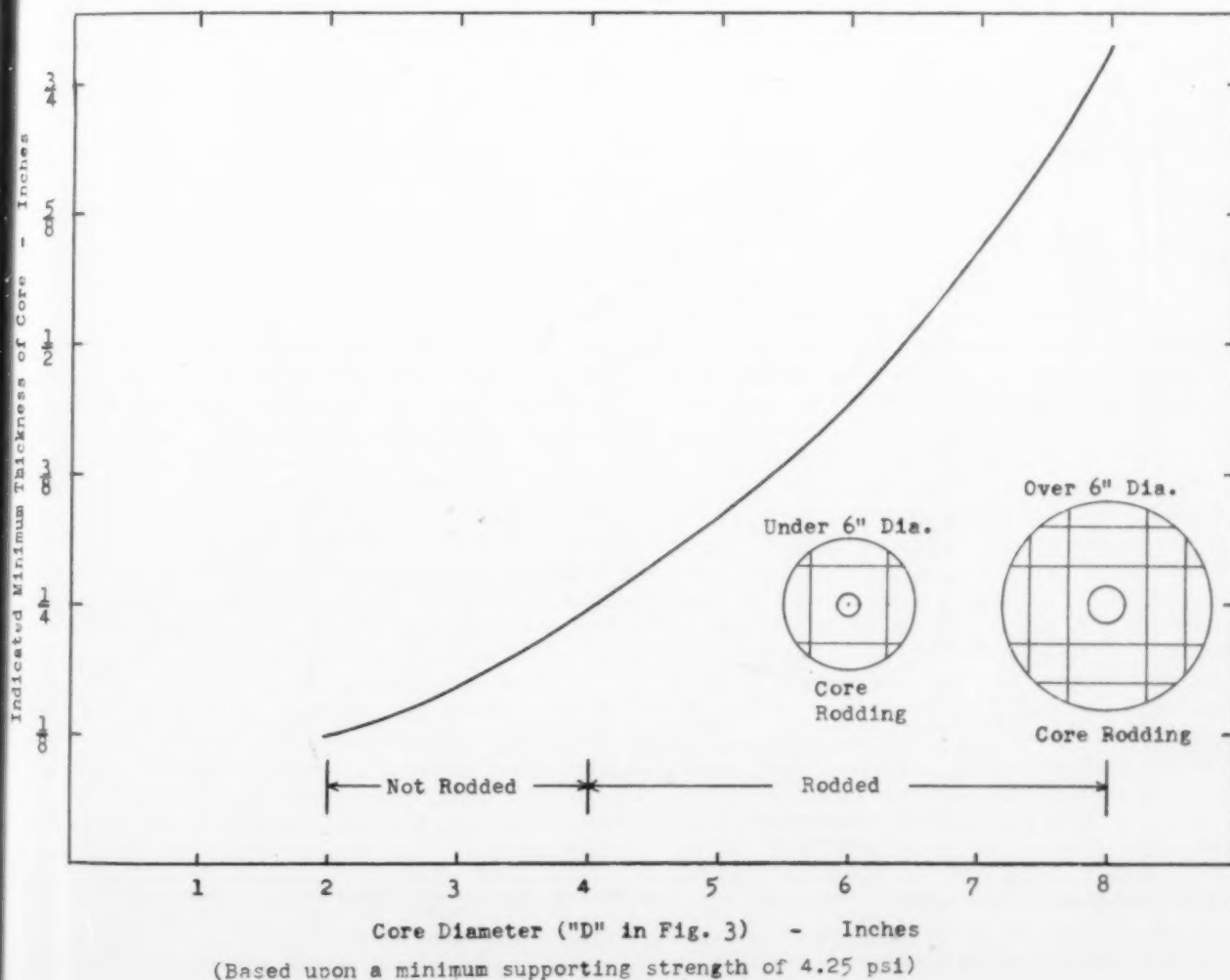


Fig. 4—Graph Showing Supporting Strength of Different Sizes of Cores.

Fig. 5—Indicated Minimum Thickness of Cores as Derived from the Data of Fig. 4.

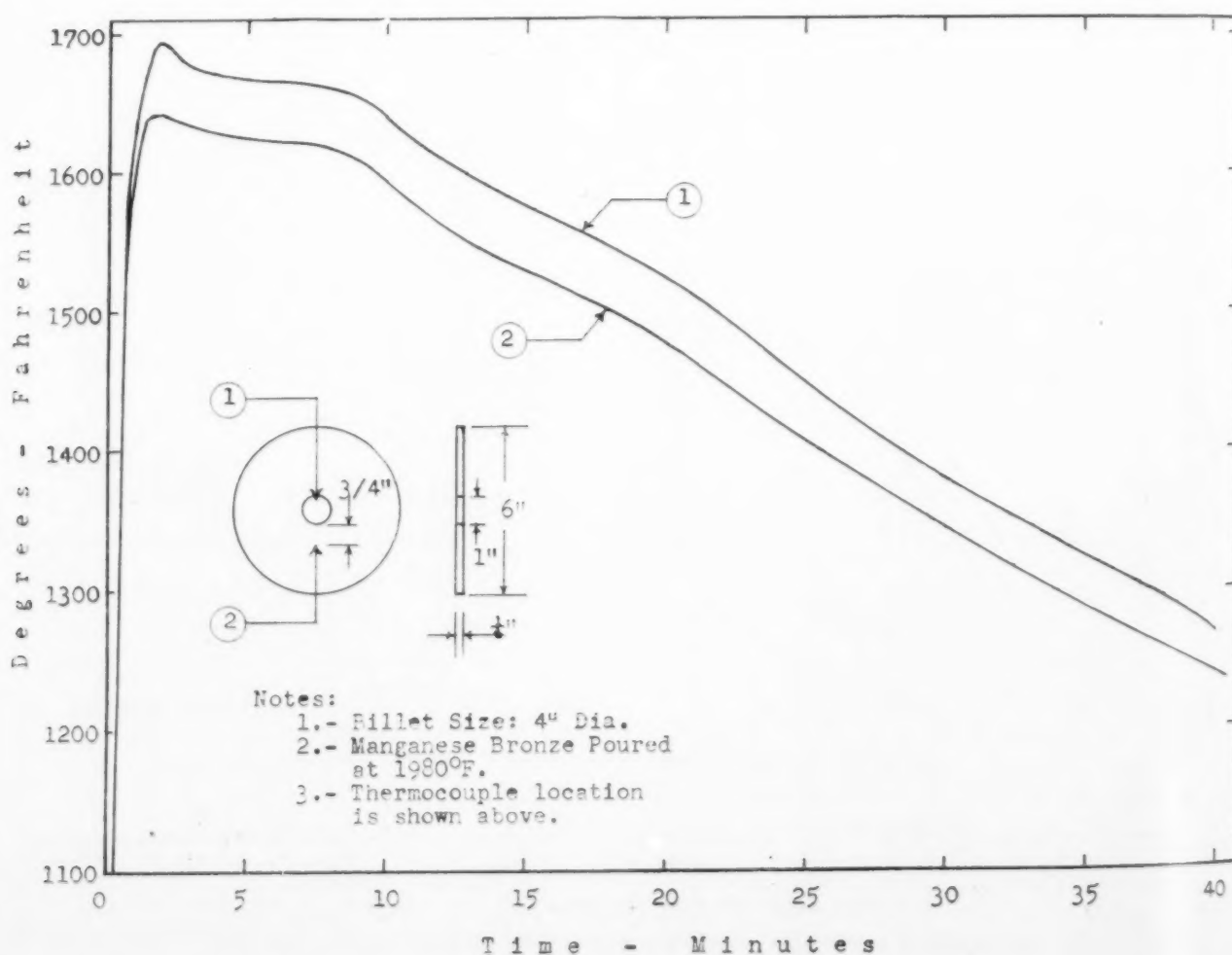


(Based upon a minimum supporting strength of 4.25 psi)



core as possible, consistent with its having the necessary strength"? Quantitative tests were made to answer this question. Figure 4 shows the resistance to breakage of different thicknesses and diameters of knock-off cores made from mix No. 2 (Table 1), and tested at room temperature in the manner illustrated in Fig. 4. In the large diameter cores the hole size was varied from 1 to  $2\frac{3}{8}$  in. without appreciable effect on strength results. Correlation of breaking strength values with actual shop tests found that any core which had a minimum supporting strength of approximately 4.25 psi, as determined by the method shown in Fig. 4, would successfully resist breakage at time of pouring when used in billets from 3 to 8 in. in diameter by 12 in. high carrying a riser 6 in. high. All things remaining equal, this strength factor more than any other seems to predetermine the minimum thickness of core needed. For example, a 4-in. diameter billet having an area of 12.6 sq. in. will require a core whose supporting strength should be at least 53 lb. ( $12.6 \times 4.25 = 53.5$ ), while a 7-in. diameter billet having an area of 38.5 sq. in. will require a core whose minimum supporting strength should approximate 163 lb. ( $38.5 \times 4.25 = 163.6$ ). From Fig. 4, it is seen

Fig. 6—Time-Temperature Curves for  $\frac{1}{4}$ -in. Thick Core Used Around the Neck of Manganese Bronze Billet.



that the thinnest 4-in. and 7-in. diameter cores which will meet the above strength requirements are  $\frac{1}{4}$  in. and  $\frac{9}{16}$  in. thick, respectively. Based on a minimum supporting strength of 4.25 psi of surface area, the thickness of core required to obtain this strength in different diameter sizes is shown in Fig. 5. The graph in Fig. 5 was derived from the data of Fig. 4 by multiplying the area in square inches of "D" diameter cores by 4.25, thereby obtaining the value of resistance to breakage needed in pounds, and from this breakage value the thickness of core just sufficient to resist such breakage. Results obtained by the authors are those which are associated with one specific method of testing, one specific core mixture, and one specific shop practice.

Cores made thinner than those recommended in Fig. 5 are liable to break on pouring, not so much in the thick sizes ( $\frac{1}{2}$  in. and over) as in the thin size. Cores made thicker than those recommended in Fig. 5 will function effectively provided the amount of thickening is properly related to the hole or neck size, and to the depth or height-to-width ratio of the casting (Fig. 17). These relations are discussed under their respective headings later in the paper. Wherever possible, knock-off cores should be made as thin as possible consistent with their having necessary strength to resist breakage in use. They should be selected according to some graph as in Fig. 5. Recent studies on segregation in small steel castings reported that

severity of segregation of carbon, and to a minor degree of other alloy elements, found in the area immediately below riser contacts of the Washburn type, varies directly with the thickness of the knock-off core. Thin cores are associated with minimum, in many cases with negligible, segregation. Thus, it behooves us to use as thin a core as is practically possible.

Besides mechanical strength, it was stated that cores had to have rapid heat penetration. The time-temperature curves in Figs. 6 and 7 are enlightening in this respect. Thermocouple No. 2 in Fig. 6 shows the rate of heat penetration from manganese bronze into the center of a  $\frac{1}{4}$ -in. thick core. Thermocouple No. 1 shows the temperature prevailing at the same instant in the outer portion of the neck. This core gave a sound casting. Figure 7, shows the rate of heat penetration from molten steel into the center of  $\frac{1}{4}$ -,  $\frac{1}{2}$ -, and 1-in. thick cores. The 5-in. diameter by 8-in. high steel billets made with the  $\frac{1}{4}$ - and  $\frac{1}{2}$ -in. thick cores came out solid whereas the same size billet made with the 1-in. thick core did not come out solid. It is significant that the maximum temperature obtained in the 1-in. thick core at no time reached the solidus temperature of the steel. Either one of two conditions might have aided in making this thick core casting solid, namely, reducing the height of the billet to equal the diameter, or enlarging the neck cross section from 1 in. to  $2\frac{1}{2}$ -3 in. in diameter. It would seem from this test that for efficient feeding of castings having height-to-width ratio in excess of 1 but less than 1.5, the thickness of knock-off cores should be such as to allow the center of the core to heat up to at least the solidus temperature of the metal.

#### Neck Size

Having determined the minimum thickness of core that can be used without fear of breakage, it remains next to find the proper size of hole for the different thicknesses of cores and size of castings. This has been ably done by other investigators<sup>9</sup> for castings of height-to-width relationship of 1 or less. The authors wish to add, however, that recommendations for knock-off core geometry which are founded on studies of normal cylinders do not apply to sections or cylinders whose height exceeds their diameter.

The size of round neck contacts for knock-off cores varies between two limits, one maximum and one minimum, namely,  $\frac{3}{8}$  in. and  $2\frac{3}{8}$  in. diameter. Practical difficulties attending pouring and running deter the use of any diameter size below  $\frac{3}{8}$  in., and, since it is difficult to knock-off any riser attached to a neck greater than  $2\frac{3}{8}$  in. in diameter, knock-off necks are necessarily restricted to diameters below this value. Summarizing the results of numerous shop and laboratory tests made on all manner of round billet castings up to 10 in. in diameter, it has been found that, for the core thicknesses recommended in Fig. 5, a satisfactory size of hole of the straight edge type is one whose diameter is approximately equal to 25 per cent of the diameter of the casting. In terms of core thickness, the diameter of round neck contacts should be approximately 3 to 4 times the thickness of the core. While these results are not in agreement with some

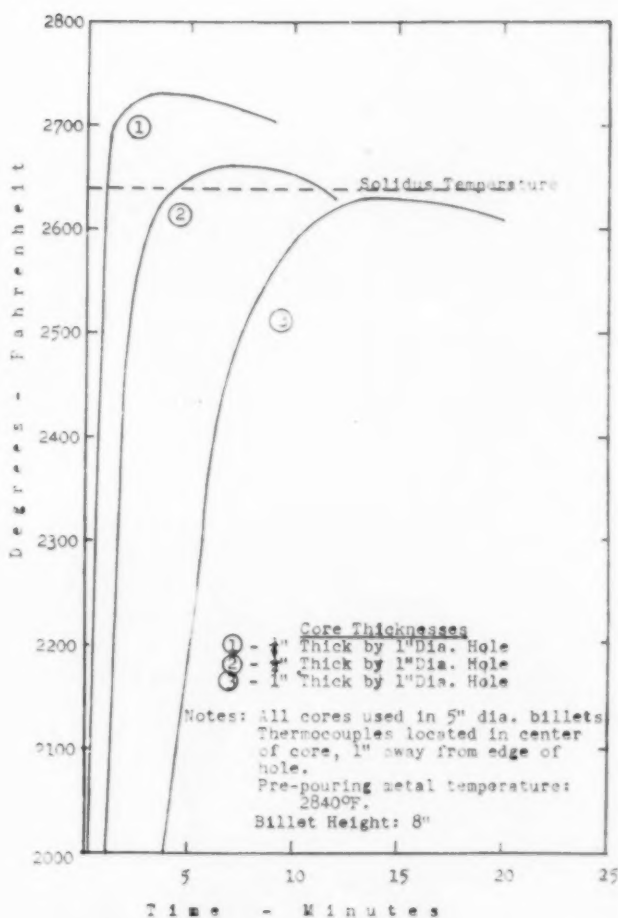


Fig. 7—Time-Temperature Curves in Different Thicknesses of Knock-Off Cores Used in Steel.

published findings, it should be remembered that the cores used by the authors of this paper are thinner by approximately one-half, and the castings which they made are somewhat deeper than those which have been reported in the literature. This difference, however, should add to, and in no way detract from, the usefulness of the data contained in this and other papers.

Sensitivity to hole size increases with core thickness. Thin cores are remarkably insensitive to changes in hole or neck size, especially when the casting is no deeper than its diameter, or when the casting is tapered for directional solidification. Examination of the cone castings pictured in Figs. 8 and 9 will bear out the validity of this observation. These cones have a base diameter of 4 in., are 6 in. high, and each is attached to a riser through a reduced neck of variable diameter but constant,  $\frac{1}{4}$ -in. high neck, indicating the employment of  $\frac{1}{4}$ -in. thick knock-off cores. The cone castings were poured in clusters of six to a mold to enable use of six different diameters for neck openings, which, from left to right, Figs. 8 and 9, are  $3\frac{3}{4}$  in., 3 in., 2 in., 1 in.,  $\frac{3}{4}$  in., and  $\frac{1}{2}$  in. At the end of a timed interval after pouring, each mold was given a 180-degree reversal to allow the unsolidified liquid metal in each to run out and thus disclose by the thickness of the remaining shell the influence of

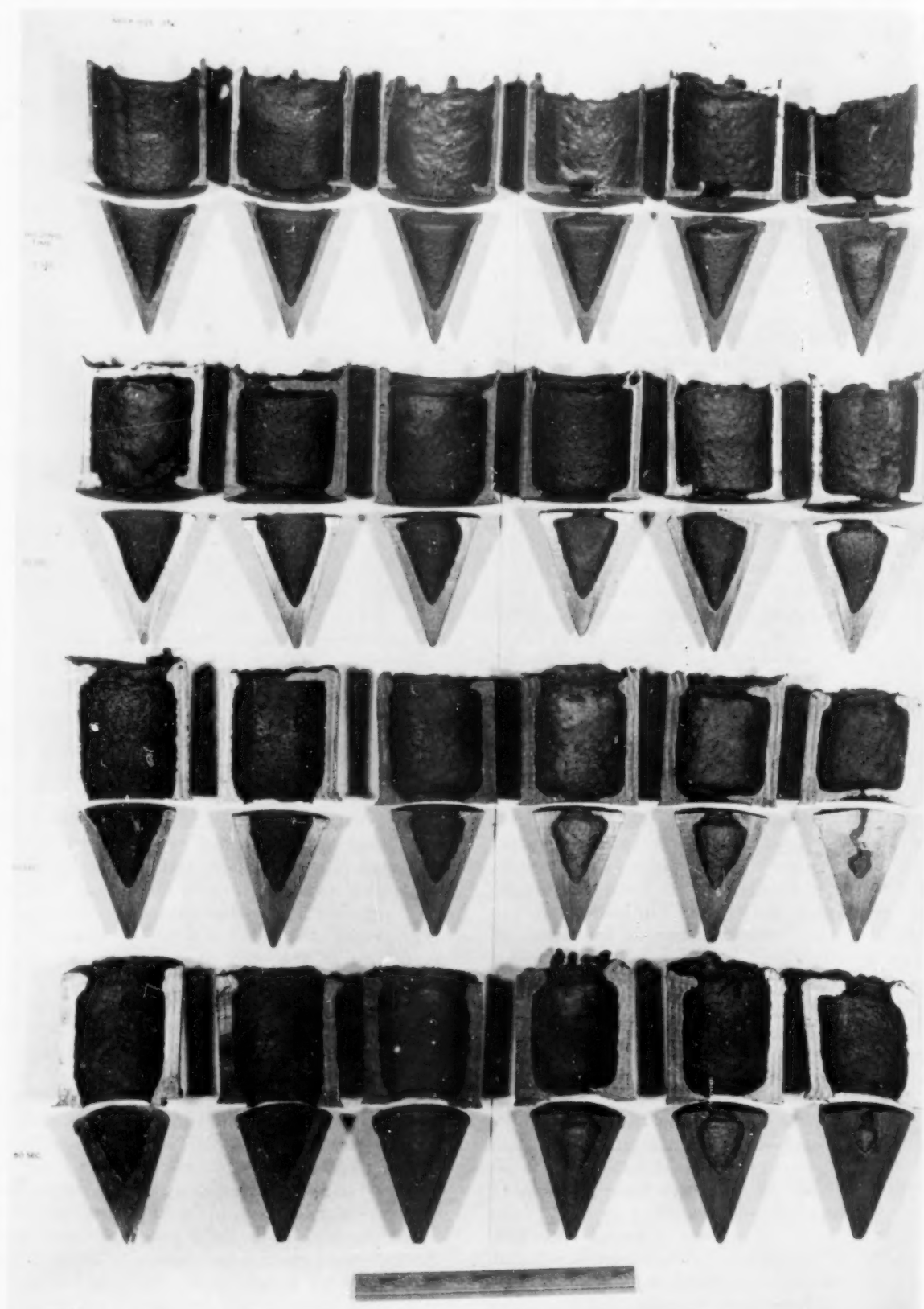


Fig. 8—Steel Cone Castings Using  $\frac{1}{4}$ -in. Thick Knock-Off Cores. Castings were bled at timed intervals after pouring.



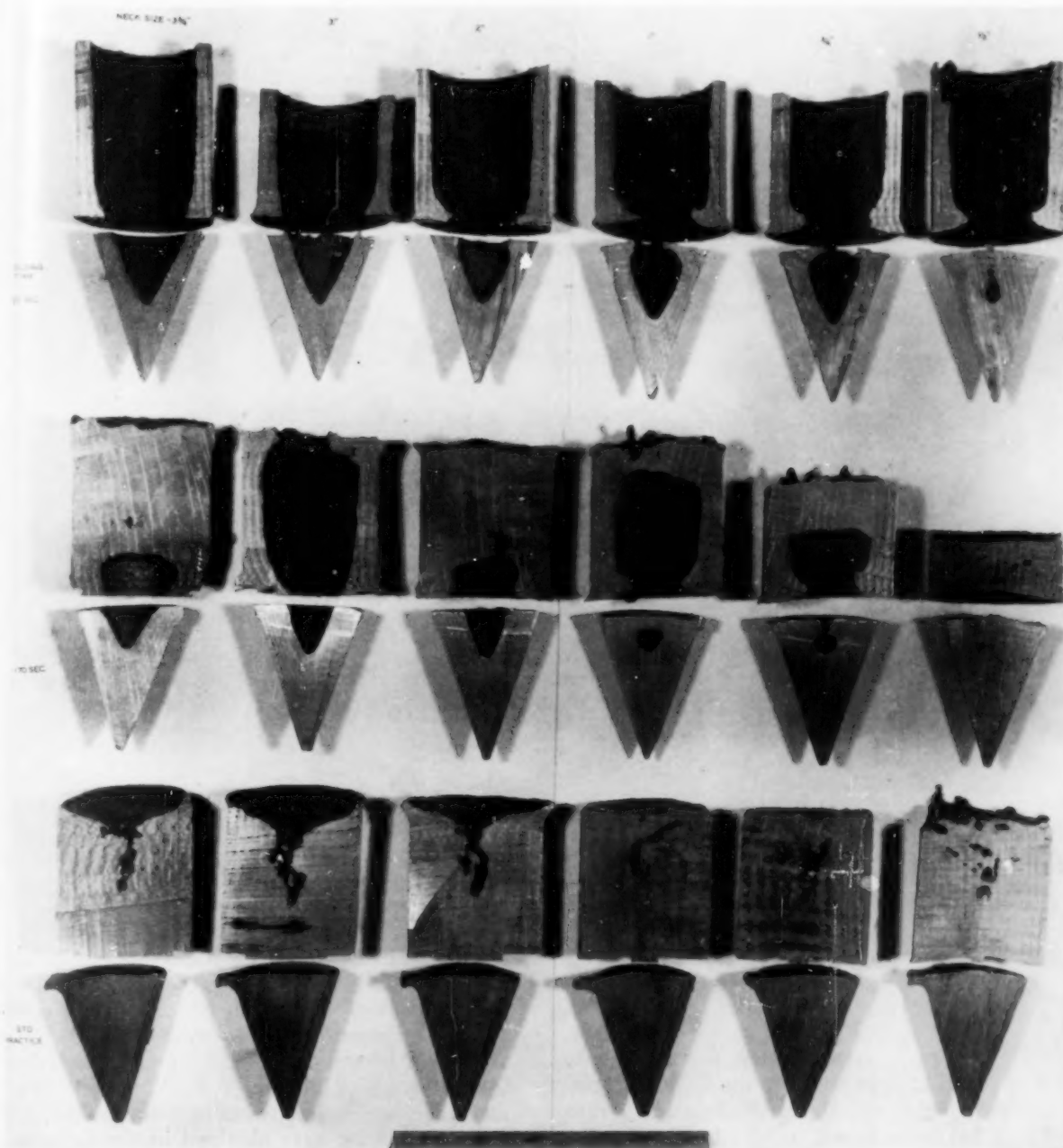


Fig. 9—Steel Cone Castings Using  $\frac{1}{4}$ -in. Thick Knock-Off Cores. Castings were bled at timed intervals after pouring.

the knock-off core upon the progress of solidification. This influence of the core on the solidification of steel in both casting and riser is shown in Figs. 8 and 9. Castings in the last row are all solid, attesting to the fact that thin ( $\frac{1}{8}$  to  $\frac{1}{4}$  in.) cores are hardly, if at all, critically related to neck diameter.

#### Casting Size

The authors have had little success in the application of knock-off risers to castings larger than 8 in. in

diameter. The neck sizes required to feed solidly castings of this size are necessarily so large that the risers which they carry are truly no longer knock-off risers, but, more appropriately, fall under the heading of necked-down risers.

#### Depth of Casting

In the many instances of feeding failures which epidemically cropped up on different but seemingly the same size design of castings, the one important

Billet	2 Dia. x 2	2 Dia. x 4	2 Dia. x 6	2 Dia. x 8	2 Dia. x 8	2 Dia. x 8	2 Dia. x 8
Core							
Thickness	$\frac{1}{2}$	$\frac{1}{2}$	$\frac{1}{2}$	$\frac{1}{2}$	$\frac{1}{4}$	$\frac{1}{4}$	None
Neck Dia.	1	1	1	1	$\frac{1}{2}$	$\frac{1}{8}$	None
Riser	$2\frac{1}{2} \times 7$	$2\frac{1}{2} \times 7$	$2\frac{1}{2} \times 7$	$2\frac{1}{2} \times 7$	$2\frac{1}{2} \times 7$	$2\frac{1}{2} \times 7$	$2\frac{1}{2} \times 7$

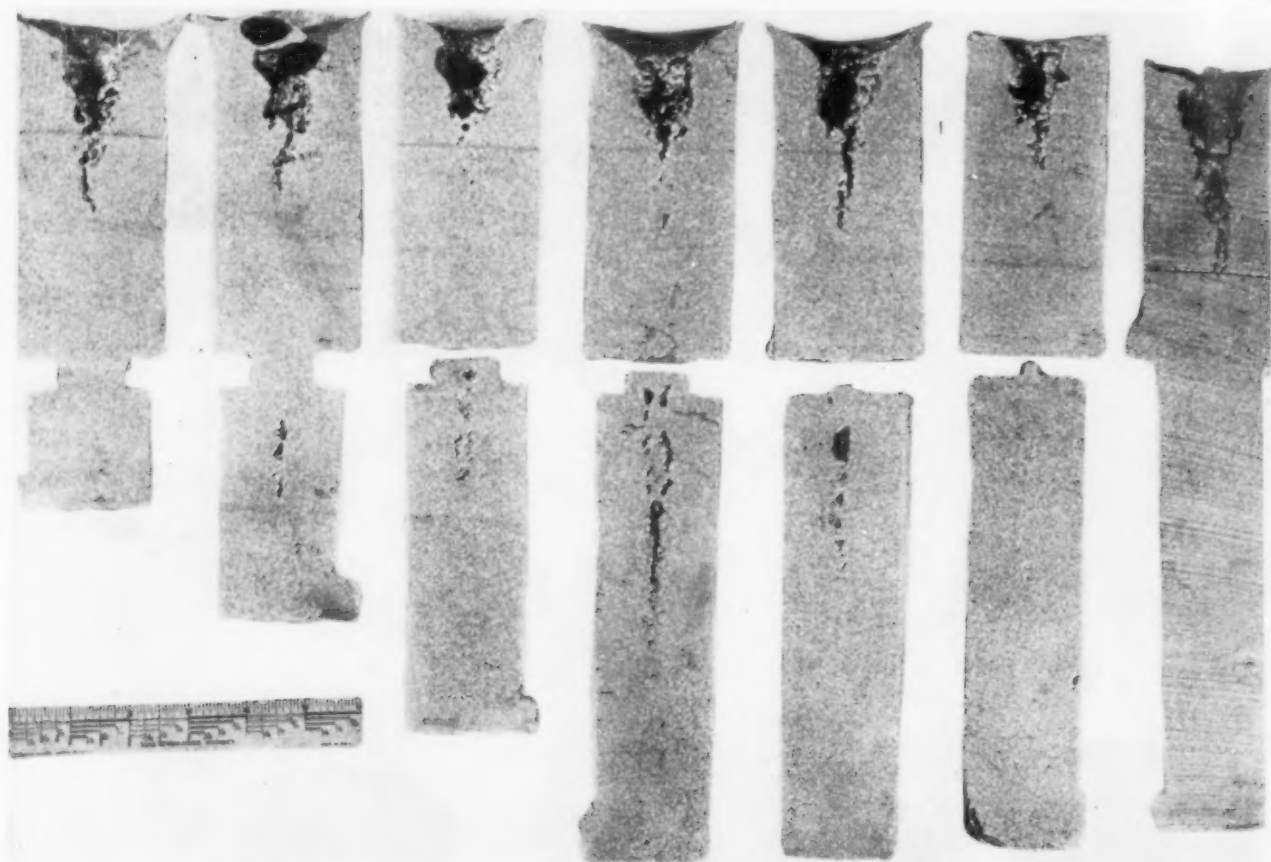


Fig. 10—Showing 2-in. Dia. Billets of Different Lengths Made with Knock-Off Cores.

variable which was overlooked and not taken into account in practical application of the knock-off principle was the height-to-width relationship of the casting. For it has been found, as will be shown, that the core geometry relationships of thickness and hole diameter which have been established as being productive of sound castings for a given height-to-width ratio are operative only so long as that height-to-width ratio remains unchanged.

In Fig. 10 are shown seven 2-in. diameter castings, the first four of which are made in four different body lengths, namely, 2, 4, 6, and 8 in. The same design of core ( $\frac{1}{2}$  in. thick, 1 in. diameter riser neck hole) was used in making the first four castings. While the first casting, 2 in. diameter by 2 in. high, is solid, the other three are not. In fact, each succeeding taller casting is progressively worse with respect to amount of internal shrinkage contained. The fifth casting was made with a thinner core and a smaller diameter neck. Though not solid, it contains less shrinkage than the fourth casting. Decreasing the thickness of the core still further, that is, to  $\frac{1}{8}$  in., resulted in

what appeared, on visible inspection, to be a solid casting. Radiographic examination, however, disclosed the presence of negligible (three spots, each about  $\frac{1}{8}$  in. in diameter) shrinkage. The seventh casting was made in the conventional manner, without the use of any knock-off core. It is not entirely free from shrinkage, but is slightly better than the sixth casting.

Similar results were obtained in the series of 4-in. diameter castings shown in Fig. 11. From this evidence, it is seen that susceptibility of castings to shrinkage increases in proportion to their length. To obtain freedom from shrinkage in billets 2 and 3, Fig. 11, either thinner cores or larger necks are needed (Fig. 13). In Fig. 12 are pictured other interesting deviations from conventional practice in the making of 4-in. diameter billets by 12 in. high. The first casting on the left in Fig. 12 is illustrative of the degree of solidity obtainable in this size with top risers without use of a knock-off core. Since some shrinkage is evident even with the full neck riser, one can hardly expect improvement in this respect with a knock-off core no matter how thin it is. The third casting from the left in Fig. 12 is representative of the best result that is obtainable with a knock-off core. Lightly padding the castings, after the manner of the last two billets in Fig. 12, aids in making the casting

Billet Dia.	4	4	4	4
Billet Height	4	6	8	12
Neck Dia.	$\frac{7}{8}$	$\frac{7}{8}$	$\frac{7}{8}$	$\frac{7}{8}$
Core Thickness	$\frac{1}{2}$	$\frac{1}{2}$	$\frac{1}{2}$	$\frac{1}{2}$
Riser	5 Dia. x 6 High	5 Dia. x 6 High	5 Dia. x 6 High	5 Dia. x 6 High

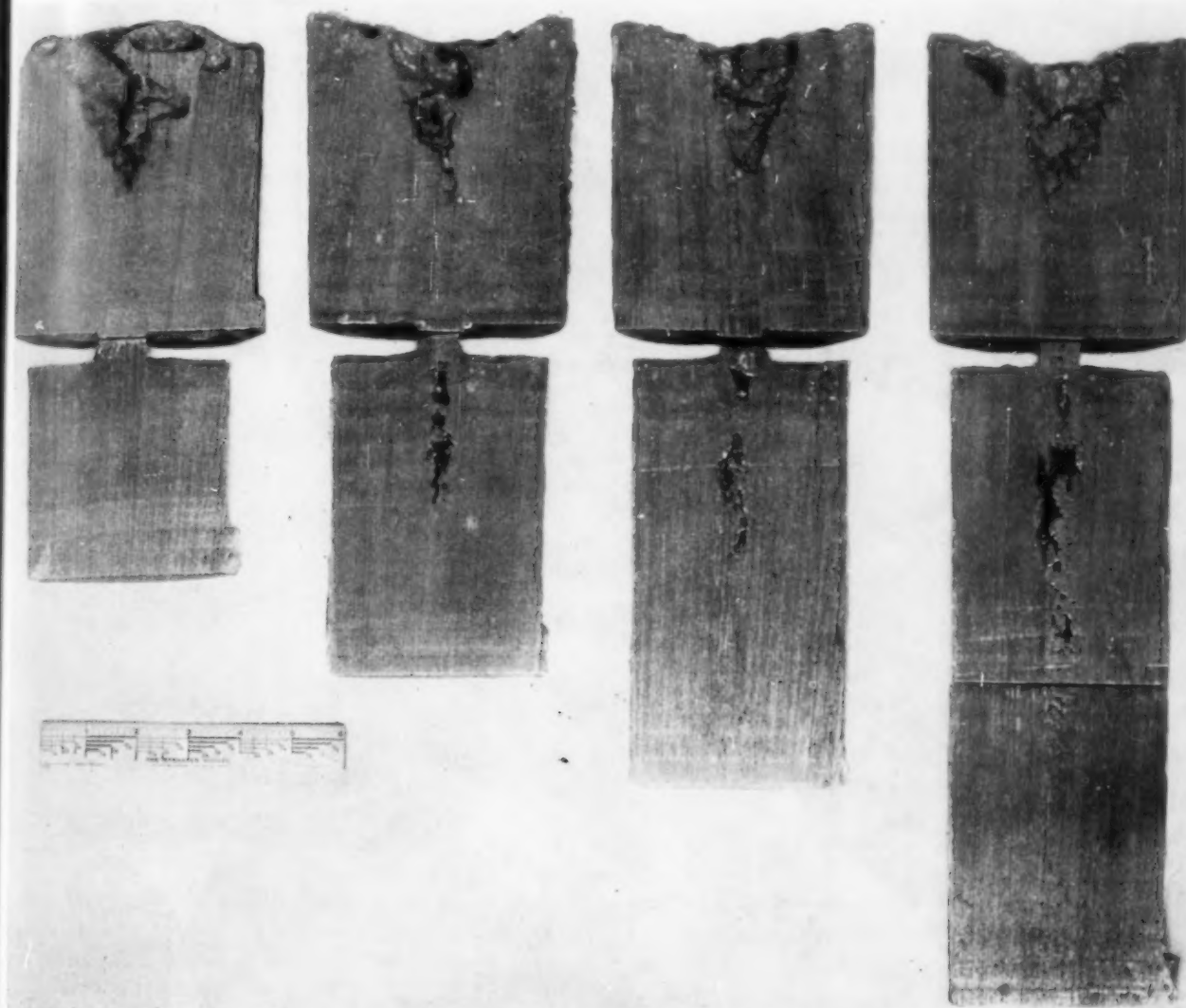


Fig. 11—Showing 4-in. Dia. Billets of Different Lengths Made with Knock-Off Cores.

sound provided the core is not too thick. In the 6-in. diameter billets, Fig. 14, the same marked tendency toward shrinkage in the longer castings is manifest. From an inspection of the third (from left to right) casting in Fig. 14, it is seen that the amount of shrinkage is materially decreased by the use of a thin core.

As for the 8-in. and 10-in. diameter billet sizes some of which may be seen in Figs. 15 and 16, considerable difficulty has been experienced in making them solid, even in the normal sizes. With neck sizes restricted to under  $2\frac{3}{8}$  in. in diameter, sound castings are at times obtained. More often than not, however, some unsoundness develops directly under the neck. Because consistently sound castings have not been obtained

in 8-in. diameter sizes, the curve for this size in Fig. 17 is shown dotted in. Larger neck sizes may be the answer to consistency of results, but not the answer to knock-off risers. Indications are that the limit of knock-off risers is reached on 8-in. diameter billets, that this appears to be where knock-off risers yield place to necked-down risers.

#### Summary

The variables, core thickness, neck size, casting size, and casting depth, which have been previously discussed, are shown in their proper relationship one to the other in Fig. 17. The curves in the graph crystallize the authors' observations on knock-off cores both with respect to shop practice and shop and field tests. The curves show the minimum size of hole and thickness of core required to feed effectively castings having the diameter and length stated on the curve. The



Billet Dia.	4	4	4	4	4
Billet Height	12	12	12	12	12
Neck Die	Solid	1 3/8	1	1 1/2	1 3/4
Core Thickness	None	3/32 Steel	1/4	3/8	1
Riser	5 x 6	5 x 6	5 x 6	6 1/2 x 8	6 1/2 x 6
Padded	No	No	No	Yes	Yes

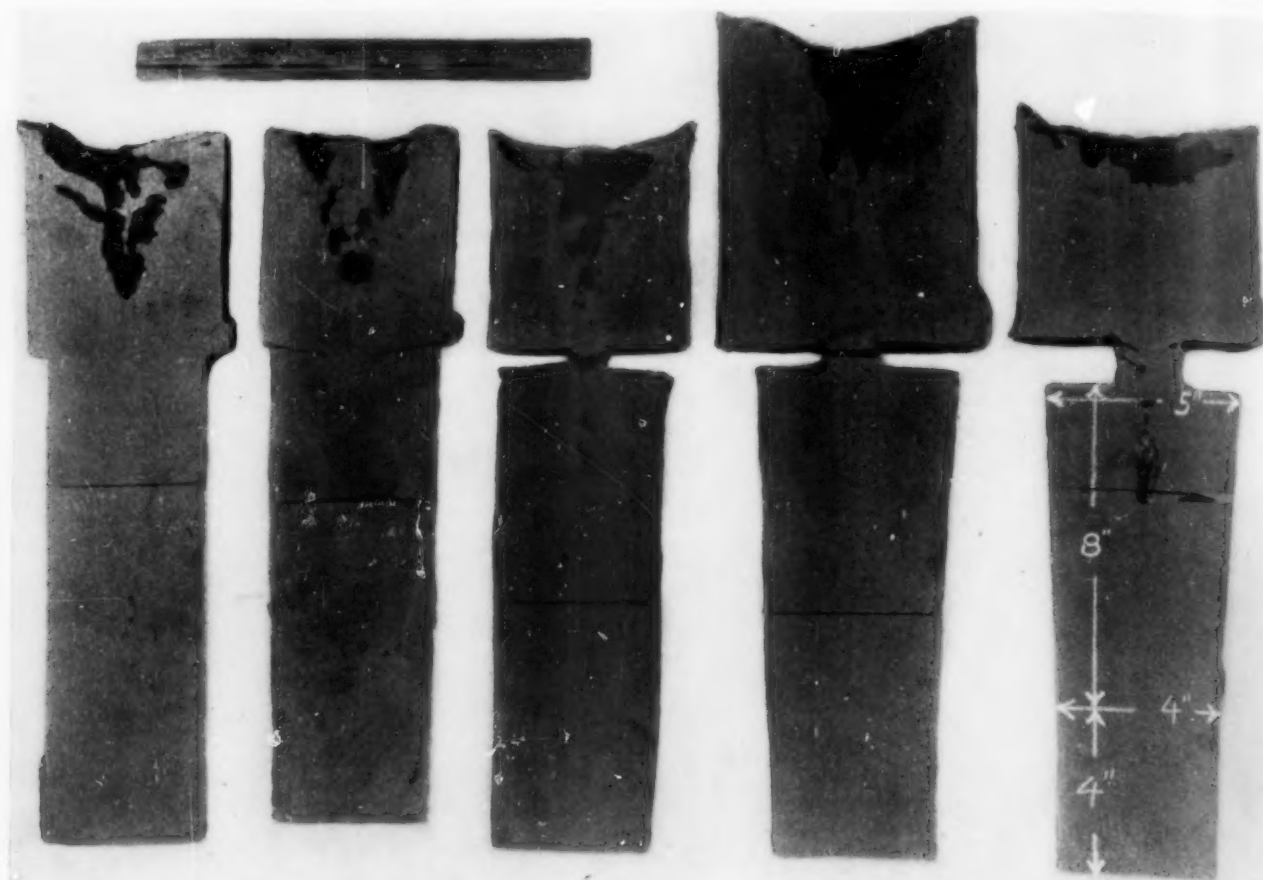


Fig. 12—Showing 4-in. Dia. Billets All 12 In. Long Made with Knock-Off Cores.

lines on the chart represent a boundary condition in that values above a line produce a solid casting, while values below a line will result in shrinkage cavities being formed in the casting. It is self-evident that castings of shorter length than those for which the curves are drawn will also be sound. The curves are therefore applicable to castings equal to, or shorter in length than those for which curves are given. Castings longer than those indicated on the curves cannot be made sound with the use of knock-off cores. The center-line shrinkage which is wont to afflict such designs must first be overcome.

The conclusions which can be drawn from the graphs presented in Fig. 17 are as follows:

1. Knock-off risers with circular neck sizes greater than 2 3/8 in. in diameter do not knock-off when hit with manually powered mauls. The diameter value of 2 3/8 in. is not a fixed value; in the absence of hot tears, it varies from 2 1/4 to 2 1/2 in.

2. The maximum depth of casting which can be fed with open knock-off risers appears to be 6 in., 8 in.,

8 in., and 8 in. respectively for 2-in., 4-in., 6-in., and 8-in. diameter billets.

3. To feed the depth of casting given above, the thickness of core should approximate that of Fig. 17. Greater thicknesses will not allow feeding to continue to the degree required to make sound castings. Short castings with length-to-width ratios of one or less than one are not critically reactive to core thickness.

4. The authors were unsuccessful in making sound 10-in. diameter billets using various thicknesses (1/2 in. to 1 1/2 in.) of knock-off cores with neck sizes from 1 1/2 in. to 2 1/2 in. in diameter. In this connection, tests are in progress using plastic fire brick cores.

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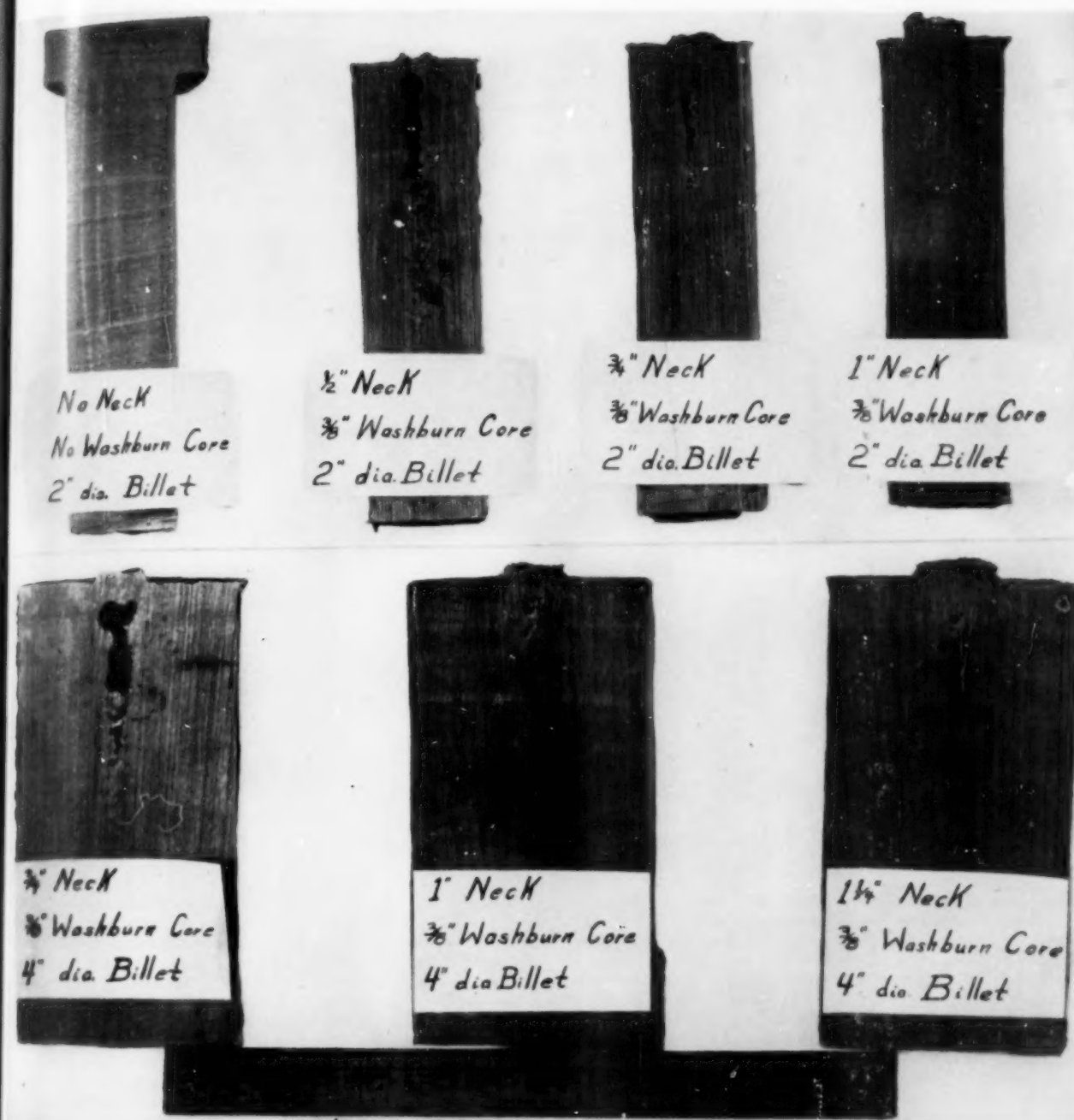


Fig. 13—Showing Some 4- and 2-in. Dia. Billets by 8 In. Long Made With Knock-Off Cores.

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## DISCUSSION

Chairman: W. W. MOORE, Burnside Steel Foundry Co., Chicago

Co-Chairman: H. F. TAYLOR, Massachusetts Institute of Technology, Cambridge, Mass.

E. C. TROY (Written Discussion):<sup>1</sup> The discussor would like to thank the authors S. W. Brinson and J. A. Duma for their contribution to much needed foundry literature.

Analysis of this paper gives confirmation for previously experienced practical difficulties with the use of the Washburn principle on neck-down risers.

Since the difficulty of feeding uniform sections increases with the length or depth of such sections under any system of risering, it is not surprising to find that necking down of the riser results in even greater unsoundness. That the reduction of the riser contact, through the use of even very thin cores, will result in a less effective system must be expected. For, as the authors point out, the core used will absorb heat. This heat is absorbed from the metal at that location where final casting solidification is desired. The work of the authors should not discourage the

<sup>1</sup> National Engineering Co., Philadelphia.

Billet Dia.	6	6	6
Billet Height	6	12	12
Neck Dia.	2	2	1 $\frac{5}{8}$
Core Thickness	$\frac{5}{8}$	$\frac{5}{8}$	$\frac{3}{4}$
Riser	8 Dia. x 6 High	8 Dia. x 6 High	8 Dia. x 6 High

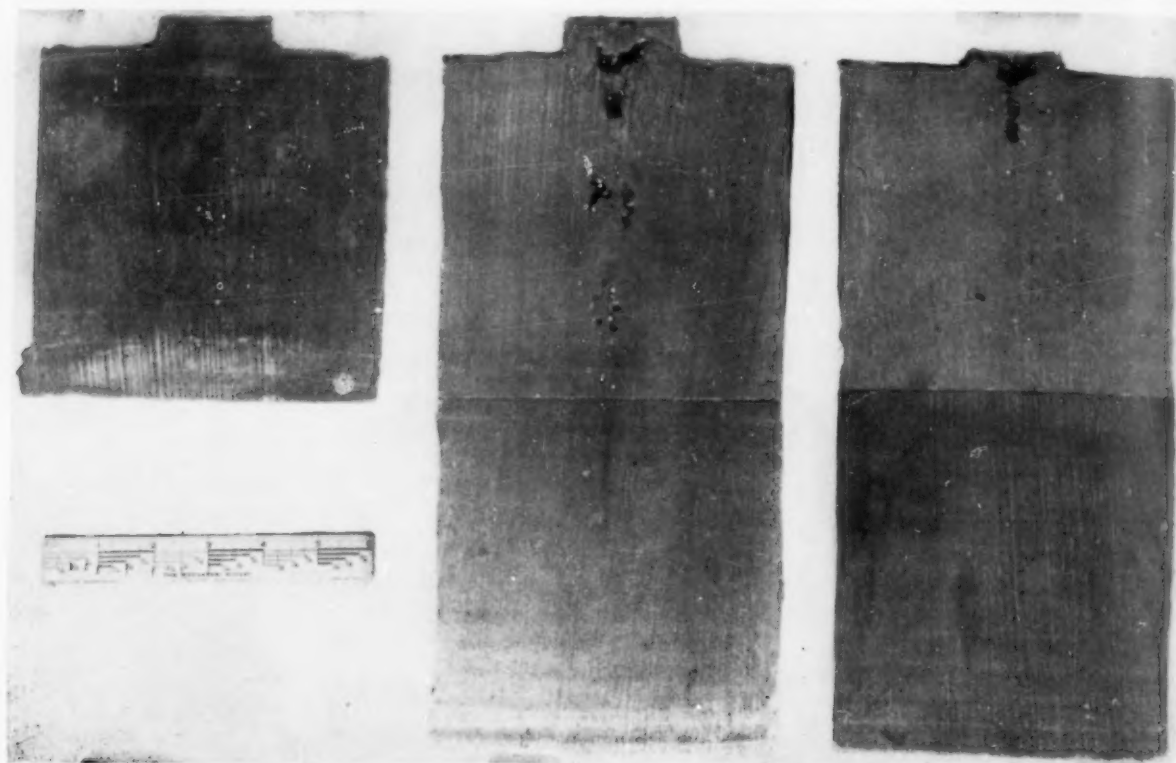


Fig. 14—Showing 6-in. Dia. Billets of Different Lengths Made With Knock-Off Cores.

use of the neck-down riser but rather aid its proper application.

In their introduction the authors state "From the data presented, it appears that the impact delivered by a manually powered 12-lb maul is inadequate to fracture round necks in excess of 2 $\frac{3}{4}$  in. in diameter." Such data does not appear in the text.

The correlation attempted by the authors between cold, static localized loading of the neck-down core portrayed in Fig. 4, and the loading by hot surging metal on either or both sides of the core in the mold, seems to the discussor inoperative. Many such cores fail several minutes after the pouring has been completed. Probably due to the destruction of the binders by the heat and yielding to the buoyant forces. The type binder and total exposed area of the core being the important variables.

In discussing the results of Fig. 7 the authors suggest that the neck-down core should heat to a temperature at its center equal to or above the solidus of the metal poured, if it is to function satisfactorily. When discussing the failure to feed increasing the diameter of the neck opening would probably result in a sound casting. Why would the larger neck opening result in a higher temperature at the center of the core?

MR. BRINSON: The only data we had was practical data. The data referred to was the result of actual demonstrations of knocking off risers with a 12-lb maul. We realize that the correlation between the cold static test and the hot test is only indicative. It is indicative to the same extent as test bars for cast steel, cast iron, or any other metal which is properly fed. Those test bars are indicative of the metal that you have in the casting. It is seldom that you can select one casting, cut a section from that casting and obtain the same results you do from the test bar. You have other variables that enter into the problem. But the curves do show whether with a certain type of core mixture the static load we apply cold will give satisfaction or not. It is just indicative. We are able to show just what

strength is required with certain types of binders and sand that will stand up under the hot surging of the metal.

In reference to the remarks on temperature, I think it is shown by the different charts in the paper that the thicker the core, the larger the diameter of the neck. That is probably due to the fact that the larger the diameter of the neck, the longer it will take for that neck to freeze, and it will feed a little longer than it would if the neck was smaller. With a thin core you can use a smaller diameter neck than you can with a thick core. If you increase the thickness of the core, you have to increase the diameter of the neck. The more mold material you have, the lower will be your equilibrium temperature; the higher conductivity you have in your sands, the lower heat you are going to reach in your mold; or the greater mass of metal you have in your mold the higher it is going to heat the mold.

We have had that point demonstrated in the following example. If you try to make a large thin casting in concrete with a large mass of mold material, you will generally have trouble due to cracks, because you do not have enough heat in the mass of the metal to properly break up the mold so it will disintegrate and yield when your casting solidifies. You will have a cracked casting instead of a good casting. We have had that happen.

We feel that this larger neck diameter would feed because of the mass of metal in that neck.

MR. DUMA: That temperature was mentioned just as an incidental result of the test. All other things being equal, two facts were noted concerning the action of the 1-in. thick core, namely, first it marked the dividing line between effective and ineffective feeding of the casting, and secondly, the temperature reached in the interior of said core was less than 2650 F. Because of this, the statement was ventured somewhat cautiously and not too assertively that if the temperature in the interior of a core did not exceed 2650 F the chances were that you would





Fig. 15—Showing 8-in. Dia. Billets by 8 In. High Made With Knock-Off Cores.

not get a solid casting. It is not axiomatic that it always will do so.

J. B. CAINE (Written Discussion):<sup>2</sup> The authors should be thanked for a valuable contribution to the proper application of necked-down risers, the effect of depth of section. The variable has undoubtedly been the cause of mysterious failures of necked down risers on certain castings in the past. However, one point should be emphasized in the interest of the casual reader. The authors have rightfully selected the minimum neck size for a given core thickness that will feed a section whose height is equal to its diameter to illustrate the effect of

increasing height. This does not mean that a necked-down core with a larger opening will not feed the section. Perhaps the opening size will be larger than the maximum prescribed by the authors for "knock-off" risers, but will still be small enough to result in an appreciable decrease in the number of square inches that must be cut and ground.

For example, the 1/2-in. thick core used in Fig. 10 with a 1-in. diam opening falls just below the line between shrinkage and solidity established in Ref. 9. In this case a thinner core is indicated and, as shown by the authors will result in a sound casting. Similarly, the 1/2-in. thick cores with 7/8-in. diam openings in Fig. 11 fall on the line between shrinkage and solidity

<sup>2</sup> Sawbrook Steel Castings Co., Cincinnati

Billet Dia.	8	10
Billet Height	8	9
Neck Dia.	2	2
Core Thickness	$\frac{3}{4}$	$\frac{3}{4}$
Riser	10 Dia. x 8 High	12 Dia. x 8 High

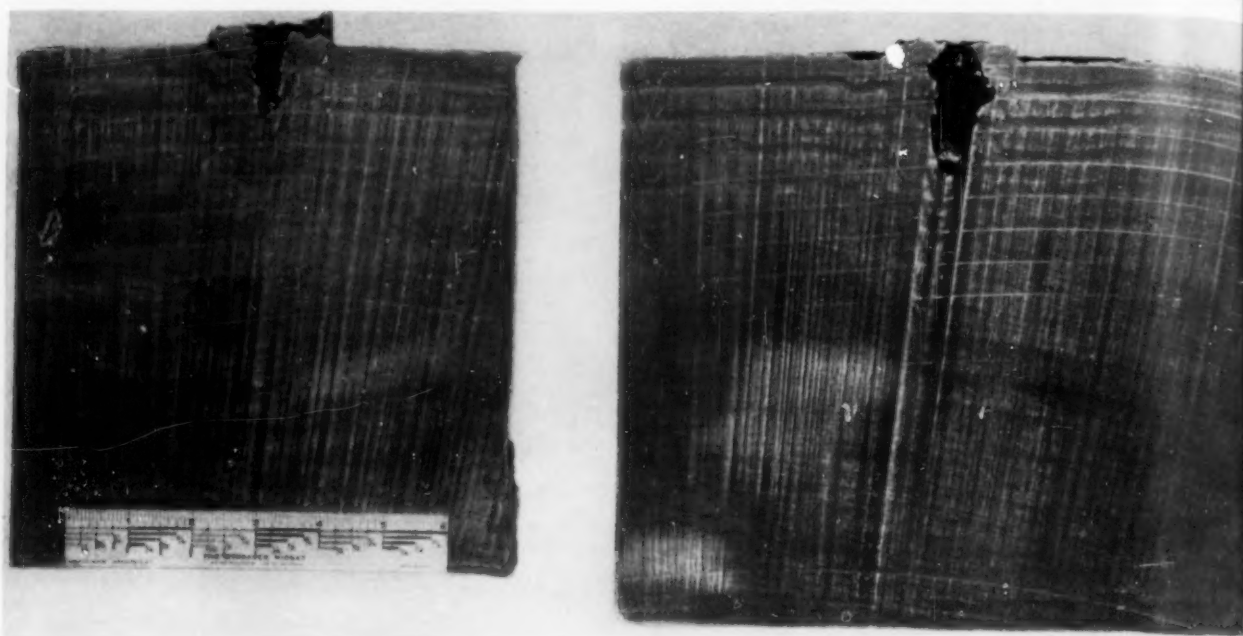


Fig. 16—Showing One 10- and One 12-in. Dia. Billets.

Fig. 17—Summary of Relations between Core Thickness, Neck Size, Casting Size, and Casting Depth. Values on or above the line produce solid castings in shorter or the same length of castings. Necks are round, type A.

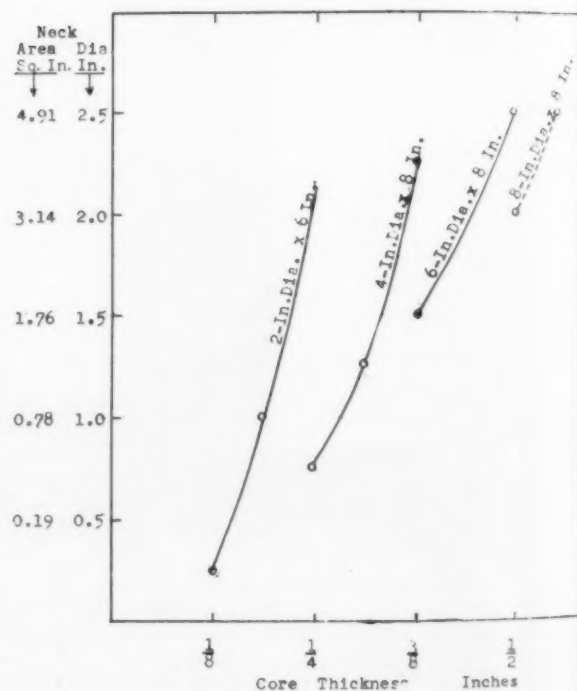
for unrodded cores established in Ref. 9. A later publication of the Steel Founders Society has shown that the size of the opening must be increased if rods are used in the cores, the increase depending on the number and size of the rods, as well as their position in relation to the neck opening. In the case of the larger castings there are practical disadvantages in the use of very thin cores as advocated by the authors. The first is bowing up of the core as shown in Fig. 15, due to softening of the core when heated to the temperature of liquid steel, even though it possesses sufficient strength at lower temperatures.

An even more important disadvantage of thin necked down cores on heavy sections, say over 3 or 4 in., is a type of surface shrinkage. These thin cores become so hot that the metal in contact with the core is the last metal to freeze. Vertical surfaces, when using side, necked down risers are the most prone to this defect and this surface shrinkage can become quite serious. Thicker cores, as thick as 2 in. seem to be the only solution. The openings are, of necessity, quite large, far above those studied by the authors. The practical advantages of cutting and grinding a 25 sq. in. opening in comparison with 314 sq. in. if the 20-in. diam riser was connected directly to the casting is obvious.

Have the authors any theories as to the reason for the variation in shrinkage due to depth of section using necked down cores? Why are the castings made with no necked-down cores not susceptible to this variation due to depth of section?

Mr. BRUNSON: We specifically explained in the introduction that in this paper we were referring to knock-off risers and not neck-down risers.

Whether you use a neck-down riser or not reverts to the principle of directional solidification of the casting or temperature



gradient. To properly feed a casting you have to establish a temperature gradient so that sections that should solidify first will do so, and sections that should solidify last will solidify last whether you have a neck-down riser, a straight riser, or any other kind of riser system.

Some years ago we made a similar test on cast iron. We put a 6-in. diam cylinder on a knock-off riser core, and then an 8-in. riser on top, 6 in. high. After that was poured we had a large shrinkage 1 1/2 in. from the top down. We decided we did not need all that riser on top, so instead of putting an 8-in. riser on

... we put a 3-in. riser on top, thinking we would get a solid casting, but we did not because we had disturbed the temperature gradient.

The neck-down riser would give you a better result and I think this has to be borne in mind in connection with the Washburn cores in that the core is at room temperature and the steel is at some elevated temperature. This core has a tendency to solidify and cool the steel first.

With the cast iron test we found that you have to have a certain section of metal in the riser as compared with the section of the metal in the casting. On the thicker cores you increase your diameter. That gives you more heat to heat up the core and the result is that the middle of that neck has to be the last to freeze.

H. F. BISHOP:<sup>2</sup> I think the reason that you get a shrinkage when you lengthen this casting is because you are changing the surface area to volume ratio. When you have a 4-in. diam, 4-in. high cylinder and lengthen it to 8 in. you are actually doubling the volume but are not doubling the surface area. Actually, the

only way that you can cool a casting is by extracting heat at the sand metal interface and when you do not have as much sand surface to extract heat, naturally, your casting is going to take longer to solidify and you need a thicker neck.

Mr. BRINSON: Maximum volume for the surface is a normal cylinder where the diameter equals the height. So on that basis you disturb the temperature gradient. When you double the volume, the surface area is greater per volume than it is for a normal cylinder.

Mr. BISHOP: In a 4-in. high, 4-in. diam cylinder you have the surface area at the circumference and at the bottom of the casting. If the length, and hence the volume, of this cylinder is doubled you are only doubling the surface area at the bottom of the two castings. In order to double the surface area along with the volume you would have to add additional surface area equivalent to this bottom area.

Mr. BRINSON: You still have the relationship of volume to surface area. The least amount of surface to the contained volume is a cylinder where the diameter equals the height. That fact is taken advantage of by can manufacturers. When they want to use the least amount of tin to get the greatest volume, they produce a can the height of which equals the diameter.

<sup>2</sup>Naval Research Laboratory, Washington, D. C.



# CORE BOX DESIGNING AND RIGGING FOR CORE BLOWING

By

H. J. Jacobson \*

IT IS MANY CENTURIES since man discovered that he could melt metal and cast it into objects to serve his needs. The molder and the patternmaker have come a long way since that time. Today the United States is the greatest industrial nation, and the foundry industry played an important part in making it the great nation it is.

The major part of this remarkable progress has been accomplished in the past thirty years. During that time a great change has taken place throughout this vast industrial empire. The foundry industry, during this period, has changed—it might be said—from a “crude art” into an “exact science.” This “exact science” may be more appropriately termed “mass production technique.”

Machines and tools are gradually taking the place of skilled craftsmen in the production phase of the foundry industry. The blowing of cores is one of the most important developments that has taken place in a great number of years. A machine and a tool, the core blower and the blower core box have taken the place of the craftsman, the core maker. When the engineer designs and rigs a core box for the blower he is building into the core box some of the skill of the craftsman it has replaced.

The basic principle in designing any tool is to have it so well engineered that no one part will wear out faster than another, and it must best serve the purpose for which it was made. Like the proverbial “one horse shay” which was a vehicle so well made that after a century of use it one day collapsed, every part having worn out at exactly the same time.

A core box is a precision tool of the foundry, made to produce accurate cores, from which precision castings are to be made. When a core box is rigged for the core blower, it is being subjected to one of the severest conditions known for any precision tool. It must withstand the pressure of a punch press and the grinding, cutting action of the sand blast.

At the present time there is a great difference of opinion on many questions regarding the design and rigging of core boxes for the blower. Like any other pattern problem it is almost impossible to lay down any standard rules covering this subject. Usually each job presents its own problems.

The purpose of this paper is to arrive at conclusions

\* Industrial Pattern Works, Chicago.

of a general nature regarding the best design, material and technique to be used for a core box of average size and shape. It is hoped that from these conclusions, knowledge will be gained for the use of larger or smaller cores of a similar nature.

The machine on which the core box is to be used must be considered. There are available many makes of core blowers of different designs, but their basic principles are the same. They embody:

1. A reservoir for sand,
2. An inlet for compressed air into this reservoir,
3. An outlet through which sand is blown into the core box.

Its very simplicity is what makes it such a highly efficient part of the modern core room. Production increases with the core blower, over the hand-rammed methods, have been phenomenal. However, in many cases, these increases have been false because expensive core boxes have been completely ruined due to improper design and rigging.

There are two types of machines for blowing small and medium-size cores.

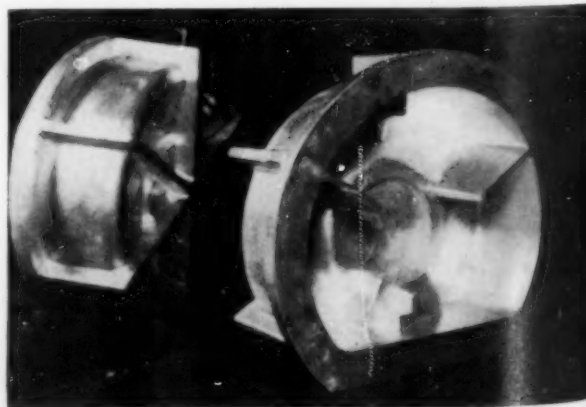
1. The machine designed for vertical split boxes,
2. The machine designed for horizontal split boxes.

## Three Core Box Types

Basically there are three types of core boxes in common use.

1. The dump core box (Fig. 1) which is a solid box, made to load and draw through the same end.
2. The 2- or 3-part split core box (Fig. 2) open on two ends, such as the gang-pin core box, usually with

Fig. 1—Dump core box positioned in front of a mirror to show the back end in reflection.



the parting in a vertical position and cores dried in the on-end position on the flat core plate, and

3. The booked-type core box, or completely enclosed box (Fig. 3) usually blown with the parting in a horizontal position. The name "booked" has been derived from the days of hand ramming where each half of the box was filled and the loaded halves were "booked" together forming a full core (usually run with a core dryer).

This paper will be confined to two general types of machines and three general types of core boxes.

In choosing an example for discussion, the field of foundry mechanization should be considered. Also, a core job should be chosen which would be typical of a great number of core problems that come up daily in the foundry business. The type of job chosen should cover the widest possible scope of the subject to be discussed.

Figures 3 and 4 show the type of core job mentioned, the pattern equipment having been engineered by one of the finest pattern engineering departments in the automotive field. Every detail was carefully studied and the design was blue printed and changes made from time to time as experience with blowing problems made it necessary. The job is being run as a 2-gang completely enclosed or booked-type box with parting in a horizontal position using two lower halves and one upper.

#### Consider Weight and Wear Resistance

The deciding factors in the choice of a core box material are its weight and wearing quality. For the size of box shown in Fig. 3 the choice would be between cast aluminum or cast magnesium. Although magnesium is a lighter metal than aluminum its wearing quality has not been definitely proven under the blower.

The aluminum used by the author is a good grade of pattern aluminum free from pin-hole porosity. Density of the metal is essential to the life of a blower core box because porosity, even if very slight, will give the sand a chance to start a cutting action.

Before any machining is done on the casting, several test cuts are made in the deepest parts of the cavity about  $\frac{1}{16}$  in. below the skin with a free-cutting tool. Usually if the cavity is sound the parting of the casting is sound also. All ingot metal should be used and heavy sections should be chilled. Nitrogen gas has been helpful in preventing gas or pinhole porosity.

The analysis of the aluminum alloy used for this purpose is as follows:

Copper	.....2.5 to 4.0 per cent
Silicon	.....5.0 to 6.5 per cent
Magnesium	.....0.15 per cent max.
Iron	.....1.0 per cent max.
Nickel	.....0.2 per cent max.
Manganese	.....0.5 per cent max.
Zinc	.....0.7 per cent max.

The high silicon makes it possible to pour at comparatively low temperature which makes it ideal for core box and pattern castings.

#### General Shape of Box

The general shape of any core box is controlled by the shape of the cavity or group of cavities. The box should not be made too great in weight, yet it

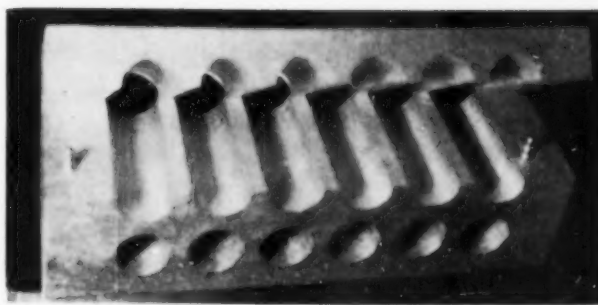


Fig. 2—Three-part split core box open on two ends.

must withstand blower pressure without distortion.

This particular box (Fig. 3) has the main ribbing around the outside back of the box, following the general contour of the cavity. This brings the pressure of the clamping action of the blower close to the cavity edge, assuring a tight seal at the core box parting. Besides this outer ribbing there are two cross ribs in the center. The wall sections are  $\frac{3}{8}$  to  $\frac{7}{16}$  in. thick.

It has been the general thought on small blow boxes to make them square or rectangular in shape, with box-type ribbing, for positioning in the blower. This is important, but much consideration should be given when designing the shape of the box, so that most pressure from the ribbing is applied directly above the cavity outline. It has been proven that this design of ribbing will add greatly to the life of the parting material.

#### Locating Box in Blower

Locating the box in the blower so that the holes in the blow plate line up with the holes in the box can be done in several different ways. The use of stops either on the table under the box or directly on the blow plate seems to be the general practice. Where the cavity shape permits making a box square or rectangular in shape, no additional provisions are necessary on the box.

In this case (Fig. 4) the bottom steel plate, which is  $\frac{3}{16}$  in. thick plus a reinforcing strip of  $\frac{1}{4} \times \frac{3}{4}$  in., is used on the inside right hand corner when the box is placed on the machine. In other words, the stops on the machine strike against this steel bottom plate of the core box which has two edges at a 90-deg. angle on one corner.

For maximum production, facilities must be provided for the operator to handle the box quickly. Here we provide a finger grip on the outside lower edge of the steel bottom plate for the operator to grip and slide box quickly from the blower.

The work table directly in front of the core blower is the same height as the blower table which makes it unnecessary for the operator to lift the box. The operation is "slide in and slide out."

The upper half of the box is made to protrude about  $\frac{1}{2}$  in. on each side, forming convenient finger grips for the operator in drawing upper half from lower half.

The size and number of blow holes and the size and number of vents tie in closely. Theoretically, when blowing a core the operator replaces the air in the cavity with sand of the proper density. The

object is to exhaust the air only through the vents provided and not through the parting of the box or through some other weak point.

#### Provide Ample Vent Areas

The faster the sand enters the cavity, the greater the impact, and therefore, the harder the core. If a book-type box without vents were to be blown, the only place the air could escape would be through the parting. So the theory is that it is much better to "over vent" a core box than to "under vent" for maximum life of parting material.

The upper and lower halves of this core box have an equal number of vents. This box is rigged to give maximum permeability to the sand core. There should be very little built up back pressure through the parting, the air having a free flow through the vent areas. Besides the slotted vents the parting is

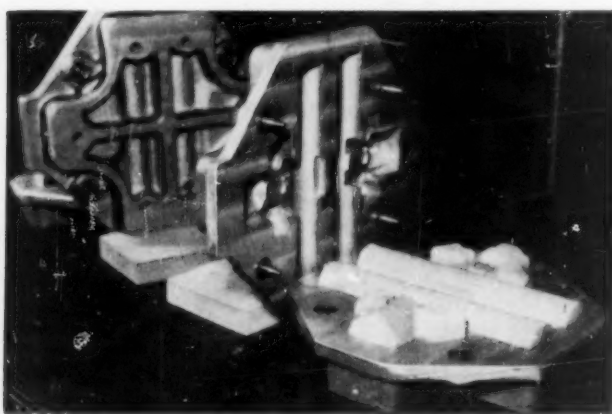


Fig. 3—Booked-type core box positioned in front of a mirror to show the back ribbing in reflection. This is the core box before it is rigged, less vents, facing, blow tubes, and inserts.

completely vented around the entire cavity with  $\frac{3}{32}$ -in. diameter holes  $\frac{1}{2}$  in. apart. This also helps to prevent a leak in the parting material causing the air to be directed through the  $\frac{3}{32}$ -in. holes instead of continuing out through the parting.

The three  $\frac{1}{2}$ -in. blow holes in each cavity equal an area of 0.59 sq. in. The total effective vent area is 2.95 sq. in. Therefore there is a ratio of approximately 5 to 1, exhaust over intake, without considering the air that might vent through the parting.

There are ninety  $\frac{3}{8}$ -in. diameter vents; four  $\frac{1}{2}$ -in. diameter vents; four  $\frac{5}{8}$ -in. diameter vents, and four  $\frac{3}{4}$ -in. diameter vents.

The blow bushings are made of cold-rolled steel, pack hardened. They are made to protrude into the cavity  $\frac{1}{16}$  in. and have a thread in the lower end so the sand in the blow hole will not draw out and become a part of the core.

#### Facing on Partings

The parting on a book-type core box is usually the weakest point and must be replaced several times during the life of the core box. Steel and brass seem to be the materials most preferred for this purpose.

There are three factors to consider when choosing a material for core box facings:

1. Maintain good seal,
2. Edges must stand up,
3. Cost of replacing.

In this particular case (Fig. 3) brass is used because it is easier to replace and bring back to the original cavity line, and it seems to wear as well as steel in this application.

The thickness of the facing should be held to standard dimensions either  $\frac{1}{16}$  (0.062) or  $\frac{1}{8}$  (0.125) in., so that when it is replaced the original dimension of the box will be held. Usually the entire parting face of a core box is covered with facing, but wherever possible, the pressure of the clamping action must seal the box at the cavity edge because if the sand cannot get into the parting, it cannot cut the facing material. In this respect (Fig. 4) a marginal facing is used  $\frac{5}{8}$  in. wide around the entire cavity of the box, directly under the outside ribbing. This will assure uniform pressure around the cavity edge without distorting the box.

None of the facing materials, regardless of hardness or abrasion-resistant qualities, will give satisfactory service unless a perfect seal is maintained during the blowing operation.

#### Steel Inserts Under Blow Tubes

It is almost always necessary on an aluminum or magnesium core box to provide steel inserts or some other material under the blow tubes. The inserts are of a size that takes in the entire lower half of the box up to the facing material and are made to be replaceable when worn out. On short-run jobs the insert can be a small round piece set in of a size large enough to cover the blast area. Thin sheet copper has been used by laying it on the top of the cavity surface under the blast. In the core box shown in Fig. 3 the author uses cold-rolled steel, case hardened.

A blower core box must be kept parallel on the clamping surfaces to insure a good seal on the parting. For this reason steel is used to prevent the box from wearing out of line. The bottom plate on the box must stand the wear of sliding in and out of the blower. The top plate must be heavy enough and supported by ribbing underneath to prevent buckling. It was found that  $\frac{1}{8}$ -in. steel was too weak, hence the present design calls for  $\frac{1}{4}$ -in. plate on top and  $\frac{3}{16}$ -in. plate on the bottom.

There is no blower core box built that will give maximum life and maximum production unless it is maintained. Maintenance of tools and machines is just as important as the designing and rigging. A regular service should be established and the core boxes inspected every 24 hr.

There has been much controversy over the wear resisting quality of one metal over another, and it would be a great advantage to the industry if this question could be settled definitely. An attempt was made to evaluate the blast resisting quality of one material over another.

1. The specimens of materials tested were approximately 4 in. sq. x  $\frac{3}{4}$  in. thick, machined smooth and true on one surface before the wear resistance test.



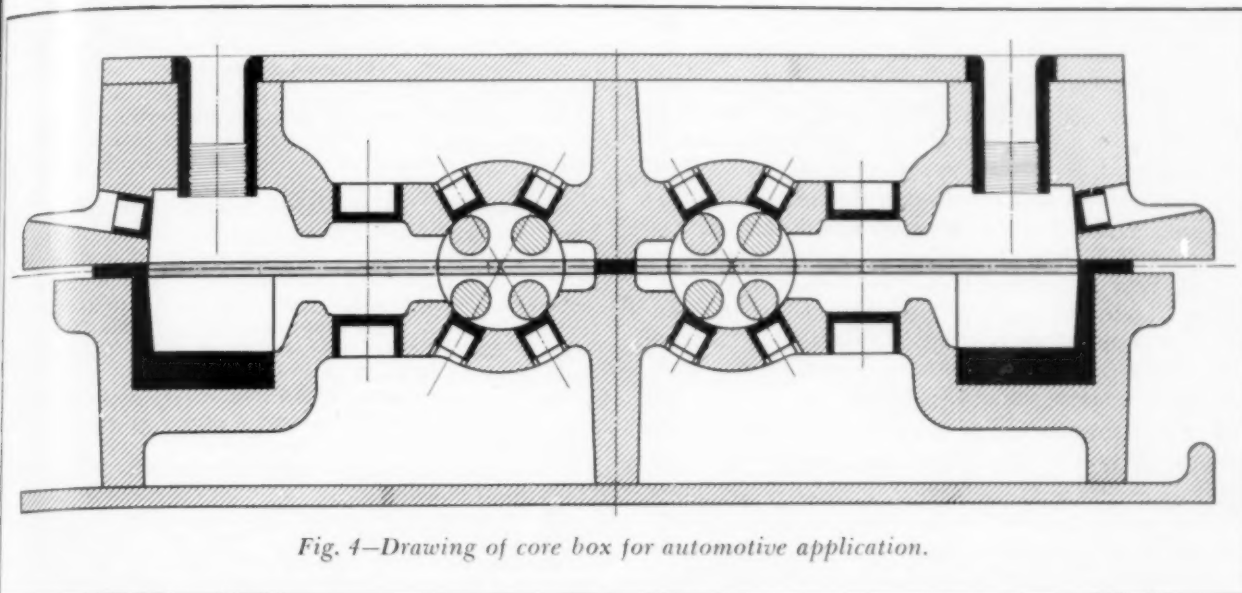


Fig. 4—Drawing of core box for automotive application.

2. A steel plate 4 in. sq. x  $\frac{1}{4}$  in. thick with a 1-in. diameter hole in the center was clamped onto the true machined surface to serve as a masking.

A sandblast machine of the open type, in which none of the sand can get back into the system, supplied all new sand throughout the test.

4. Washed silica core sand No. 60 grit was used under 90-lb pressure through a nozzle opening of  $\frac{1}{8}$  in. diameter directed into the center of the 1-in. diameter opening in the steel plate.

5. The nozzle was clamped into a rigid position and the specimens were all placed 2 in. from the end of the nozzle.

6. Using a stop watch each specimen was given a 2-min. blast at a point near the center and two 1-min. blasts at two other points on the same surface.

7. The pressure gage on the tank and the pressure gage on the sandblast machine were watched to see that the pressure did not vary.

8. The depth of cut was measured with a ball point dial indicator to 0.0001 in. to the lowest point of cut (excluding pits).

9. Depths of cuts of the two 1-min. blasts were added to that of the one 2-min. blast and an average 2-min. reading obtained.

10. Table 1, Column 3, shows the average depth of cut at the lowest point of wear.

11. Column 4 shows the percentage of efficiency of each specimen, using cold-rolled steel as 100 per cent, the other materials rating from this point. This would mean that if a core box made of cold-rolled steel lasted 100 hr, one made of sheet brass or rolled brass would last 76.5 hr.

12. Column 5 shows the condition of the surface of each specimen after the test.

Although this test does not exactly simulate actual conditions of the sand action when blowing a core, it should tell something of the wear resisting quality of one machineable material over another.

It is interesting to note that in both cases where steel was hardened its resistance to abrasion was lessened. Drill rod steel drops from 145 to 50.6 per cent

and cold-rolled steel drops from 100 to 63.7 per cent after hardening. Samples of cold-rolled steel were tested after cyanide hardening and salt bath carburizing with almost the same results.

The first test was made with cast iron comparable to the iron that would be purchased by a pattern shop for a core box casting. Because of the comparatively poor rating of this particular iron, a piece of close grain inoculated iron was used. This specimen

TABLE 1—BLAST RESISTANCE TEST OF CORE BOX MATERIALS

Item No.	Material	Average Depth of Cut, In.	Per Cent of Efficiency	Condition of Surface
1.	Drill Rod	0.0029	145.	Very smooth
2.	Copper (Hard drawn)	0.0031	135.	Very smooth
3.	Cold Rolled Steel (flat)	0.0042	100.	Very smooth
4.	Brass (Sheet)	0.0055	76.5	Very smooth
5.	Aluminum (Sheet)	0.0060	70.	Slight roughness (peened effect)
6.	Cold Rolled Steel (Hard)	0.0066	63.7	Very smooth
7.	Brass (85-5-5-5)	0.0083	50.6	Slightly rougher than No. 5
8.	Drill Rod (Hard)	0.0083	50.6	Very smooth
9.	Aluminum (Chill Cast)	0.0095	44.2	Smooth
10.	Inoculated Cast Iron	0.0099	42.5	Same roughness as No. 7
11.	Aluminum (Not Chilled)	0.0102	41.2	Pitted, $\frac{1}{64}$ holes, $\frac{1}{16}$ to $\frac{3}{16}$ apart
12.	Magnesium (Sheet)	0.0120	35.	Very smooth
13.	Cast Iron	0.0130	32.3	Very rough
14.	Magnesium	0.0210	20.	Smooth
15.	Lead	0.0410	10.2	Rough

LIGHT METALS COMPARISON			
1.	Aluminum (Sheet)	0.0060	158.
2.	Aluminum (Chill Cast)	0.0095	100.
3.	Aluminum (Not chilled)	0.0102	93.
4.	Magnesium (Sheet)	0.0120	79.
5.	Magnesium (Cast)	0.0210	45.2

tested slightly under aluminum (chill cast), and the aluminum specimen had a much better surface.

A gang-pin core box made of rolled aluminum stock rated 70 per cent would outlast the best iron box rated at 42.5 per cent. Due to the difference in material weights, it is possible to make the aluminum box in a larger size with a greater number of cavities.

It will be noted that in all cases rolled material has a better rating than cast material. Rolled brass rated 76.5 per cent against cast brass (85-5-5-5) 50.6 per cent. Rolled aluminum rated 70 per cent against the best cast aluminum of 44 per cent. Cold-rolled steel rated 100 per cent. Rolled magnesium rated 35 per cent, whereas cast magnesium rated 20 per cent.

The cast magnesium does not rate as high as aluminum. However this would not necessarily mean that magnesium is not a good core box metal. When the need is for large core boxes where weight is an important factor and where the blast is not too close to the opposite side of the cavity, or where provisions can be made to protect the blast area, magnesium could still be considered a good core box material.

### Conclusions

1. Complete cooperation between pattern shop and foundry so that design can be predetermined at blue print stage.
2. Density of the core box material is important.
3. Design of ribbing and general shape of box should be given careful consideration.
4. For maximum production consideration should be given as to whether it is necessary to provide special features to enable operator to handle box.
5. Proper location and size of blow holes and vents should be recognized as a means of prolonging the life of a blower core box.
6. Consideration should be given to the advisability of using brass or copper as a parting material.
7. When using light metals, provisions should be made to protect the area under the blow hole by a means which would be in direct proportion to the production requirements expected from the core box.
8. Machines and tools must be kept in first class condition for maximum production.

### DISCUSSION

Chairman: A. F. PFEIFFER, Allis-Chalmers Mfg. Co., Milwaukee.

Co-Chairman: L. F. TUCKER, City Pattern & Foundry Co., South Bend, Ind.

DANIEL BYRNE<sup>1</sup>: Did the author have experience with plastic or rubber core boxes? The only advantage that I can see in using plastics is that a new core box can be made in a couple of hours, just by putting the plastic in the box.

MR. JACOBSON: Yes, that is the thought in using plastics. I have heard that the plastic seems to stand up under the blower, but you must use a plastic that can be tooled or machined. We have to put dowel pins and vents in it; and if the plastic cannot be handled or tooled, then it would not be practical to use.

W. J. FINSETH<sup>2</sup>: Table 3 shows that the author used aluminum and magnesium experimentally for core boxes. What was the composition of these alloys?

MR. JACOBSON: The aluminum alloy was the same alloy we use for pattern castings. The analysis is shown on page 603. The magnesium used was alloy C and alloy H.

W. H. RUTEN<sup>3</sup>: What is the life of the brass facing on the joint of the box?

MR. JACOBSON: With proper maintenance you can get about 40,000 cores from one brass facing. Those boxes are taken out of production every 24 hr and carefully inspected. If there is any indication of a leak they are repaired immediately.

MR. RUTEN: On the matter of brass facing, I have concluded from the author's remarks that the joint is to be entirely air tight. What would happen if the joint did leak slightly? There are numerous vents in the box. Would a little additional venting at the joint make much difference?

MR. JACOBSON: The parting is the weakest point in a core box for core blowing. When a leak once starts at the parting it progresses rapidly due to the abrasive, cutting action of the sand blown into the core box. You cannot allow a leak to start in a core box, and the only way you can do that is through proper inspection and maintenance.

MEMBER: We had occasion to make several alloy iron core boxes, and we have abandoned aluminum almost entirely for core blowing. We find that by cutting the metal section down and ribbing properly it works out satisfactorily and the maintenance on them is negligible.

MR. JACOBSON: Gray iron is a good material for core boxes. However a gray iron box would weigh considerably more.

A. REYBURN<sup>4</sup>: Does the author have experience in putting rubber inserts around the edge of the box to prevent leakage?

MR. JACOBSON: No I have not. I understand it has not been successful.

F. W. SHIPLEY<sup>5</sup>: Does the author have experience with neoprene blow holes? We have an idea that it will not only prevent wear of the blowholes themselves but it will pinch off the sand so that the sand will not have to be knocked off the top of the core.

MR. JACOBSON: No we have no experience in using neoprene for this purpose.

<sup>1</sup> W. D. Allen Mfg. Co., Bellwood, Ill.

<sup>2</sup> Light Alloys Ltd., Renfrew, Ontario, Canada.

<sup>3</sup> Polytechnic Institute of Brooklyn, Brooklyn.

<sup>4</sup> Massey-Harris Co., Ltd., Brantford, Ontario, Canada.

<sup>5</sup> Caterpillar Tractor Co., Peoria, Ill.

# AN APPROACH TO QUALITY CONTROL IN CASTINGS PRODUCTION

By

V. A. Simpson \* and G. K. Eggleston \*\*

THE PURPOSE OF THIS PAPER is to present a new approach to the often distressing problem of maintaining cooperation between the production, metallurgical and inspection personnel, in their combined attempt to keep quality up and costs down.

While the authors' plant makes rough and machined castings on a jobbing basis, the bulk of the castings is used as components of items manufactured in the plant. Only the castings used in the plant's own products will be discussed.

The authors' interpretation of quality control is defined as the necessary inspections or process controls to produce a product to predetermined chemical, physical, dimensional and visual standards at the lowest cost consistent with the standards set. A large contributing factor to low costs is the keeping of rejects to a minimum.

## Quality Control Responsibility

Possibly the most important consideration in setting up a quality control program is determining to whom the quality control personnel shall be responsible. This subject can become controversial. However, the authors made the quality control responsible to the chief engineer, reasoning that the engineering department designs the products and should have some means of assurance that the products they have designed will function as intended when they reach their customers.

Therefore, the man heading up the quality control is assistant chief engineer responsible only to the chief engineer. The quality control man has absolutely no authority over production personnel. Production personnel are held responsible for the quality of the parts produced by their departments and the maintenance of all equipment and processes in the department. It is the quality control's responsibility to supply the production departments with all tests, analyses and other technical information they may need in their operation, such as alloy analyses, complete sand analyses for both foundry and core room, solution analyses for electroplating and the procurement and maintenance of all gages and testing equipment.

Production supervisors are free to make any changes in their manufacturing procedures or processes whenever they feel it will reduce costs. However, the change must be submitted to the quality control man in

writing. He will then make the necessary changes in engineering records. This insures against unauthorized changes being made.

On the other hand, quality control does have absolute authority over the quality of work the production departments produce. This authority expresses itself in two ways; first, it is their responsibility to detect and reject all defective castings. These are sorted according to defect or cause of rejection and charged against the department responsible for the rejection, such as foundry, machine shop, polishing, plating or assembly. A casting rejected on the assembly line because of foundry fault is charged to the foundry, not at the value of the original castings, but with all accumulated manufacturing costs up to the time it is scrapped. Scrap reports are turned in every day so an error or carelessness on the part of a production supervisor or his men is immediately observed.

The second and more drastic means the quality control has at its disposal is the authority to stop production on an item in any stage of manufacture, if the rejects become excessive. This usually brings corrective results immediately as almost without exception, it happens on a job that is being expedited.

The quality control man, unless he has tact and is a fair student of psychology, can become disliked, "at least during working hours," instead of being an important spoke in a smooth running wheel of his manufacturing business.

## Control Is Loss Preventive

Setting up a quality control program and getting it into operation may meet with considerable opposition from production personnel. This is especially true after the production man is called to account for excessive scrap or shut down. But if he is open-minded, he will realize that quality control can keep him out of more trouble than it will get him into.

Quality control, if it is to mean *quality* and *control*, must, of necessity, have consideration in the very beginning of a product's design. Quality control at the authors' plant starts before the drawings leave the drafting boards in the engineering department. If a new casting is required, the drawing and design must be approved by three people; the engineer in charge of design, the engineer in charge of quality control and the chief engineer. In this way the quality control man is given the opportunity to question finish allowance or dimensional tolerance he is to maintain

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when the casting gets into production. It is equally costly to maintain unnecessarily close tolerances as it is to slight those that should, of necessity, be close. Therefore, it is important to establish limits that are real and practical.

### Standards and Process Controls

After establishing the standards to which the casting is to be manufactured, the inspection or process controls must be set up in every department through which the pieces must pass, from the foundry to the shipping department. To accomplish these inspection and process controls, the quality control man has at his disposal the chemical laboratory, foundry control laboratory, physical testing laboratory, reports on all process data, and departmental inspectors.

Simplicity is the key to a workable plan, which must, in most part, be executed by personnel in the shop who are not required to have technical skill. It is the job of quality control to collect the facts on a casting as it goes through the factory until it is a completed part of an assembly. In this way a history can be had on which to base design, minimizing past trouble in patterns, tooling, finishing and even electroplating. After the drawings have cleared the engineering department, the next step is to check the pattern. No pattern is permitted to be put into production until a rough casting made therefrom has been checked, machined and rechecked, by quality control.

An effort is made to get the molders and melters to follow the same procedure from one hour to the next. It has been found that posting each molder's daily production and scrap on a large and conspicuously located bulletin board in the foundry has injected a spirit of competition among the molders, not only in quantity or production but in the percentage of scrap produced. It also has the effect of forcing the plant to keep pattern and core equipment in top condition as the molder will complain if he thinks his scrap was high because of faulty patterns or cores.

Large dial type recording pyrometers are used for checking pouring temperatures. They also serve the purpose of recording the temperature of each pot of metal poured. It is remarkable how consistently some operators can keep the temperature of the metal they melt, and exasperating how far others can miss it.

Let us retrace the services the quality control performs in and for the foundry. The foundry laboratory takes care of the necessary metal analyses, test bars or other physical tests when required, as well as all phases of sand testing both for the foundry and the core room. The foundry patrol inspector is depended upon for the source of most of the information to run down and stop scrap in the foundry. He checks the foundry scrap daily and posts it on the bulletin board. Every pattern plate and each casting on a plate carries a control number. A record is kept showing the amount of scrap and good castings produced daily from each pattern plate.

### Accurate Records Essential

The records make it possible to check patterns, plate and molder separately. Often, just by switching patterns or plates, it is possible to find whether the molder

or the pattern or the individual plate is at fault. Castings found defective anywhere along the line can be traced back to the plate, pattern, molder and floor.

These details aid in clearing up poor production practices. When a man sees his record is not as good as that of another, he will make an effort to do better.

Ideas of individuals, innocently put into practice, will often create an upsetting influence. The patrol inspector makes it his business to fill in his report completely, including such things as changed angle of pour-off racks, venting procedure changed, vents changed on flask, adding water to mold facing or failing to use facing, and many others. Also, the weather and temperature are recorded to help settle melting time and core blow disputes. Cores absorb moisture in damp weather, and exceptionally cold weather will change time required for melting.

The problem in the cleaning room is simple. It involves checking for proper cleaning, over or under-grinding and the removal of core wires, which can cause serious tool breakage in the machine shop.

A similar approach is used in tracing down trouble in the machine shop and on the assembly line. The patrol inspector fills out a daily form recording all tool changes, time down, reason for being down, operator and machine number. At the end of the day, the inspector and the foreman sign the report.

The department scrap is marked and recorded by the man who rejects the castings. These reports from each department are correlated into one daily sheet. The scrap tag is stamped, showing in which department the scrap was produced.

### Daily Production Samples Finished

Machining and polishing are done from current daily stock as much as possible. For instance, if the foundry makes 800 pieces per day, and it is possible to machine or polish only 500 pieces, 300 are put in stock. In this way a large sample of each day's production is obtained. This is only fair to foundry and causes very little extra work. This procedure was instituted after several experiences with running large quantities of rough castings which went into stock and later had to be scrapped because something had gone wrong in the foundry.

On castings that are to be polished and electroplated, it is the practice to put a small sample of each day's foundry run through the polishing operation even if the machining has to be skipped, as the delay between foundry and polishing is too great.

The patrol inspector in the foundry is the key man for he is familiar with all patterns, plates, cores, molders and floor conditions. He is also familiar with rejects and many of their causes. He correlates all the daily reports and records the scrap. The monthly summary sheets are also filled in by him. They give an overall picture of the scrap condition and become an invaluable historical record against which the plant can judge progress or lack of progress.

In conclusion, it is felt that the quality control system outlined in this paper has proved profitable for the company in reducing costs by the reduction of scrap and has practically eliminated the return of material from the field.

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